

Beam Shaping in Laser-Based Powder Bed Fusion of Metals

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Preface

This doctoral thesis originated during my work as a research associate and research group leader at the Professorship of Laser-based Additive Manufacturing (LBAM) from 2020 to 2025. These years were an extraordinary journey marked by challenges, discoveries, and countless moments of learning that shaped both this work and me personally.

First, I want to thank Prof. Dr.-Ing. Katrin Wudy. She gave me the opportunity to pursue my doctorate in her newly established unit, affording me a remarkable amount of creative freedom. Her great ambitions and inspiring “whatever it takes” conviction gave the doctoral period incredible drive. Additionally, I would like to thank Prof. Adam Thomas Clare, Ph.D., and Prof. Dr. techn. Jan Torgersen for acting as the co-examiner. Furthermore, I want to thank Prof. Dr.-Ing. Christoph Meier for chairing the examination board.

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Abstract

Additive manufacturing is becoming increasingly established as a new class of industrial manufacturing processes. Thereby, it enables the production of three-dimensional component geometries that cannot be manufactured using conventional manufacturing methods such as machining. Nowadays, laser-based powder bed fusion of metals (PBF-LB/M) is the most industrially mature additive manufacturing process. However, PBF-LB/M still faces numerous challenges, including long manufacturing times, narrow process windows, and limited utilization of potential process freedoms, such as microstructure tailoring. In the state of the art, these are partly attributed to the Gaussian laser spots commonly used in PBF-LB/M.

To tackle these challenges, this doctoral thesis addresses the application of ring-shaped beam profiles in PBF-LB/M. The aim is to gain *a fundamental understanding of the relationship between beam profile characteristics, a modeled process signature, and the dynamic process behavior, as well as the resulting component properties.*

For the experimental implementation, a programmable laser beam source with switchable ring-shaped beam profiles is used to process the material AISI 316L. The focus on ring-shaped beam profiles is chosen, as this type of beam profile can be implemented in PBF-LB/M machines with comparatively little integration effort. Thus, a beam shaping technology is used that has one of the highest technology readiness levels and can therefore be used in future industrial applications of additive manufacturing.

The research findings are derived from a hierarchical approach that can be structured into five solution modules, each represented by a peer-reviewed journal paper. An initial feasibility study shows that ring-shaped beam profiles enlarge the process window and potentially enable higher process speeds by modifying the melt track dimensions and surface topology. This initial study uses input-output correlations. To overcome this black box approach often encountered in the state of the art, the second solution module creates a model that enables the comparability of processes with different beam shapes. For this purpose, a semi-analytical heat conduction model is employed to predict the melt pool geometry, validated using the width and depth of single melt tracks. In the third solution module, this heat conduction model is extended to the three-dimensional process by considering the rapid scanning of hatched scan vectors. Beam profile- and size-independent process maps can be created using the parameters employed in this extended model. Within these new process maps, the introduced semi-analytical model can be used to predict process window boundaries by evaluating geometric characteristics of the modeled melt pools. This new approach forms the basis for ensuring compatibility of processes using

different beam profiles across various sizes by providing a modeled process signature. The simplified semi-analytical model approach does not provide any information about the process dynamics. Nor can the component properties be derived directly from the model. Therefore, the influence of ring-shaped beam profiles on the spatter formation is investigated in the fourth solution module. In the fifth solution module, the resulting microstructure is examined. For both studies, the model parameters from the second and third solution modules serve as comparative parameters. Key outcomes are that beam profiles with a low peak-to-mean intensity ratio tend to result in processes with a reduced amount and size of spatter. This indicates an increased process reliability for ring-shaped beam profiles. Moreover, the beam profile characteristics have a significant influence on the resulting microstructure. State-of-the-art processing strategies using Gaussian beam profiles typically produce microstructures with little texture or with $\langle 100 \rangle$ or $\langle 110 \rangle$ textures in the build direction in PBF-LB/M. Using ring-shaped beam profiles enables the generation of $\langle 111 \rangle$ textures in the build direction. Furthermore, it is demonstrated how the microstructure can be locally adapted by using different beam profiles for various segments within one component.

The holistic approach of the doctoral thesis, which first creates a modeled process signature and then utilizes the parameters employed for this purpose to compare different processes with varying beam profiles and sizes, represents a key novelty compared to the state of the art. This enables the understanding of the relationship between beam characteristics and the resulting process and component properties, in contrast to the frequently used black-box approaches. The work thus represents an important contribution to understanding the influence of beam shaping in PBF-LB/M, as well as in other laser-based melting processes.

The models presented, along with the demonstrated influences on process and component properties, introduce tools that have the potential to reduce the effort required for future experimental investigations by providing model-based estimates of the impact of new beam profiles. Furthermore, the findings can be exploited industrially with minimal implementation effort for process optimizations and enhancements, such as microstructure tailoring.

Kurzfassung

Die additive Fertigung etabliert sich zunehmend als neue Klasse industrieller Fertigungsverfahren. Sie ermöglicht die Herstellung dreidimensionaler Bauteilgeometrien, die mit herkömmlichen Fertigungsverfahren wie zerspanenden Verfahren nicht herstellbar sind. Heute ist das laser- und pulverbettbasierte Schmelzen von Metallen (PBF-LB/M) das industriell ausgereifteste additive Fertigungsverfahren. Allerdings stehen PBF-LB/M noch zahlreiche Herausforderungen gegenüber, darunter lange Fertigungszeiten, kleine Prozessfenster und eine begrenzte Ausnutzung potenzieller Prozessfreiheiten, etwa der gezielten Mikrostrukturanpassung. Im Stand der Technik werden diese teilweise auf die bei PBF-LB/M üblicherweise verwendeten gaußförmigen Laserspots zurückgeführt.

Um die beschriebenen Herausforderungen anzugehen, untersucht diese Dissertation die Anwendung ringförmiger Strahlprofile im PBF-LB/M. Das übergeordnete Ziel besteht darin, *ein grundlegendes Verständnis der Beziehung zwischen den Eigenschaften eines Strahlprofils, einem semi-analytischen Modell der Prozesssignatur und dem dynamischen Prozessverhalten sowie den resultierenden Bauteileigenschaften* zu gewinnen.

Für die experimentelle Umsetzung wird eine programmierbare Laserstrahlquelle mit umschaltbaren ringförmigen Strahlprofilen verwendet, um den Werkstoff AISI 316L zu verarbeiten. Der Fokus liegt auf ringförmigen Strahlprofilen, da diese Art von Strahlprofilen mit vergleichsweise geringem Integrationsaufwand in PBF-LB/M-Maschinen implementiert werden kann. Damit wird eine Strahlformungstechnologie genutzt, die einen der höchsten Technologiereifegrade aufweist und entsprechend in zukünftigen industriellen Anwendungen der additiven Fertigung eingesetzt werden kann.

Die Forschungsergebnisse basieren auf einem hierarchischen Ansatz, der sich in fünf Lösungsmodule unterteilen lässt, die jeweils durch einen begutachteten Fachartikel repräsentiert werden. Eine erste Machbarkeitsstudie zeigt, dass ringförmige Strahlprofile durch die Veränderung der Schmelzspurabmessungen und -oberflächentopologie das Prozessfenster vergrößern und potenziell höhere Prozessgeschwindigkeiten ermöglichen. Diese erste Studie verwendet Eingabe-Ausgabe-Zusammenhänge. Um diesen häufig im Stand der Technik verwendeten Black-Box-Ansatz zu überwinden, wird im zweiten Lösungsmodul ein Modell implementiert, das die Vergleichbarkeit von Prozessen mit unterschiedlichen Strahlformen ermöglicht. Zu diesem Zweck wird ein semi-analytisches Wärmeleitungsmodell verwendet, um die Geometrie des Schmelzbades vorherzusagen. Das Modell wird anhand der Breite und Tiefe einzelner Schmelzspuren validiert. Im dritten Lösungsmodul wird dieses Wärmeleitungsmodell auf den dreidimensionalen Prozess erweitert, indem das schnelle Abfahren benachbarter Belichtungsvektoren berücksichtigt

wird. Mit den in diesem erweiterten Modell verwendeten Parametern lassen sich strahlprofil- und spotgrößenunabhängige Prozesskarten erstellen. Innerhalb dieser neuen Prozesskarten kann das semi-analytische Modell verwendet werden, um die Grenzen des Prozessfensters anhand geometrischer Eigenschaften der modellierten Schmelzbäder vorherzusagen. Dieser neue Ansatz bildet die Grundlage, die Vergleichbarkeit von Prozessen mit unterschiedlichen Strahlprofilen über verschiedene Strahlgrößen hinweg zu ermöglichen, indem eine modellierte Prozesssignatur bereitgestellt wird. Der vereinfachte Modellansatz liefert allerdings keine Informationen zur Prozessdynamik. Auch lassen sich die Bauteileigenschaften nicht direkt aus dem Modell ableiten. Daher wird im vierten Lösungsmodul der Einfluss ringförmiger Strahlprofile auf die Spritzerbildung analysiert. Im fünften Lösungsmodul wird die resultierende Mikrostruktur untersucht. Für beide Untersuchungen dienen die Modellparameter aus den zweiten und dritten Lösungsmodulen als Vergleichsparameter. Die Kernergebnisse zeigen, dass Strahlprofile mit einem niedrigen Verhältnis von Spitzen- zu mittlerer Intensität tendenziell zu Prozessen mit geringerer Spritzermenge und -größe führen. Das deutet auf eine erhöhte Prozessstabilität bei ringförmigen Strahlprofilen hin. Darüber hinaus beeinflussen die Eigenschaften des Strahlprofils die resultierende Mikrostruktur signifikant. Standardbelichtungsstrategien mit gaußförmigen Strahlprofilen erzeugen im PBF-LB/M in der Regel Mikrostrukturen mit geringer Textur oder mit $\langle 100 \rangle$ - oder $\langle 110 \rangle$ -Texturen in Aufbaurichtung. Die Verwendung ringförmiger Strahlprofile ermöglicht die Erzeugung von $\langle 111 \rangle$ -Texturen in Aufbaurichtung. Darüber hinaus wird gezeigt, wie die Mikrostruktur lokal angepasst werden kann, indem unterschiedliche Strahlprofile für verschiedene Segmente eines Bauteils eingesetzt werden.

Der ganzheitliche Ansatz der Dissertation, bei dem zunächst eine modellierte Prozesssignatur erstellt wird und anschließend die dafür verwendeten Parameter zum Vergleich von Prozessen mit verschiedenen Strahlprofilen und -größen herangezogen werden, stellt eine wesentliche Neuheit gegenüber dem Stand der Technik dar. Dies ermöglicht, im Gegensatz zu den häufig verwendeten Black-Box-Ansätzen, ein Verständnis der Beziehung zwischen den Strahleigenschaften und den daraus resultierenden Prozess- und Bauteileigenschaften. Dadurch leistet die Arbeit einen wichtigen Beitrag zum Verständnis des Einflusses der Strahlformung im PBF-LB/M sowie in anderen laserbasierten Schmelzprozessen.

Die vorgestellten Modelle sowie die nachgewiesenen Einflüsse auf Prozess- und Bauteileigenschaften stellen Werkzeuge dar, die das Potenzial haben, den Aufwand für zukünftige experimentelle Untersuchungen zu reduzieren, indem sie modellbasiert den Einfluss neuer Strahlprofile abschätzen. Außerdem können die Ergebnisse mit geringem Implementierungsaufwand industriell für Prozessoptimierungen und neue Prozessfreiheitsgrade eingesetzt werden, wie z. B. der lokalen Anpassung der Mikrostruktur.

Nomenclature and abbreviations

List of symbols

Symbol	Description
A	Absorption
$A_{\text{intersected}}$	Area of grains intersected by the beam profile switching plane
$A_{\text{intersected},\%}$	Area fraction of grains intersected by the beam profile switching plane
A_{new}	Area of newly emerged grains
c_p	Specific heat capacity
$d_{10,3}$	Volume-based 10 th percentile powder particle diameter
$d_{50,3}$	Volume-based 50 th percentile powder particle diameter
$d_{90,3}$	Volume-based 90 th percentile powder particle diameter
d^*	Dimensionless melt pool depth
d_h^*	Dimensionless melt pool depth at the half hatch
ΔE	Energy introduced
E^*	Dimensionless energy input
E_V	Volume energy density
g	Dimensionless temperature field
G	Temperature gradient
h	Hatch distance
h^*	Dimensionless hatch distance
h_s	Enthalpy at melting
ΔH	Enthalpy introduced
$\Delta H/h_s$	Dimensionless enthalpy
$\Delta H_{\text{max}}/h_s$	Maximum dimensionless enthalpy
I_{mean}	Mean intensity
I_{peak}	Peak intensity
k	Thermal conductivity
l^*	Dimensionless melt pool length
L_f	Latent heat of fusion
M	Magnification
O_{relative}	Relative overlap
P_{laser}	Laser power
P_{laser}^*	Dimensionless laser power
Pe	Péclet number
Pe_{max}	Maximum Péclet number

Symbol	Description
Pe_{mean}	Mean Péclet number
Pe_{min}	Minimum Péclet number
$r_{50,2}$	Area-weighted equivalent grain radius
R	Growth rate
S_a	Arithmetical mean height
t	Integration variable
T	Temperature
T_m	Melt temperature
T_{peak}^*	Dimensionless peak temperature
T_0	Ambient temperature
v_{scan}	Scanning speed
V	Volume
w	Melt pool width
w^*	Dimensionless melt pool width
w_0	Spot radius
x, y, z	Spatial coordinates
x^*, y^*, z^*	Dimensionless spatial coordinates
$\bar{x}$	Scan vector length
$\bar{x}^*$	Dimensionless scan vector length
α	Thermal diffusivity
γ	Surface tension
δ_{diff}	Thermal diffusion depth
δ_{layer}	Layer thickness
δ_{layer}^*	Dimensionless layer thickness
ρ	Material density
τ	Interaction time
φ_{rot}	Layer rotation angle
ψ	Peak-to-mean intensity ratio

List of abbreviations

Abbreviation	Description
AOD	Acousto-optical deflectors
BD	Build direction
cw	Continuous wave
DOE	Diffraction optical element
EBSD	Electron backscatter diffraction
fcc	Face-centered cubic
FWHM	Full width at half maximum
HIP	Hot isostatic pressing
IPF	Inverse pole figure
LoG	Laplacian of Gaussian
PBF-EB/M	Electron beam-based powder bed fusion of metals
PBF-LB/M	Laser-based powder bed fusion of metals
RQ	Research question
SLM	Spatial light modulator
SM	Solution module

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1 Introduction

Headlines such as "3D Printing Is Spurring a Manufacturing Revolution" (New York Times [1]) or "3D Metal Printing: A Revolution in the Factory" (original "3D-Metalldruck: Eine Revolution in der Fabrik", FAZ [2]) recognize that additive manufacturing holds great potential as an industrial manufacturing process. Nevertheless, the overall market share of additive manufacturing processes is relatively low. The global market size of additive manufacturing was estimated at USD 21.58 billion in 2024 [3]. With a global manufacturing market size of USD 14.16 trillion in 2024 [4], this represents a share of 0.15 % for additive manufacturing processes. One reason for this is that the potential freedoms of additive manufacturing processes, such as local microstructure tailoring, have yet to be fully exploited. Limited process reliability and productivity exemplify the challenges preventing further industrialization of additive manufacturing processes.

With a market share in systems of approximately 80 %, laser-based powder bed fusion of metals (PBF-LB/M according to DIN EN ISO/ASTM 52900) has the greatest industrial maturity in metal-based additive manufacturing [5]. With PBF-LB/M, three-dimensional components are produced layer by layer by selective fusing of powder material using a laser beam. Due to the layer-by-layer process sequence, workpieces with complex geometries that exceed the capabilities of conventional manufacturing methods can be produced.

However, laser exposure causes highly complex physical phenomena in the created melt pool, resulting in high process dynamics. In state-of-the-art PBF-LB/M, lasers with Gaussian intensity distributions are used [6]. These Gaussian laser beams are focused on microscopic spatial scales (in a two- to low three-digit micrometer range [7]) and deflected at high scanning velocities (up to single-digit m/s range). This results in very short beam-material interaction times, ranging from two to three digits of microseconds. Even minor disturbances or instabilities can lead to process irregularities and component defects that impair reproducibility and productivity in these small scales. In addition to the process and melt pool dynamics, the global and local temperature distributions determine the cooling conditions, solidification, and, thus, the microstructure and resulting component properties.

The temperature distribution and its progression are determined by the material properties, the scanning strategy, and the laser beam profile. Consequently, one opportunity to modify these process influences is to adapt the beam profile by using beam shaping techniques. Therefore, beam shaping and the influence of the adapted beam profiles on laser material processing are current research interests, particularly with respect to PBF-LB/M. However, the state of the art largely provides input-output correlations from a black-

box perspective. In contrast, this thesis aims to gain *a fundamental understanding of the relationship between beam profile characteristics, a modeled process signature, and the dynamic process behavior, as well as the resulting component properties* in PBF-LB/M processing the exemplary material AISI 316L. To achieve this, the thesis is structured as outlined below:

First, the “State of the art” section presents the fundamentals of the PBF-LB/M process, the basics of beam shaping, the application of beam shaping in laser-based melting processes, and the fundamentals of various simplified laser material processing models. Subsequently, five research questions and solution modules arising from the state of the art are presented and interconnected in the “Research approach” section. This is followed by the “Materials and methods” section, which describes the feedstock material and the experimental boundary conditions. The findings obtained from the five solution modules represented by five peer-reviewed journal publications are summarized in the “Research findings” section. Finally, a concluding discussion with respect to the state of the art is presented in the section “Critical discussion and conclusion“. Additionally, the Appendix includes reprints of all the included publications.

2 State of the art

This section introduces the laser-based powder bed fusion of metals (PBF-LB/M according to DIN EN ISO/ASTM 52900) process, its characteristics, typical process limits, and resulting microstructures. Subsequently, the basics of beam shaping and its application in laser-based melting processes are presented. The section closes by discussing various simplified process models for laser material processing.

2.1 Laser-based powder bed fusion of metals

PBF-LB/M is a single-step additive manufacturing process from the process category of powder bed fusion processes with metallic feedstock materials. A laser beam is used as the energy source for selectively fusing the material in a melted state within a powder bed (see Figure 1).

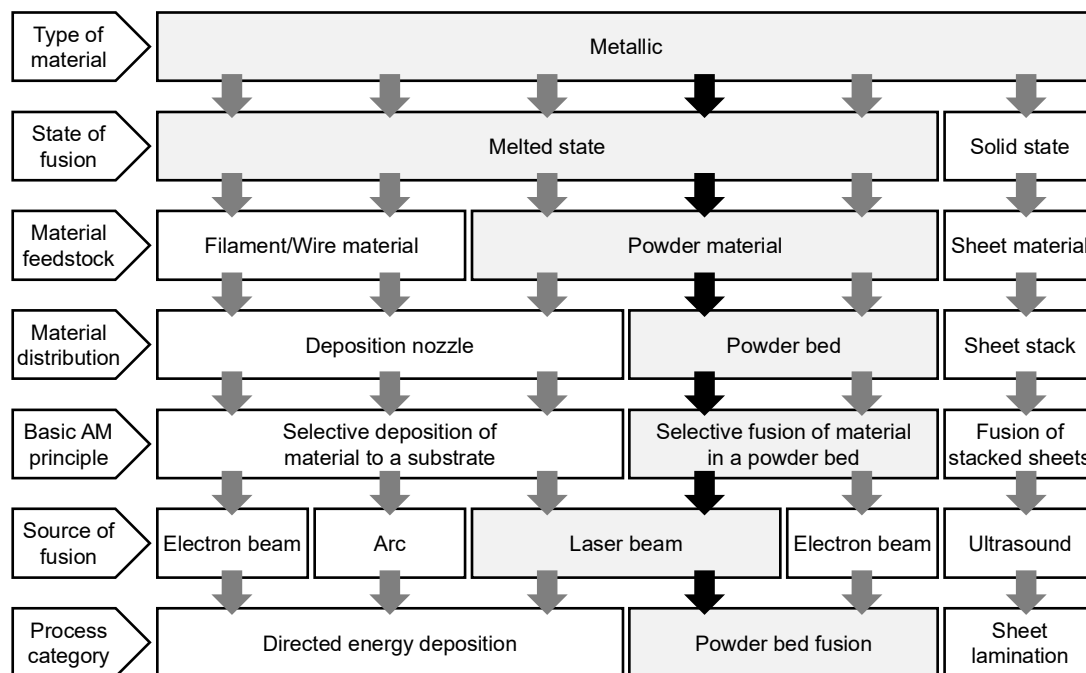


Figure 1: Classification of the principles of single-step additive manufacturing for metallic materials, adapted from DIN EN ISO/ASTM 52900:2022-03.

The entire process flow can be divided into pre-process, in-process, and post-process [8]. In the pre-process, the system technology and the feedstock are prepared. This includes programming the system according to the parts to be manufactured and assigning the process parameters. In-process covers the actual manufacturing. This in-process can be divided into 2–3 sub-steps. The corresponding sequence is shown in Figure 2 using an exemplary machine layout.

At the core of the PBF-LB/M machines is the build space, which is typically designed as a cuboid or cylinder. In this build space, a powder layer is first applied to a build platform or, in subsequent layers, to an existing powder bed (see Figure 2a). To achieve this, powder is supplied by a dosing system, such as powder supply platforms or systems that dispense powder from the top. The powder layer is typically coated on the build plane using rubber lips, blades, or brushes [9]. Excess powder is collected in an overflow. After the recoating, the powder layer is exposed by a laser (see Figure 2b). The material absorbs certain fractions of the incident radiation, causing the temperature to rise and the powder material to melt and fuse in desired areas. The exposure is vector-based, with a hatching of the areas to be fused and one or more contour exposures. Typically, optical setups are used that consist of a solid-state beam source (commonly single-mode fiber lasers) and a scanning system [6]. The scanning system deflects the laser beam using galvanometer scanners and focuses it in the build plane using dynamic focusing systems or flat field lenses. After the exposure step, the build platform is lowered by the desired layer thickness, and the next layer of powder is applied (see Figure 2c).

Once the components have been completely built layer by layer, post-processing follows. During post-processing, the components and the build platform are cleaned of unfused powder and removed from the system. This is followed by material-specific heat treatments where applicable. As the components are fused to the build platform, they are separated, and support structures are removed. Further post-process steps are optionally possible.

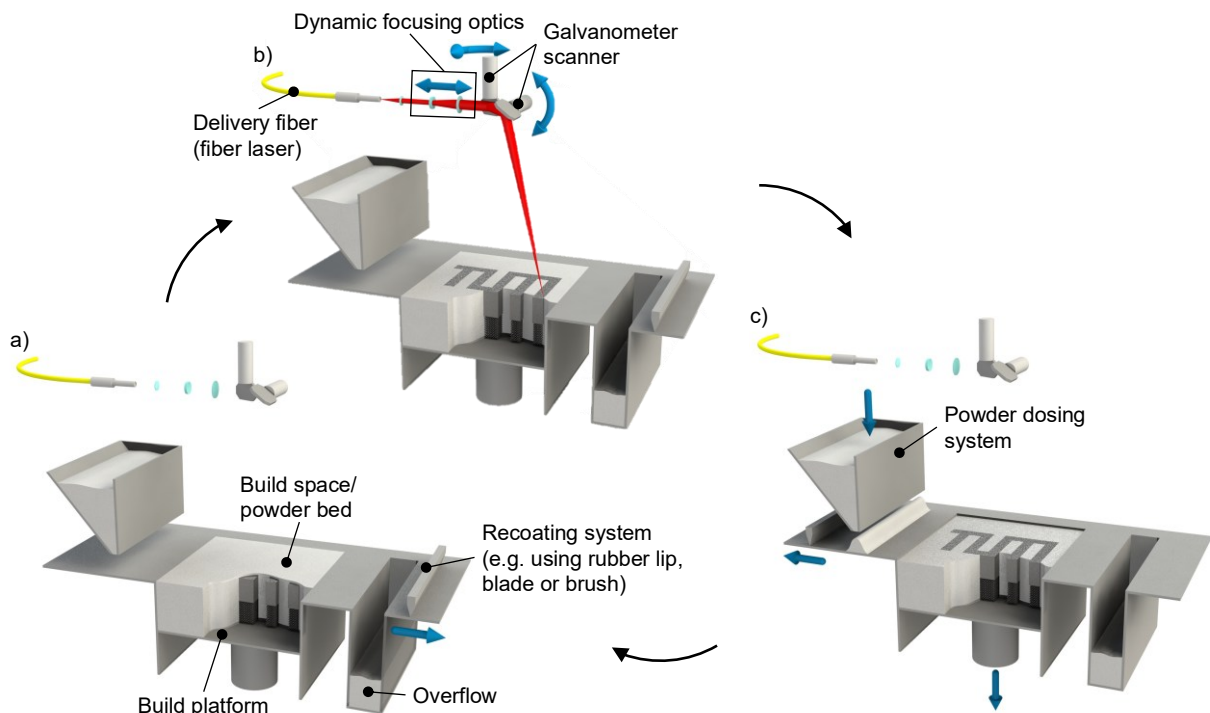


Figure 2: In-process procedure of PBF-LB/M a) recoating powder layer, b) exposing build area, c) lowering build platform.

2.1.1 Characteristic process properties and process limits

With a focus on beam shaping, this thesis examines the exposure step and its corresponding process behavior in detail. Exposing the powder bed yields a range of physical effects (see Figure 3).

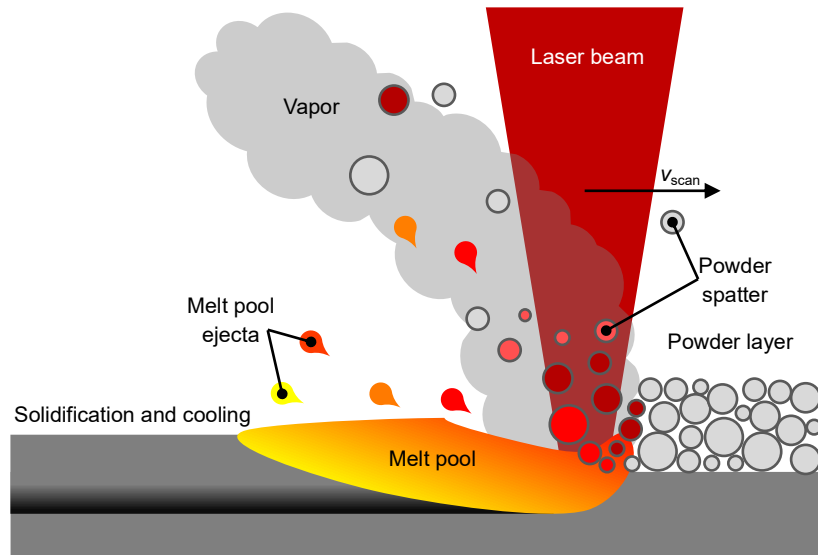


Figure 3: Schematic overview of the process zone during the exposure step.

Melt pool formation and melting modes

At the beginning of the exposure, the laser radiation is absorbed by the newly applied powder layer. Due to multiple reflections at the surfaces of the powder particles, the coupled power is higher than it would be in the case of Fresnel absorption [10]. Due to heat conduction, the absorbed energy is distributed in a microscopic area. As soon as the melting temperature is reached, a melt pool forms. This changes the coupling behavior. The laser radiation is coupled to the surface of the melt pool, resulting in different outcomes depending on the shape and absorption properties of the material used [11, 12]. Additionally, the presence of the liquid alters the heat flow in the process zone [13]. Temperature differences within the melt pool cause Marangoni flows. Marangoni flows describe the movement of fluids from areas of lower surface tension to areas of higher surface tension [14]. The surface tension depends on the temperature [15]. Consequently, there is a more pronounced Marangoni flow within the melt pool at higher temperature gradients than at lower temperature gradients. The direction of the flow relative to the temperature gradient depends on the materials used. Many metals and alloys processed using PBF-LB/M reduce their surface tension as the temperature rises [16]. This means that the melt flows from hotter to colder areas. When these flows within the melt pool, particularly in the boundary area adjacent to the solid material, are highly distinctive, convection effects extend the heat flow [17] (see Figure 4). The convection effects can cause the shape of the melt pool to change.

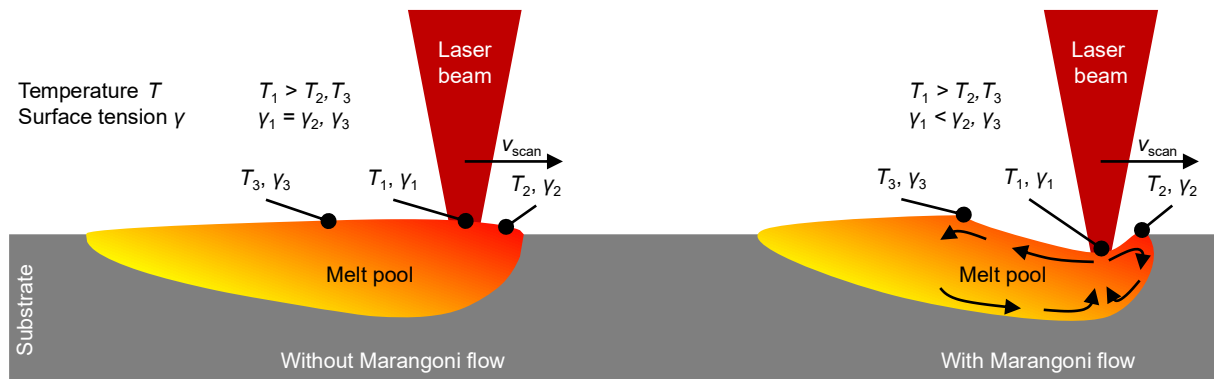


Figure 4: Schematic illustration of the influence of Marangoni flows on the melt pool shape.

Besides melt formation, the high intensity of the laser radiation in the PBF-LB/M (average intensity in the range of 10^6 to 10^7 W/cm²) results in partial evaporation of the material. The metal vapor streaming out of the process zone adds a recoil pressure to the already described physical effects on the melt pool, which additionally changes the flow in the process zone [13]. As a result of the recoil pressure, the melt is pushed down into the base material. Combined with the Marangoni flows, this alters the shape of the melt pool cross-section and the shape of the melt pool surface. At moderate pressures and flows, the melt pool forms a bowl-shaped structure. If the vapor pressure increases, this bowl in the melt pool surface deepens until a deep, narrow vapor capillary is formed [18]. This so-called keyhole is characteristic of laser deep penetration welding [19]. To summarize, as with all laser-based melting processes, two main melting modes can occur with PBF-LB/M, with a transition mode in between:

- **Conduction mode:** In conduction mode, heat transfer is driven by heat conduction. Convection plays a subordinate role. The resulting melt track cross-sections have a width-to-depth aspect ratio of less than 0.5 [20]. In this melting mode, heat conduction welding processes are typically performed.
- **Transition mode:** In transition mode, melt flows and the recoil pressure begin to deform the melt pool shape. The resulting width-to-depth aspect ratios are in the range between 0.5 and 0.8 [21]. In PBF-LB/M, this melting mode is often employed.
- **Keyhole mode:** In keyhole mode, the recoil pressure leads to the creation of a deep, narrow vapor capillary referred to as the keyhole [22]. This results in melt tracks with depth-to-width aspect ratios larger than 0.8 [23]. In PBF-LB/M, keyhole mode is typically avoided because keyhole collapsing creates the risk of entrapped gas bubbles, potentially resulting in porosity [18].

The melting modes are characteristic of the process behavior. However, this does not necessarily define the process limits.

Process window

As briefly described in the melting mode overview, PBF-LB/M is typically performed in transition mode. However, the process limits for manufacturing three-dimensional components are typically defined by defects that occur as a result of the energy input.

The lower limit of the PBF-LB/M process concerning introduced energy is determined by the fusion behavior. The applied energy must be high enough to completely melt the powder layer and fuse it to the existing solid (build platform or previously manufactured layer). The hatch distance between two neighboring melting tracks must be small enough to ensure a sufficient overlap of the individual tracks [24]. If less than the required energy is applied, if the powder layer is too thick, or if the hatched tracks are positioned too far apart, lack of fusion porosities occur in the components. These are characterized by irregularly shaped cavities (see Figure 5d).

When using higher laser powers in combination with high scanning speeds, balling, humping, and bulging of the melt pool can occur. There are various definitions of the terms balling, humping, and bulging, some of which are used synonymously. Balling and humping refer to single tracks, while bulging describes the effect on areas. Balling originally describes the effect of the molten powder layer not bonding completely with the substrate due to poor wetting behavior [25]. Instead, the powder particles partially fuse and stay on the surface as large balls of molten and solidified material [26]. Humping [27] and often synonymously balling [28] describe the phenomenon in which melt pools begin to break up longitudinally due to Plateau-Rayleigh capillary instability. The effect is often triggered when a specific length-to-width ratio of the melt pool is reached [29]. The corresponding melt tracks exhibit a rough surface with separated melt accumulations along the track [30, 31]. Both phenomena are driven by the surface tension of the melt and the associated desire to maintain the surface energy, thus lowering the ratio of surface area to volume. Similarly, bulging results in a regular or irregular elevation of the surface of an exposed area. This is typically caused by vast melt pools, which are created by the accumulated heat of several consecutive exposed scan tracks [32]. Due to the comparatively large spot sizes [33] and scanning speeds [34], bulging is a particularly known process defect in powder bed fusion of metals using an electron beam (PBF-EB/M). With state-of-the-art PBF-LB/M, bulging is a rather uncommon process defect, as typically small Gaussian spots with diameters in the range of 100 μm and average scan vector lengths in the range of 10 mm are used. Scanning speeds are in the range of 1 m/s when processing steels and nickel-based alloys, which typically do not result in melt pool sizes and shapes prone to bulging. However, larger spot sizes, which are examined in this doctoral thesis, result in larger melt pools, rendering bulging as a relevant

process defect [35]. The bulging of the top surfaces leads to increased roughness or, depending on the cut-off thresholds, waviness of the top surface. However, the samples initially still have a high relative density (see Figure 5b).

The upper process limit with respect to the energy input is determined by the keyhole porosity. As briefly mentioned in the description of the keyhole melting mode, a typical side effect of such keyholes is a regular collapse, which can trap gas bubbles [18]. If the melt solidifies while gas bubbles are present in the material, they remain as uniformly distributed, spherical pores in the finished part (see Figure 5a). Therefore, this unstable keyhole melting mode is avoided with PBF-LB/M.

In conclusion, with PBF-LB/M, processing is aimed at a condition between the described processing limits. Consequently, the parameter sets are designed to prevent keyholing and bulging top surfaces. At the same time, complete fusion of the powder particles is aimed for. The corresponding processes result in highly dense components with smooth top surfaces (see Figure 5c). In addition to these process window limits, which are defined by the resulting component properties in terms of relative density and top surface characteristics, the energy input can also influence the process dynamics, such as spatter formation.

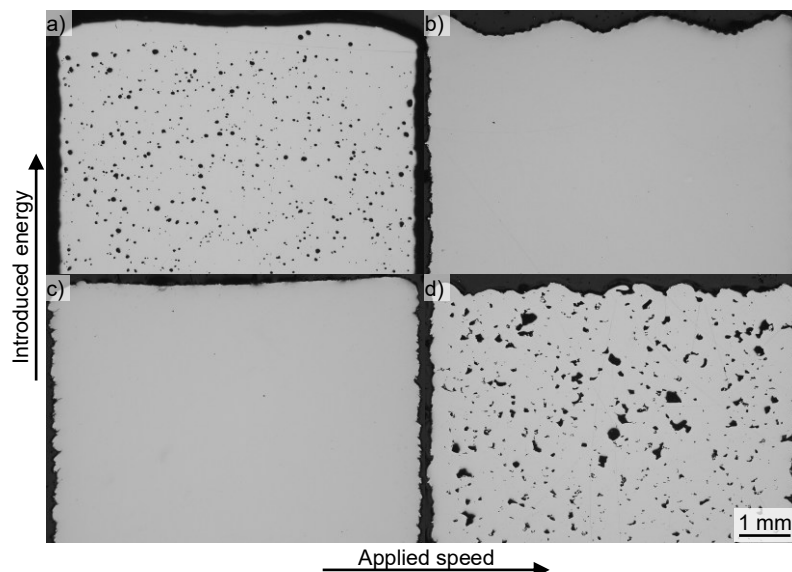


Figure 5: Exemplary micrographs of test specimens resulting in a) keyhole porosity, b) a bulging top surface, c) dense material with a smooth top surface, and d) lack of fusion porosity, micrographs published in [13].

Spatter formation

In addition to entering the keyhole mode, outflowing metal vapor causes powder denudation and spattering in PBF-LB/M. Powder denudation means the entrainment of powder particles by the surrounding gas flow in the direction of the process zone [36]. It is driven by the Bernoulli effect induced by the evaporation during the laser-material interaction. Neither powder denudation nor spatter formation is a strict process limit. However, both

phenomena may reduce the reliability of the PBF-LB/M process. Powder denudation can lead to a lack of fusion in the components due to the lack of feedstock material. Similarly, spatters larger than the initial powder particles can result in a lack of fusion or an altered microstructure due to a changed solidification behavior [37]. Moreover, the recoater may crash into large adhering spatter, leading to process abortions [38]. Additionally, large spatter particles are sieved out during powder preparation, thereby reducing the reusability of the material [39]. Furthermore, an increased oxygen content in reused spatter particles can change the resulting material properties of components, additionally reducing powder reusability [40].

Spatters during PBF-LB/M can be classified according to various categories. Firstly, a distinction can be made according to the origin of the spatter, whether it originates from the melt pool or the powder bed. Secondly, a distinction can be made according to the temperature of the spatter, meaning whether the spatter reached the melting temperature and became liquid or remained at ambient temperature [41]. The different spatter types are shown in Figure 6. Cold spatters originate from denudated powder in powder beds. Cold spatters are powder particles drawn into the process zone by the vacuum created by the outflowing metal vapor and heated shielding gas [36, 39]. This primarily redistributes the powder in the powder bed. When the entrained powder particles move through the laser beam and consequently heat up, the cold powder spatters become hot powder spatters. The heating may cause the powder particles to reach their melting temperature, and several particles may agglomerate or combine to form larger spatter particles [41]. In addition to the spatters originating in the powder bed, there are spatters generated by melt pool instabilities, which are known from laser welding processes. This spatter type is caused by the recoil pressure and Marangoni flows present in the melt pool [13, 39]. These hot melt pool ejecta are sometimes referred to as droplet spatters [39].

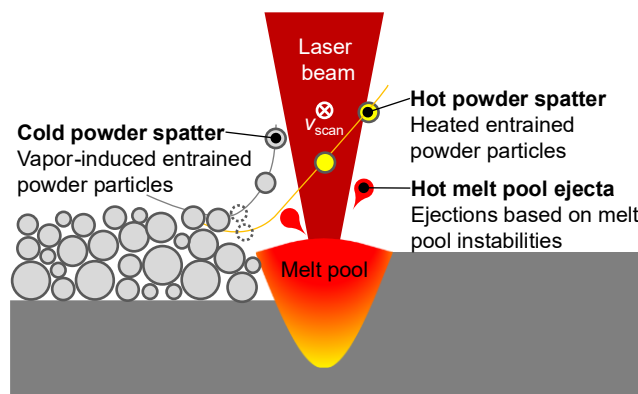


Figure 6: Schematic overview of the three different spatter types during PBF-LB/M: cold powder spatter, hot powder spatter, and hot melt pool ejecta, adapted from [41].

The composition and properties of the respective types of spatters depend on the process parameters used. Ly et al. [41] investigated the spatter distribution in the PBF-LB/M processes of AISI 316L and Ti6Al4V in the various spatter categories. The study used laser powers of 200 and 300 W, combined with scanning speeds of 1500 and 2000 mm/s. Approximately 60 % of the resulting spatters were hot powder spatters, around 25 % were cold powder spatters, and only around 15 % were melt pool ejecta.

Besides the typical process characteristics associated with metal evaporation and spatter formation, the operating principle of PBF-LB/M results in component properties that are characteristic of its microstructure. For this reason, microstructure formation in PBF-LB/M is described in more detail in the following section.

2.1.2 Characteristic microstructure

In general, the microstructure of metallic materials can be subdivided into different scales. Metallic materials typically solidify into a polycrystalline structure. In this case, the finest of these scales is at the crystal lattice level. The crystal structure may exhibit different lattice structures depending on the material and alloy composition. The austenitic stainless steel AISI 316L investigated in this thesis has a face-centered cubic (fcc) lattice. Depending on the growth rate and thermal gradient, the lattice structures form larger, varied shapes (see Figure 7).

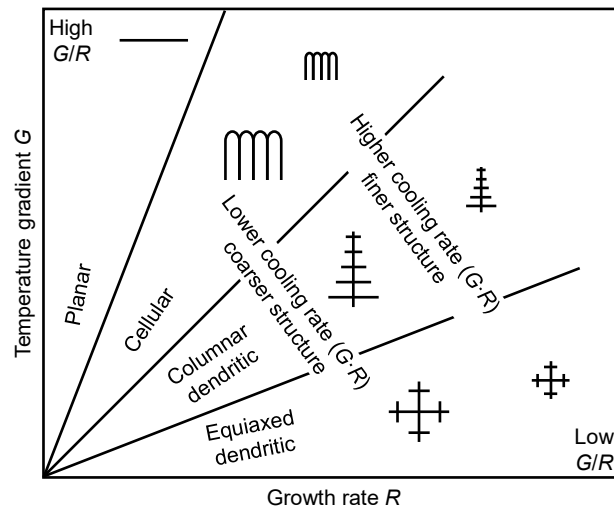


Figure 7: Influence of temperature gradient G and growth rate R on the microstructure formation.

In PBF-LB/M, the growth rate and thermal gradients are typically in a range to form columnar and equiaxed dendritic structures. If several of these dendritic structures grow towards and meet each other with a misalignment angle, grain boundaries are formed, which separate different grains from each other (see Figure 8). Consequently, the grains form the coarsest scale of the microstructure. The size of the individual grains primarily depends on

the cooling rate, represented by the product of the temperature gradient and growth rate, $G \cdot R$ (see Figure 7). Sections of the workpiece with a faster cooling rate form a finer microstructure. Component sections subject to a slower cooling rate form a coarser grain structure [42]. The grain shape and size [43] and the crystallographic texture [44] influence the macroscopic component properties. Therefore, it is essential to understand the formation of these microscopic substructures if the influence on the component properties is to be understood. Thus, the core mechanisms of the considered material, AISI 316L, are briefly revisited.

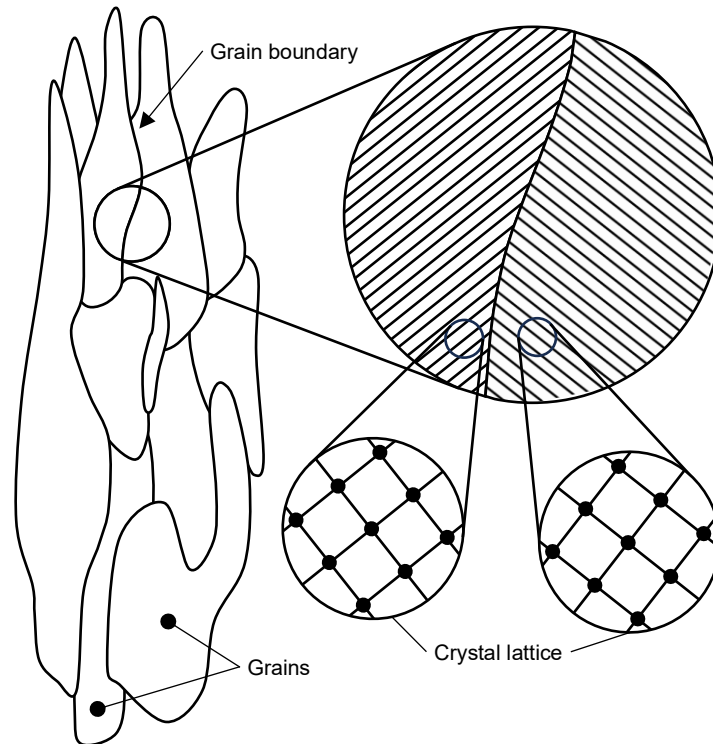


Figure 8: Schematic representation of the different scales of the microstructure, adapted from [15].

The microstructure of PBF-LB/M components is a result of the entire thermal history experienced by the components. The fine-tuning of a desired microstructure of PBF-LB/M components is a research area by itself (such as in [45]). Typically, the microstructure is adjusted to meet the desired resulting properties using heat treatments during post-processing. This can include techniques such as hot isostatic pressing (HIP), recrystallization annealing, hardening, stress relief annealing, ageing, and more. This thesis considers the microstructure exclusively created by the PBF-LB/M process (as-built condition).

The austenitic stainless steel AISI 316L does not undergo any further phase transformation during cooling after initial solidification. Therefore, the microstructure in the as-built condition mainly depends on the solidification microstructure of the remaining parts of the individual melt tracks that form the component [10, 46]. The microstructure of these

non-remelted parts primarily depends on the initial solidification and partial remelting during the build process [46]. The solidification microstructure is primarily influenced by the geometric shape of the melt pool and the solid and solidifying phases of the melt present at the melt pool boundary. In PBF-LB/M processing AISI 316L, it is assumed that austenite is the phase present in both cases [10]. This means that no nucleation is required for crystallization, but the grains at the melt pool boundary can continue to grow. This enables the epitaxial grain growth characteristic of PBF-LB/M.

Concerning the crystal structure, AISI 316L solidifies preferentially in the $\langle 100 \rangle$ directions of the crystal lattice [47]. In individual melt tracks, $\langle 100 \rangle$ oriented grains therefore grow perpendicular to the melt pool boundary along the thermal gradient towards the center of the melt pool. The fluctuating growth rate and the temperature gradient during PBF-LB/M result in an inhomogeneous microstructure, which typically ranges from a columnar microstructure at the outer edge of a single melt track to an equiaxed section in the center near the surface [48]. Since with PBF-LB/M partial melting of previously manufactured layers occurs, the equiaxed area is often remelted, resulting in a microstructure consisting of long columnar grains growing through multiple layers [49].

When transferring the microstructure development at the single-track level to three-dimensional components, the global temperature gradient within a component is an additional influence. With simple scanning strategies (e.g., bidirectional exposure with 90° rotation between the layers), columnar grains with a $\langle 100 \rangle$ texture in the build direction (BD) can be produced [50]. Bidirectional scanning of the hatch vectors creates a mean temperature gradient directed in the build direction. As a result, the grains preferentially grow with a $\langle 100 \rangle \parallel$ BD orientation (see Figure 9) [51]. This basic mechanism can be modified according to the scanning strategy employed [52, 53]. For example, if the remelting zone is set by selecting process parameters that favor grain growth at 45° in the $\langle 100 \rangle$ direction and the corresponding areas are not remelted, a $\langle 110 \rangle \parallel$ BD texture can be generated [54, 55]. Textures in the x-y plane can be broken by applying rotating scan strategies (e.g., a 67° rotation between layers), which can result in a fiber texture with $\langle 110 \rangle \parallel$ BD orientation. [56].

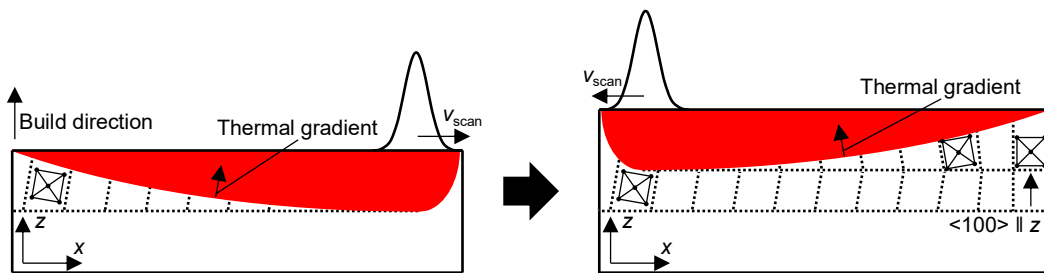


Figure 9. Schematic representation of the $\langle 100 \rangle \parallel$ BD texture build up in state-of-the-art PBF-LB/M, adapted from [51].

2.1.3 Conclusion from the state of the art in laser-based powder bed fusion of metals

All statements made in section 2.1 refer to the state-of-the-art PBF-LB/M using Gaussian beam profiles for exposure. However, the beam profile determines the local distribution and temporal progression of the energy input and thus the essential step of the PBF-LB/M process [57]. Furthermore, Gaussian beam profiles are frequently attributed to typical process limitations of PBF-LB/M such as local overheating, evaporation, balling, and spatter formation [58]. Therefore, the following sections deal with the topics of beam shaping and the application of the created beam profiles in laser-based melting processes.

2.2 Basics of beam shaping

Beam shaping is defined as the redistribution or modification of the intensity and phase of optical radiation [59]. In this thesis, the focus is on modifying the intensity distribution referred to as the beam profile, while the phase change is a secondary effect. As in PBF-LB/M, Gaussian beam profiles are generally the most common beam shape used in laser material processing [60]. One reason for this is their low divergence, which enables small focused spots [61]. However, beam shaping offers the opportunity to explicitly design beam profiles, allowing the laser material processing to be tailored to specific requirements [60].

Various optical working principles, such as reflection, refraction, diffraction, or fiber-based effects, can be utilized to implement beam shaping. Reflective optics could, for example, consist of deformable mirrors [62, 63]. As refractive optics, cylindrical lenses [64, 65], axicons [66, 67], or more complex optical assemblies [68, 69] can be used for beam shaping. Diffractive systems can be either rigid, e.g., diffractive optical elements (DOEs) [70, 71] or multi-plane light conversion [72, 73], or flexible, such as spatial light modulators (SLMs) [74, 75], acousto-optical deflectors (AODs) [76, 77], or coherent beam combining [78]. Another method for performing beam shaping is to utilize multi-core fibers [79, 80]. Here, designs can be used in which the laser radiation can be distributed to different rotational-symmetric guide regions of a fiber [81]. This enables beam profiles consisting of a Gaussian core and a surrounding ring (see Figure 10).

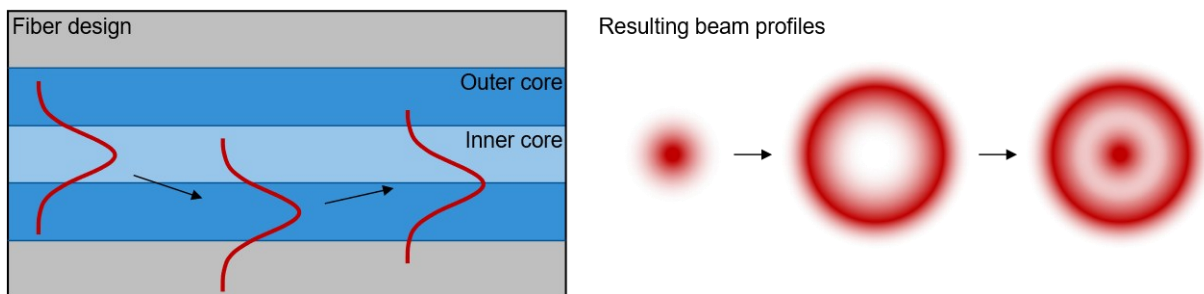


Figure 10: Schematic illustration of the fiber design of a multi-core fiber and the resulting beam profiles.

The various methods have different drawbacks, limiting their industrial applicability. For example, deformable mirrors are more likely to distort the beam profile than to actually shape it. Switching the beam shape in rigid systems such as lenses, DOEs, or multi-plane light conversion is only possible by mechanically adding or removing optics. Systems like SLMs exhibit strong thermal lensing effects and can therefore only operate stably at low power levels [82]. Additionally, flexible diffractive systems, such as AODs and coherent beam combining, generate side orders in the diffraction patterns, which can have undesirable effects [77]. Furthermore, beam shaping in these approaches is based on free-space optics, entailing increased sensitivity to contamination and misalignment [81]. When using multi-core fibers, the design freedom is determined by the fiber design. However, the beam shaping optics can be flexibly adjusted within the given design frame, do not require any additional optics, and accordingly do not cause any adverse optical effects. For this reason, beam sources including multi-core feeding fibers are increasingly being used in research and industry [83].

2.3 Beam shaping in laser-based melting processes

Preliminary remark: The research area concerning the application of beam shaping techniques in laser-based melting processes has grown considerably during the period of research of this doctoral thesis (2020–2025). Therefore, some of the literature presented was published in similar time periods to the publications included, resulting in a partial overlap of the findings with those of this doctoral thesis.

As outlined in Section 2.1, Gaussian beam profiles are increasingly being attributed to process limitations of state-of-the-art PBF-LB/M [58]. The beam profile can be adapted to the process requirements by using beam shaping [60]. However, a comprehensive investigation of the influence of different beam profiles in PBF-LB/M is challenging. The PBF-LB/M process has many degrees of freedom due to the succession of melt tracks in a plane and the stacking of these planes to form a volume. Therefore, for a systematic investigation, preliminary studies are often conducted on PBF-LB/M single melt tracks or laser welding, collectively referred to as laser-based melting processes. Accordingly, the following sections are structured to present the findings from single-track experiments first, followed by an examination of the PBF-LB/M process.

The use of beam shaping in laser-based melting processes is motivated by various objectives. Depending on the motivation and objective, the publications examine different influences. The core statements on the influence of beam shaping in laser-based melting processes can be classified into the following four basic categories:

- Accelerating processes
- Increasing process reliability through larger process windows or reduced process dynamics
- Affecting the microstructure
- Enabling the processability of new and hard-to-process materials

2.3.1 Process accelerations and reliability increase

Mi et al. [62] showed the potential of a deformable mirror to specifically influence the melt pool length, width, and depth by distorting the spot during the processing of Ti6Al4V. This potentially enables control of the melt pool size and shape. In order to produce deep and wide cross-sections in deep penetration welding, Hansen et al. [84] generated various multi-spot distributions using a beam-splitting approach with a diffractive optical element to process the two steels EN DC01 and AISI 304. This approach enabled the adjustment of the melt pool width according to the number of laser spots used, while maintaining the penetration depth. Slodczyk et al. [85, 86] transferred this approach of widening the melt pool to the PBF-LB/M process of AISI 316L. Single-track experiments revealed that the beam-splitting approach could increase the melting rate [85].

Rasch et al. [70] were able to show experimentally that ring-shaped beam profiles yield wider, flatter melt pools compared to Gaussian beam profiles during heat conduction welding of an aluminum-copper alloy. This increased the process reliability and reduced surface roughness. Similarly, Nahr et al. [87] demonstrated that the process window is enlarged and the occurrence of humping is reduced when processing the high-strength aluminum alloy EN AW-5083 using ring-shaped beam profiles.

In contrast to the beam shaping in the focal plane, Volpp and Vollertsen [88] employed a beam-shaping approach with several focal planes offset in the direction of the beam axis to process EN AW-6082. With this multi-focus beam shaping, it was possible to avoid keyhole collapses, resulting in a reduced spatter formation. With the same aim of reducing process dynamics, the use of centric top-hat or Gaussian profiles in combination with surrounding ring-shaped beam profiles is being increasingly investigated in laser deep penetration welding of various steels and aluminum alloys. By using these point-ring-shaped beam profiles, the keyhole size, shape, and dynamics can be adjusted, resulting in a stabilized keyhole and reduced spatter formation (see Figure 11) [89, 90]. The underlying mechanism is a widened keyhole opening [91], resulting in a stabilized welding process with smooth weld bead surfaces [92, 93].

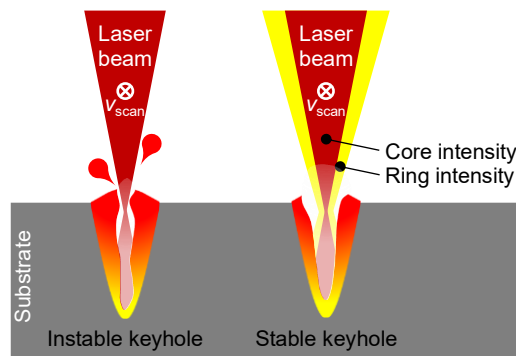


Figure 11: Schematic illustration of keyhole widening and stabilization by using ring-shaped beam profiles, adapted from [94].

These single-track studies demonstrate the fundamental potential of beam shaping to specifically influence the dynamic behavior of the process and the geometry of the melt pool. This allows, for example, the width of the melt tracks to be increased, which can be used to achieve higher process speeds. Additionally, existing keyholes have been stabilized, which can improve the reliability in PBF-LB/M through larger process windows and reduced spatter formation. The following section demonstrates the potential of beam shaping at PBF-LB/M.

Historically, the investigation of the potential of non-Gaussian beam profiles in PBF-LB/M started with using multi-mode beam sources with top hat beam profiles [95]. As early as 2011, Buchbinder et al. [96] demonstrated the potential of beam sources with top hat profiles and powers of up to 1 kW to increase build rates by more than a factor of 4 compared to the state of the art at the time, processing AlSi10Mg. Schleifenbaum et al. [97] combined those beam sources with the use of layer thicknesses of up to 250 μm when processing the steels AISI 316L and H11. The build rates could be increased by a factor of 15, whereby a reduced component density accompanied the increased build rate [97]. Loh et al. [98] used a similar optical setup to show the potential for reducing process time when processing aluminum alloy EN AW-6061. Using a larger top-hat beam profile with faster scanning speeds resulted in wider melt pools compared to a state-of-the-art Gaussian beam profile, allowing for a larger hatch distance. The change in beam profile in these early cases was linked to the beam quality of the multi-mode beam sources used.

Tumkur et al. [66] used Bessel beams during PBF-LB/M of AISI 316L. Compared to Gaussian beams, they found greater process reliability due to a reduced tendency to form keyhole pores and reduced process dynamics, resulting in less wavy track surfaces and reduced spatter formation. Since Bessel beams are not diffraction-limited, the melt pool dimensions were less dependent on focal plane shifts, additionally increasing process reliability. The altered intensity distribution led to a reduced thermal gradient, which stabilized the keyhole and diminished Marangoni flows, thereby reducing process dynamics.

One of the first research groups investigating the influence of ring-shaped beam profiles on the PBF-LB/M process was at Moscow State University of Technology. Gusarov et al. [99] stated that the establishment of the PBF-LB/M in the industrial environment is hindered by the low productivity. Motivated by this, Gusarov et al. [99], Metel et al. [100], Zhirnov et al. [69], and Okunkova et al. [68, 101] investigated the influence of top-hat and ring-shaped beam profiles on various targets. All studies used CoCrMo powder as feedstock material. Zhirnov et al. [69] presented larger process windows for top-hat and ring-shaped beam profiles, compared to a Gaussian beam profile. Gusarov et al. [99] and Metel et al. [100] showed that top-hat and ring-shaped beam profiles can accelerate the process. Using a top-hat beam profile, a power and speed increase of 14 % was achieved. With a ring-shaped beam profile, a 43 % increase in power and speed was possible. Okunkova et al. [68, 101] demonstrated experimentally that using top-hat and ring-shaped beam profiles can qualitatively reduce spatter formation and powder denudation. For these investigations, a monitoring setup was employed, consisting of an off-axis high-speed camera that imaged the process in the visible wavelength range and an on-axis near-infrared camera that recorded the thermal radiation of the process zone [102]. An explanation for the reduced spatter tendency was provided by a numerical simulation, which demonstrated that ring-shaped and top-hat beam profiles reduce evaporation losses [103].

Similarly, reduced spattering was observed when processing Ti6Al4V using ring-shaped beam profiles instead of Gaussian beam profiles [104]. Yuan et al. [105] used high-speed X-ray imaging to investigate the process dynamics in the PBF-LB/M of Ti6Al4V with ring-shaped beam profiles. They found an overall reduction in the dynamics of the process zone. There was less spattering, less fluctuation in the melt pool surface, and the keyhole was wider, with no unstable cavities due to the changed intensity distribution. In addition, Ślodziński et al. [106] demonstrated a reduction in the total amount of spatter when using a beam-splitting approach to process AISI 316L. In contrast, Kippels et al. [107] reported a higher spatter amount with smaller spatter particles when using ring-shaped beam profiles instead of Gaussian beam profiles for processing IN625. However, the paper does not discuss possible causes for this behavior not being reported before.

Wischeropp et al. demonstrated the potential of ring-shaped beam profiles in the PBF-LB/M of AlSi10Mg [71, 108] and AlSi16Sc [108]. For these aluminum alloys, the process increased reliability due to larger process windows through the use of ring-shaped beam profiles. A process acceleration could not be demonstrated experimentally due to a lack of data points. Vanmunster et al. [109] investigated the combination of ring-shaped beam profiles with increased layer thicknesses in PBF-LB/M of AISI 316L. Especially with thin

powder layers below 60 μm , ring-shaped beam profiles have potential for process acceleration. For powder layers of 60 μm and above, the build rates achieved were comparable for the Gaussian and ring-shaped beam profiles used. The identified process accelerations can be utilized to achieve both time and energy savings. Thus, Cozzolino et al. [110, 111] examined the influence of using ring-shaped beam profiles on energy consumption in PBF-LB/M. When ring-shaped beam profiles were used, the build rate increased by up to 46 %, while the energy consumption of the laser was reduced by 16 % [111]. Although the laser used required higher power, energy consumption was reduced due to the shorter exposure time [110].

The referenced studies demonstrate the significant potential of beam shaping to enhance process reliability in PBF-LB/M by reducing spatter formation and expanding process windows. Moreover, process speed can be increased through higher melting and build rates. While the studies primarily evaluated mesoscopic component properties such as component density and surface roughness, the change in the beam profile also influences the solidification behavior, which can lead to a modified microstructure.

2.3.2 Effects on the microstructure

Ebrahimi et al. [57] showed by means of a simulation that using beam shaping during heat conduction welding of AISI 316L influences the melt-pool behavior, and thereby the formation of the solidified melt track cross-section and the resulting grain morphology. In addition, the grain size in the solidified melt track can be influenced via the choice of beam profile by changing the temperature distribution and the cooling rate [112]. This can be attributed to the solidification conditions changed by beam shaping during microstructure formation, such as the solidification rate and temperature gradient [113].

Esmailzadeh et al. [74] demonstrated how the thermal gradient in the PBF-LB/M of Ti6Al4V can be controlled by tailored beam profiles, creating a lamellar $\alpha+\beta$ structure instead of the typical martensitic structure. Mohebbi and Ploshikhin [114] showed that an almost entirely equiaxed fine-grained microstructure can be produced by using ring-shaped beam profiles when processing a Sc-Zr-modified Al alloy (Scalmalloy®). One hypothesis to explain the results is that a reduced cooling rate, resulting from the ring-shaped beam profile, leads to the primary precipitation of the L12-Al₃(ScxZr1-x) phase, thereby enhancing grain refinement. Galbusera et al. [115] examined the influence of ring-shaped beam profiles on the resulting microstructure when processing AlSi7Mg0.6. A dependence of the melt track depth on the beam profile used was found. The melt track depth, in turn, correlated with mechanical parameters such as the ultimate tensile strength and elongation at break. Consequently, mechanical parameters could be influenced by the choice of beam profile.

In addition, Galbusera et al. [116] showed that ring-shaped beam profiles change the grain shape aspect ratios and result in different crystallographic textures compared to Gaussian intensity distributions when processing a low-Si alloyed steel. Moreover, when processing austenitic materials (such as AISI 316L or IN718), beam shaping approaches have the potential to change the grain sizes and shapes [65, 117]. A research group at the Lawrence Livermore National Laboratory showed simulatively [118] and experimentally [64] that the nucleation behavior of AISI 316L and thus the grain size can be influenced by using elliptical beam profiles instead of circular Gaussian beam profiles. The use of elliptical beam profiles with a scan direction along the minor axis resulted in smaller and equiaxed grains, which was attributed to lower thermal gradients. Additionally, the crystallographic texture of the microstructure depends on the beam profile. Wang and Shi [119] and Montero-Sistiaga et al. [120] found that large top-hat beam profiles combined with high laser powers result in highly textured microstructures of AISI 316L and IN718 with a $\langle 100 \rangle \parallel \text{BD}$ orientation and a large grain size in the millimeter range. Similarly, Cozzolino et al. [110] report a $\langle 100 \rangle \parallel \text{BD}$ texture when using a ring-shaped beam profile processing IN718. In contrast, Pérez-Ruiz et al. [121] showed that by using ring-shaped beam profiles, a $\langle 111 \rangle \parallel \text{BD}$ texture in IN718 can be generated. To achieve this texture, a ring-shaped beam profile was used in combination with a suitable hatch distance, which remelts a large portion of the generated melt track cross-section. Thereby, no rotation of the scan vectors between the layers was performed. Mirzabeigi et al. [122] demonstrated that different types of microstructures can be generated by selecting a specific beam profile for processing IN718. When using a line-shaped and a chair-shaped beam profile, a fine-grained and weakly textured microstructure is formed. In contrast, a coarse-grained and highly $\langle 100 \rangle \parallel \text{BD}$ textured microstructure was produced when using a ring-shaped beam profile.

In summary, beam shaping can enable microstructure adaptation of materials established in PBF-LB/M. Additionally, the adapted solidification behavior can be utilized to process new or challenging materials for PBF-LB/M.

2.3.3 Processability of new and hard-to-process materials

As indicated in the previous section, beam shaping affects the temperature distribution and progression in the process zone during the solidification [57]. This offers the potential to adapt the process, enabling the processing of new and challenging materials for PBF-LB/M.

Thus, Stoll et al. [123] demonstrate that using ring-shaped beam profiles can reduce the tendency towards longitudinal cracking when processing tungsten by adapting the solidification and cooling behavior. Baldi et al. [79] investigated the processability of pure zinc to produce thin structures for use as anodes in battery applications. Similar to more

established materials, such as steels or aluminum alloys, the use of ring-shaped beam profiles reduced the tendency to enter the keyholing process regime, resulting in a stable process. Cloots et al. [124, 125] investigated experimentally the processability of the nickel-based alloy IN738LC, which is prone to hot cracking. The use of ring-shaped beam profiles resulted in reduced crack density at low scanning speeds compared to the use of Gaussian beam profiles. However, crack-free processing was not achieved. With a similar aim, Zenz et al. [72] simulated the processing of the nickel-based alloys CM247LC and IN713LC. Based on the modeled processes, a rotationally symmetric beam profile was designed, resulting in a reduced solidification crack density in test specimens.

2.3.4 Conclusion of beam shaping in laser-based melting processes

The referenced studies demonstrate the potential of beam shaping to affect PBF-LB/M. This can be used to achieve higher process speeds and stabilized keyholes, larger process windows, reduced spatter, and thus greater process reliability. Furthermore, the temperature distribution and its progression during processing can be adjusted. On the one hand, this allows the microstructure to be modified. On the other hand, it can be used to process challenging materials.

However, the majority of studies examine the influence of beam shaping on a black box basis using input-output correlations. As a result, processes examined with different beam profiles are not always comparable. Furthermore, the effects of the beam profile cannot be isolated from other process influences. This explains why some contradictory findings are reported in the state of the art. For instance, there are reports of reduced spatter formation [104, 105], as well as a higher number of spatters [107]. Moreover, in some cases, larger beam profiles are reported to lead to an increased number of nucleations and thus to smaller grains [64, 65, 114], while in other cases, the use of larger beam profiles results in larger grains [118, 120]. To compare laser material processing with different spot sizes, the literature on scaling laws and dimensionless parameters is explored in the following section.

2.4 Dimensionless parameters for laser material processing

The process behavior and the component properties of the PBF-LB/M process depend on numerous influencing variables and parameters. The black box approach, frequently used in the state of the art to determine stable processing windows and draw conclusions about the resulting component properties, requires extensive parameter studies when new materials are processed or new system technologies are employed. Particularly when something as fundamental to the process as the beam profile is altered. This also contributes to the limited

transferability of processes between systems. For parameter studies regarding the PBF-LB/M process, many influencing variables (such as material, laser spot size, and layer thickness) are typically kept constant and a limited number of parameters (such as laser power, scan speed, and hatch distance) are varied. Often, the standard PBF-LB/M process parameters, laser power P_{laser} , scanning speed v_{scan} , hatch distance h , and layer thickness δ_{layer} , are combined to form the volume energy density E_V according to:

$$E_V = \frac{P_{\text{laser}}}{v_{\text{scan}} \cdot h \cdot \delta_{\text{layer}}} \quad 2-1$$

The target value (e.g., the relative component density) is observed as a function of this combined parameter, which does not include any dynamic or temporal aspects. To counteract this, the generated data can be used to create process maps, representing the process results (e.g., the relative component density), with the scanning speed as an abscissa and the laser power as an ordinate. As this type of representation only has two dimensions, additional process maps are required for each parameter variation. The high number of degrees of freedom in the PBF-LB/M results in an unmanageable number of process maps, which limits the comparison of the contents. Neither of these common approaches includes any information about the laser beam profile used (neither shape nor size), which has an enormous influence on the process result [20]. Therefore, Bakhtari et al. [58] stated that the use of dimensioned parameters, such as volume energy density, is not suitable for comparing processes using beam shaping.

To reduce the experimental effort and enable comparability of laser material processing, dimensionless process parameters were introduced in the 1990s. These parameters include process- and material-related parameters, such as the laser spot size and thermophysical material properties, as well as process parameters, including the laser power and feed rate. Since the resulting parameters enable the scaling of process parameters, the corresponding relationships are sometimes referred to as scaling laws.

One of the first approaches to dimensionless process maps was published by Ion et al. [126] in 1992. The axis dimensions of the corresponding process maps are formed by the dimensionless laser power P_{laser}^* and the dimensionless traverse speed equal to the Péclet number Pe . The parameters are defined as analogous to

$$P_{\text{laser}}^* = \frac{A \cdot P_{\text{laser}}}{2w_0 \cdot k \cdot (T_m - T_0)} \quad 2-2$$

$$Pe = \frac{v_{\text{scan}} \cdot 2w_0}{\alpha} \quad 2-3$$

with the absorption A , the spot radius w_0 , the thermal conductivity k , the delta between melt and initial temperature $T_m - T_0$, and the thermal diffusivity α .

With the resulting dimensionless process maps, Ion et al. [126] were able to coarsely cluster three basic regimes of laser material processing. Similar to the melting modes defined in section 2.1, a distinction was made between processes that only heat the material but do not convert it into the molten phase, processes that melt the material, and processes in which the material is (partially) evaporated. The qualitative classification of the processes is consistent with the estimated resulting peak temperature on the substrate surface during processing. To estimate the peak temperature, Ion et al. [126] used the approach of Rykalin et al. [127] based on a heat conduction model of a moving Gaussian heat source on a semi-infinite substrate, which was adapted by Ashby and Easterling [128].

Patel and Vlasea [21] extended the existing approach to include the powder layer present in PBF-LB/M. For this purpose, the dimensionless powder layer thickness δ_{layer}^* was used, as described by Thomas et al. [129] as:

$$\delta_{\text{layer}}^* = \frac{2 \cdot \delta_{\text{layer}}}{w_0} \quad 2-4$$

The dimensionless power was transformed into the dimensionless energy input E^* as:

$$E^* = \frac{P_{\text{laser}}^*}{\delta_{\text{layer}}^*} \quad 2-5$$

For the model-based estimation of the melting mode, the approach for calculating the peak temperature by Ashby and Easterling [128] was used analogously to [126]. Patel and Vlasea [21] used the dimensionless depth z^* (depth/spot radius) at which a material-specific empirically assigned dimensionless peak temperature T_{peak}^* is reached to assign the melting mode. The limit values of the dimensionless depths z^* are analogous to the crossover values of the melt track depth-to-width aspect ratio of the melting modes (0.5 and 0.8), as defined in section 2.1. Figure 12 shows an exemplary dimensionless process map. A significant challenge in Patel and Vlasea's approach [21] is that the absorption coefficient A is assumed based on the spot size and the aspect ratio of the melt track depth to width. This procedure is certainly physically justifiable, but it requires a cross-section of a melt track to be available for estimating the melting mode. On the one hand, this causes increased experimental effort. On the other hand, the melting mode is defined by the aspect ratio. This means that, given the cross-section, an estimation of the melting mode is no longer necessary, which reduces the added value of the approach in this respect. However, the approach shows that the boundaries of different melting modes in PBF-LB/M can be determined by applying a heat conduction model.

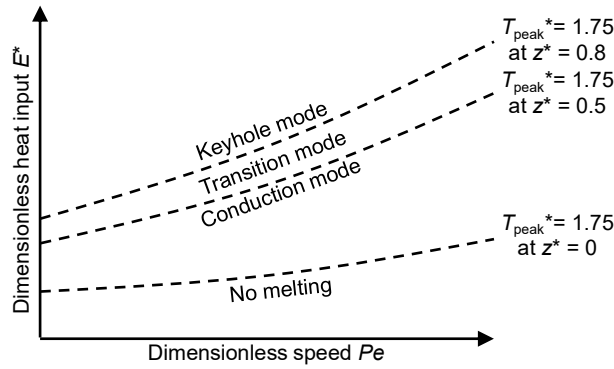


Figure 12: Schematic dimensionless process map with predicted melting modes based on reached dimensionless peak temperatures T_{peak}^* at defined dimensionless depths z^* for AISI 316L according to Patel and Vlasea [21].

Another approach based on the work of Ion et al. [126] was the introduction of dimensionless enthalpy by Hann et al. [130]. The dimensionless enthalpy relates the energy introduced into a specific volume of material, respectively, its mass, to the enthalpy at the melting temperature. There are various approaches in the literature for calculating the enthalpy introduced, all of which are proportional to each other. In the following, the calculation of the dimensionless enthalpy is derived for comprehensibility:

The enthalpy introduced results from the energy ΔE that is absorbed during the passage of the laser spot and the volume V , respectively, the mass connected to the volume via the material density ρ , into which this energy is diffused, as:

$$\Delta H = \frac{\Delta E}{\rho \cdot V} \quad 2-6$$

The introduced energy is calculated from the absorbed laser power during the interaction time τ as:

$$\Delta E = A \cdot P_{\text{laser}} \cdot \tau \text{ with } \tau = \frac{2w_0}{v_{\text{scan}}} \quad 2-7$$

The volume in which the heat can spread is assumed to be a horizontal elliptical half cylinder (see Figure 13) with one half-axis represented by the beam radius w_0 and the other half-axis represented by the thermal diffusion depth δ_{diff} . The corresponding volume is calculated as:

$$V = v_{\text{scan}} \cdot \tau \cdot \frac{\pi}{2} \cdot w_0 \cdot \delta_{\text{diff}} \text{ with } \delta_{\text{diff}} = \sqrt{\alpha \cdot \tau} \quad 2-8$$

This gives the introduced enthalpy ΔH as:

$$\Delta H = \frac{A \cdot P_{\text{laser}} \cdot \tau}{\rho \cdot v_{\text{scan}} \cdot \tau \cdot \frac{\pi}{2} \cdot w_0 \cdot \sqrt{\alpha \cdot \frac{2w_0}{v_{\text{scan}}}}} = \frac{A \cdot P_{\text{laser}}}{\rho \cdot \frac{\pi}{4} \cdot \sqrt{v_{\text{scan}} \cdot \alpha} \cdot (2w_0)^3} \quad 2-9$$

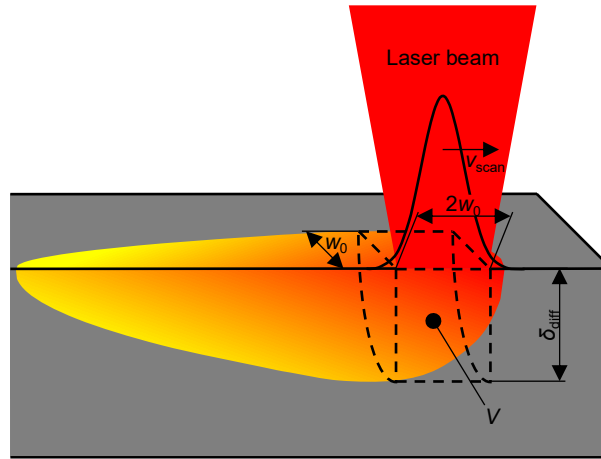


Figure 13: Sketch of heat conduction during laser-material interaction, including the relevant geometric dimensions for calculating the dimensionless enthalpy.

This variant differs from the variant of Hann et al. [130] by a factor of $2/\sqrt{2\pi}$, which is equal to a factor of approximately 0.8. This means it is proportional. The resulting deviation is in a range of differences of estimated absorption values in the literature. However, to be consistent with existing literature, the introduced enthalpy ΔH in this doctoral thesis is calculated according to:

$$\Delta H = \frac{A \cdot P_{\text{laser}}}{\rho \cdot \sqrt{\pi} \cdot v_{\text{scan}} \cdot \alpha \cdot (2w_0)^3} \quad 2-10$$

The enthalpy at melting h_s has several definitions in the literature. Some approaches define the enthalpy at melting h_s as the enthalpy present at the melting temperature (e.g., King et al. [20]). Others define the enthalpy at melting h_s as the enthalpy required to heat the material from an ambient temperature T_0 to the melting temperature T_m (e.g., Weaver et al. [131]). In this thesis, an extended definition including the latent heat of fusion L_f (e.g., Coen et al. [132]) is used, resulting in

$$h_s = c_p \cdot (T_m - T_0) + L_f \quad 2-11$$

with the specific heat capacity c_p . Thus, the dimensionless enthalpy results as:

$$\frac{\Delta H}{h_s} = \frac{A \cdot P_{\text{laser}}}{(c_p \cdot (T_m - T_0) + L_f) \cdot \rho \cdot \sqrt{\pi} \cdot v_{\text{scan}} \cdot \alpha \cdot (2w_0)^3} \quad 2-12$$

Rubenchik et al. [133] introduced an approach based on those dimensionless parameters combined with the Eagar-Tsai heat conduction model [134] of a moving Gaussian heat source on a semi-infinite substrate. The extension of Rubenchik et al. [133] was the transfer to calculate a dimensionless temperature field with a normalized spatial coordinate system. The normalization of the coordinates was implemented according to:

$$x^* = \frac{x}{w_0}, y^* = \frac{y}{w_0}, z^* = \frac{z}{\delta_{\text{diff}}} \text{ with } \delta_{\text{diff}} = \sqrt{\frac{\alpha \cdot 2w_0}{v_{\text{scan}}}} \quad 2-13$$

In this dimensionless coordinate system, the function g represents the dimensionless temperature field as:

$$g = \int_0^\infty \frac{\exp\left(-\frac{z^{*2}}{4t} - \frac{y^{*2} + (x^* - t)^2}{\frac{4\sqrt{2}t}{Pe} + 1}\right)}{(\frac{4\sqrt{2}t}{Pe} + 1)\sqrt{t}} dt \quad 2-14$$

The melt pool boundary within g can be determined via the condition:

$$g = \frac{\pi}{\sqrt[4]{2^3} \cdot \frac{\Delta H}{h_s}} \quad 2-15$$

The equations shown provide the basis for estimating the melt pool dimensions based on dimensionless process parameters and pure heat conduction. However, the model is based on and has only been tested for single tracks produced with Gaussian beam profiles. An extension and validation regarding the suitability of this basic approach for modeling the PBF-LB/M process with additional process-specific boundary conditions, such as rapid scanning of adjacent scan tracks and the use of beam shaping, is missing in the current literature.

3 Research approach

The following section outlines the content of the doctoral thesis. First, the research gap is identified, and the scientific objective is defined. This is followed by a description of the approach and structure of the thesis.

3.1 Research gap and scientific objective

The state of the art indicates that the PBF-LB/M process reaches its limits in customizing certain process properties and resulting component properties when employing standard process strategies with Gaussian beam profiles. This hinders the further establishment of PBF-LB/M as an industrially relevant manufacturing process. Moreover, the state of the art reveals a potential for targeted process alterations and new processing possibilities through an adjusted energy input by using beam shaping techniques. However, partly contradictory observations are described in the literature. For example, the microstructures produced using beam shaping are, in some cases, reported to be finer [64, 65, 114], in other cases, to be coarser [118, 120] compared to standard exposure using Gaussian beam profiles. Similarly, with regard to spatter formation, some studies show a reduction in the amount of spatter [104, 105], while others report a higher number of spatters [107] when applying ring-shaped beam profiles.

The partly contradictory findings suggest that the fundamental understanding of the influence of beam shaping in PBF-LB/M is limited. One reason for this is probably the lack of comparability between some of the processes investigated. In general, *a fundamental understanding of the relationship between beam profile characteristics, a modeled process signature, and the dynamic process behavior, as well as the resulting component properties*, is missing in the current literature. Filling this gap is the objective of this doctoral thesis. The work focuses on utilizing ring-shaped beam profiles, as these can be implemented in PBF-LB/M systems with little effort and thus have one of the highest technology readiness levels.

To achieve the objective, five research questions (RQs) are answered:

- RQ1 How do ring-shaped beam profiles influence melt track geometries and surface topology?
- RQ2 Which methods and simplified models enable the prediction of the melt pool geometries and the estimation of the melting mode during PBF-LB/M with ring-shaped beam profiles?
- RQ3 How can the process results of PBF-LB/M with ring-shaped beam profiles be compared independently of beam profile and size?

RQ4 How do ring-shaped beam profiles influence the process dynamics during PBF-LB/M?

RQ5 How can ring-shaped beam profiles modify the as-built microstructure of PBF-LB/M-manufactured components?

The answers to the research questions are elaborated using a hierarchical approach, presented in the following section.

3.2 Approach and structure

The overall goal of the doctoral thesis is to gain *a fundamental understanding of the relationship between beam profile characteristics, a modeled process signature, and the dynamic process behavior, as well as the resulting component properties*. For this, hierarchical investigations of the influence of the ring-shaped beam profiles on different process, component, and material properties are conducted. The approach can be divided into five solution modules (SM), which can be grouped into three phases (see Figure 14).

Phase 1: At the time of starting the doctoral thesis, the state of the art in beam shaping for PBF-LB/M was quite limited. Therefore, the basic concept of beam shaping influencing the PBF-LB/M process is proven first:

SM1 Starting with single-track and single-layer bare plate experiments, melt track cross-sections are characterized in terms of the cross-sectional melt track shape and dimensions. The data generated is used to derive the fundamental potential of different ring-shaped beam profiles in influencing the process in terms of productivity and reliability.

Phase 2: The comparability of processes with beam shaping is limited in the literature. Therefore, an extended semi-analytical process model is implemented, providing a basis for comparing processes with different beam shapes. The approach closes the gap between computationally expensive high-fidelity process models and oversimplified metrics like the volume energy density:

SM2 Different approaches for simple process modeling on a single-track basis, using dimensionless parameters, are examined. The underlying concept is to extend existing models to incorporate ring-shaped beam profiles. For this purpose, SM2 is divided into two sub-solution modules:

SM2.1 Different modeling approaches are investigated concerning the feasibility of predicting the melt pool dimensions with low computational costs when using different beam profiles in various sizes. Additionally, a model extension is proposed to estimate the resulting melting mode. The modelling approaches are validated on single melt tracks.

SM2.2 At the same time, approaches and methods for characterizing the beam profiles are investigated in order to describe the ring-shaped beam profiles in a suitable manner for the modeling approach.

The result of SM2.1 is a model that allows for predicting the process result in terms of melt pool dimensions and the melting mode (conduction, transition, or keyhole mode), depending on the process parameters, beam profile, and spot size. The appropriate methods for characterizing the beam shape and size are the result of SM2.2.

SM3 In the third solution module, the single-track model from SM2 is extended to the three-dimensional PBF-LB/M process. This extension is realized by considering the rapid scanning of hatched scan tracks. From the parameters used in the model, a new dimensionless process map representation is introduced, allowing for the comparison of process results across different beam profiles and various sizes. To predict the process limits within the dimensionless process map, geometric characteristics of the modeled melt pool dimensions are evaluated. The approach is validated against an independently created database.

Phase 3: Based on the parameters for comparable processes from SM2 and SM3, the influences on process dynamics and the resulting microstructure are investigated exploratively.

SM4 In the fourth solution module, the influence of ring-shaped beam profiles on the process dynamics of PBF-LB/M is investigated. This evaluation focuses on the spatter formation using high-speed imaging and image processing. The result is the quantitatively proven influence of the investigated beam profiles on the spatter formation, representing the process dynamics.

SM5 The fifth solution module examines the impact of ring-shaped beam profiles on microstructural properties of PBF-LB/M manufactured parts. For this, the resulting crystallographic texture and grain sizes are analyzed. With knowledge of the influence of beam profiles on the microstructure, the mutual influence of different emerging microstructures on each other is investigated when segments with distinct microstructures are combined into a single component. This approach enables the evaluation of the potential for local microstructure tailoring in PBF-LB/M.

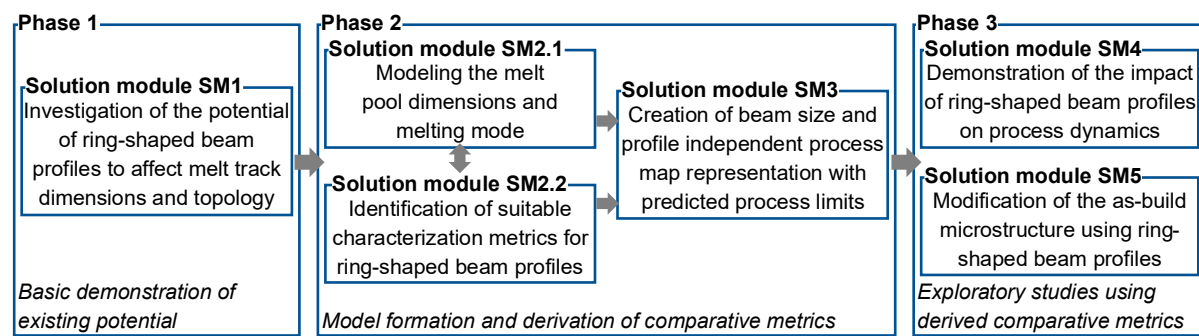


Figure 14: Structure and connection of the solution modules and the corresponding structure of the doctoral thesis.

A key innovation of the research approach in this doctoral thesis is the consistent linking of beam profile characteristics with process and component properties through a modeled process signature derived from a semi-analytical process modelling. Each solution module is represented by a separate peer-reviewed journal paper, meaning each solution module contains an additional novelty factor, which is highlighted in the respective “Research findings” section. The novelties, impact, and limitations of the research approach are evaluated and discussed in the “Critical discussion” section. The following publications are included:

- [I1] J. Grünewald, F. Gehringer, M. Schmöller, K. Wudy, Influence of Ring-Shaped Beam Profiles on Process Stability and Productivity in Laser-Based Powder Bed Fusion of AISI 316L, *Metals* 11 (2021) 1989. <https://doi.org/10.3390/met11121989>. **Main content of SM1**
- [I2] J. Grünewald, M. Wittemer, K. Wudy, Scaling laws for ring-shaped beam profiles in laser-based powder bed fusion of metals, *Advances in Industrial and Manufacturing Engineering* 10 (2025) 100164. <https://doi.org/10.1016/j.aime.2025.100164>. **Main content of SM2**
- [I3] J. Grünewald, K. Wudy, Dimensionless process windows in laser-based powder bed fusion of AISI 316L using ring-shaped beam profiles, *Additive Manufacturing Letters* 14 (2025) 100284. <https://doi.org/10.1016/j.addlet.2025.100284>. **Main content of SM3**
- [I4] J. Grünewald, J. Reimann, K. Wudy, Influence of ring-shaped beam profiles on spatter characteristics in laser-based powder bed fusion of metals. *Journal of Laser Applications* 35 (2023) 042009. <https://doi.org/10.2351/7.0001153>. **Main content of SM4**
- [I5] J. Grünewald, R. Pfund, D. Röhrer, L. Hitzler, J. Torgersen, K. Wudy, Tailored microstructures in laser-based powder bed fusion of 316L using beam shaping: From microstructure formation in individual specimens to their transition in segmented parts, *Additive Manufacturing* (2025) 105026. <https://doi.org/10.1016/j.addma.2025.105026>. **Main content of SM5**

4 Materials and methods

This section describes the experimental boundary conditions for all included studies. First, the feedstock material AISI 316L is briefly described. This is followed by a presentation of the test setup and the system technology used. The specific methods used for the individual studies are described in the “Research findings” section and the attached publications.

4.1 Material

For the investigations, the austenitic stainless steel AISI 316L was used as exemplary material. The material is well investigated in additive manufacturing, especially in PBF-LB/M. Due to its processability and high corrosion resistance, AISI 316L is used in a wide range of applications, e.g., chemical and oil industries, as well as in jewelry and medical technology.

Gas-atomized powder from two different suppliers was used in the investigations of solution modules SM2, SM3, SM4, and SM5. The powder particle size distribution was determined using laser diffraction (Mastersizer 2000, Malvern Panalytical Ltd, Malvern, United Kingdom) and static automated imaging (Morphologi 4, Malvern Panalytical Ltd, Malvern, United Kingdom). Both powders exhibited comparable particle size distributions. The powder used in SM2 and SM4 (MetcoAdd 316L-A, Oerlikon Metco Europe GmbH, Raunheim, Germany) had particles with diameters of $d_{10,3} = 19.0 \mu\text{m}$, $d_{50,3} = 29.6 \mu\text{m}$, and $d_{90,3} = 45.7 \mu\text{m}$ (see Figure 15a). The powder used in SM3 and SM5 (LPW-316-1, LPW Technology Ltd., Runcorn, United Kingdom) exhibited a particle size distribution with diameters of $d_{10,3} = 19.5 \mu\text{m}$, $d_{50,3} = 28.7 \mu\text{m}$, and $d_{90,3} = 44.8 \mu\text{m}$ (see Figure 15a). Additionally, the SEM image in Figure 15b reveals the spherical shape of the powder.

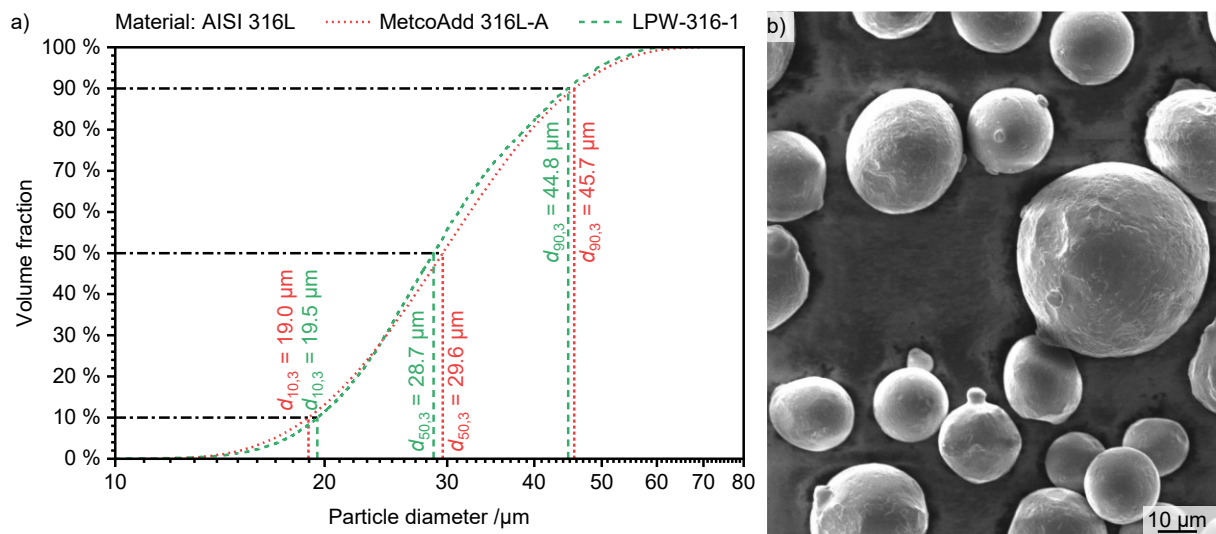


Figure 15: Characteristics of the AISI 316L powder feedstock material: a) particle size distributions of MetcoAdd 316L-A and LPW-316-1, and b) SEM image of LPW-316-1.

For the calculation of the dimensionless parameters in solution modules SM2, SM3, SM4, and SM5, the thermophysical properties of AISI 316L are required. These were assumed as summarized in Table 1 in accordance with Mills [135].

Table 1: Thermophysical properties of AISI 316L according to Mills [135].

Property	Value Unit
Melting temperature T_m	1691 K
Ambient temperature T_0	293.15 K
Density ρ	7269 kg/m ³
Specific heat capacity c_p at T_m	700 J/kg K
Thermal conductivity k at T_m	31.1 W/m K
Thermal diffusivity α at T_m	$6.11 \cdot 10^{-6}$ m ² /s
Latent heat of fusion L_f	$2.60 \cdot 10^5$ J/kg
	$1.89 \cdot 10^9$ J/m ³
Enthalpy at melting h_s	$9.00 \cdot 10^9$ J/m ³

4.2 Experimental framework

A modular PBF-LB/M test bench was designed and built for the investigations covered in this work (see Figure 16). The base of this setup is a welding table (Bernd Siegmund GmbH, Oberottmarshausen, Germany) containing a control cabinet. The control cabinet includes the hardware components for controlling all axes and safety-related releases, as well as further digital and analog inputs and outputs (Beckhoff Automation GmbH & Co. KG, Verl, Germany). The system is designed as a soft PLC operating via an additional industrial PC (Beckhoff Automation GmbH & Co. KG, Verl, Germany).

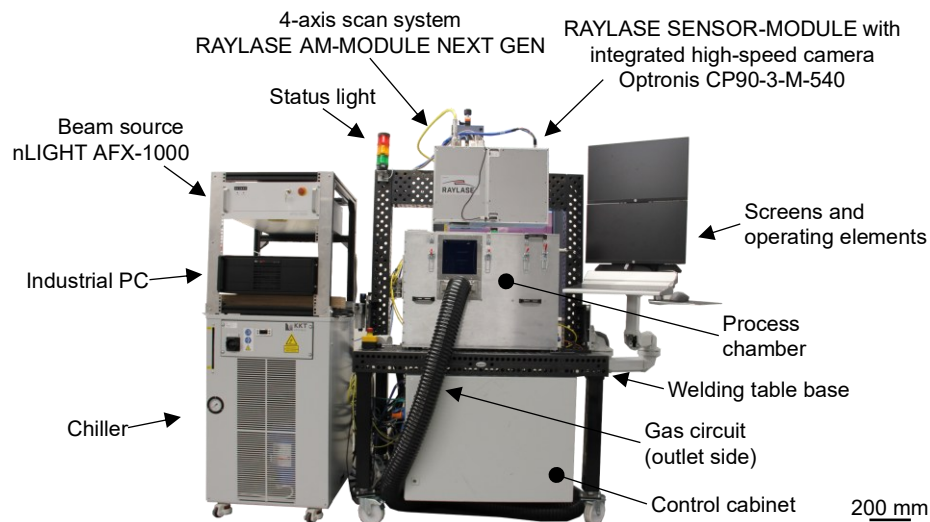


Figure 16: Labeled photograph of the experimental setup used within the scope of this doctoral thesis.

To ensure the PBF-LB/M in-process flow and enable multilayer experiments, custom control software has been designed and implemented on the industrial PC. This software is based on an open-source machine control framework (Autodesk Machine Control

Framework, Autodesk Inc., San Rafael, CA, USA) and can be operated via separately attached operating elements. With the software, it is possible to execute multilayer experiments under industrial conditions as build jobs, which are completely processed via a predefined 3MF file.

The core of the test bench is the interchangeable process chamber. Different process chambers were used depending on the focus of the investigation. Due to the lower powder demand and the adjusted substrate size, a miniaturized process chamber was used in the single-track and single-layer experiments SM1, SM2, and SM4 (see Figure 17a). For the studies in SM3 and SM5, which involved the manufacturing of test specimens, a process chamber design was used on a scale comparable to industrial production (see Figure 17b). The concept of this chamber is based on a state-of-the-art PBF-LB/M system with powder dosing via a supply platform. Recoating is performed using a rubber lip analogous to the system of the machine manufacturer TRUMPF (TRUMPF Laser- und Systemtechnik GmbH, Ditzingen, Germany; TRUMPF Additive Manufacturing Italia s.r.l., Schio, Italy). The shielding gas flow is directed perpendicular to the recoating direction.

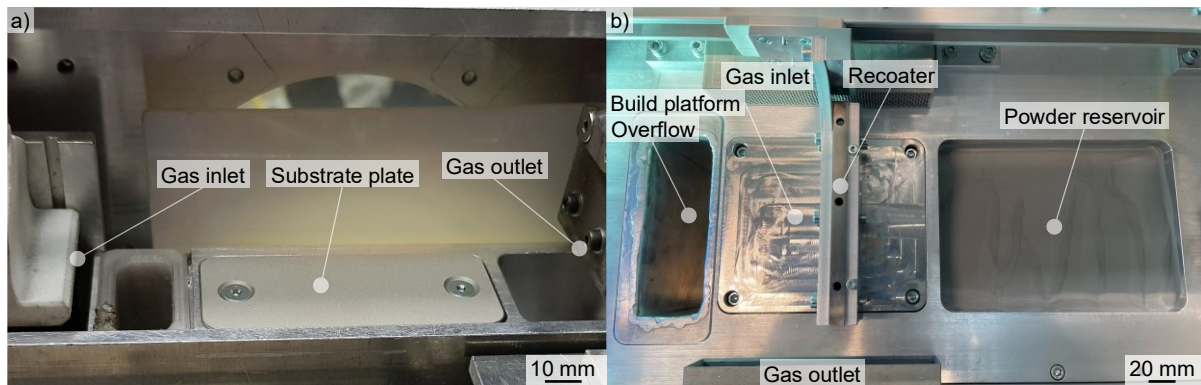


Figure 17: Labeled photo of the experimental process chamber layouts, a) miniaturized chamber, including a substrate plate and a gas inlet and outlet, adapted from [I1], b) industrial scale chamber, including a build platform, a powder reservoir with a supply platform, an overflow, a recoater, and a gas inlet and outlet, adapted from [I3].

Similarly, the scanning system is designed to be interchangeable. In the preliminary study in SM1, a state-of-the-art scanning system with three optical axes (Elephant 50, ARGES GmbH, Wackersdorf, Germany) was used for laser deflection and focusing. For the studies in SM2, SM3, SM4, and SM5, a combined scanning system (AM-MODULE NEXT GEN and SENSOR-MODULE, RAYLASE GmbH, Wessling, Germany) with a total of five optical axes is used to deflect the laser beam, position the focus, and manipulate the laser spot size. The five axes of the scanning system are assigned to an x-axis and a y-axis (galvanometer scanner), two z-axes for collimation and focusing of the high-power laser, which acts as a dynamic telescope to widen the focused beam, and an additional z-axis for focus adjustment of an integrable on-axis camera (see Figure 18). Consequently, the scanning system allows

an on-axis camera to be installed, keeping the working plane in focus coaxially, regardless of the position of the laser spot. A high-speed camera with a monochrome CMOS chip (CP90-3-M-540, Optronis GmbH, Kehl, Germany) is integrated, which enables frame rates of up to 37 700 fps depending on the pixel area read out. Additionally, the scanning system can be adjusted in height via a spindle.

A fiber laser with switchable beam profiles (AFX-1000, nLIGHT Inc., Vancouver, WA, USA) was used as the beam source, coupled to the scanning systems. The laser offers the possibility of distributing a power of up to 1200 W in 7 discrete steps between a single-mode core and a surrounding ring, as outlined in section 2.2. The beam profiles in this thesis are named by the power distribution between the central core and the surrounding ring. Consequently, a beam profile with a power fraction of approximately 20 % in the core and approximately 80 % in the surrounding ring is referred to as beam profile 20/80. Combined with the scanning system, seven different beam profiles in a magnification M , ranging from 1 to 2, can be generated and scanned (see Figure 18). In this way, various laser beam profiles can be created across different sizes with the setup. Determining the characteristics of the beam profiles using specific methods is covered in SM2.2 and is discussed accordingly in the section on the core results of [I2].

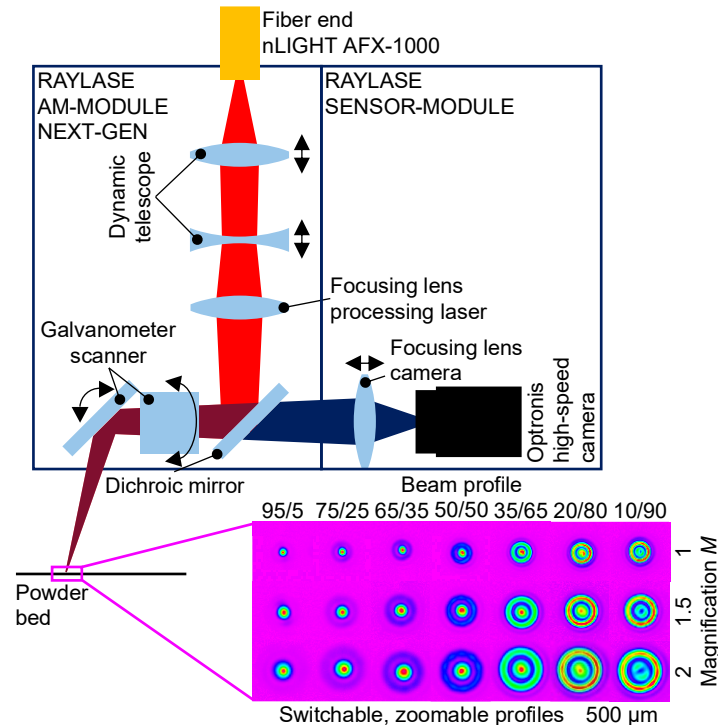


Figure 18: Schematic representation of the combined five-axis scanner system, including the integrated high-speed camera and the possible beam profiles in the various sizes, adapted from [I4].

5 Research findings

The research findings are structured according to the solution modules in Figure 14. Each solution module contains the content of essentially one peer-reviewed journal paper. This section provides an overview of the key results of the corresponding publications. For detailed information, please refer to the reprints provided in the Appendix.

5.1 Influence of ring-shaped beam profiles on melt track dimensions and topology

[I1] J. Grünewald, F. Gehringer, M. Schmöller, K. Wudy, *Influence of Ring-Shaped Beam Profiles on Process Stability and Productivity in Laser-Based Powder Bed Fusion of AISI 316L*, *Metals* 11 (2021) 1989. <https://doi.org/10.3390/met11121989>.

This section includes the key findings of publication [I1]. For detailed information, please refer to the reprint provided in the Appendix.

The subject of solution module SM1 is demonstrating the fundamental potential of ring-shaped beam profiles to influence the melt track dimensions and top surface topology during PBF-LB/M thereby answering research question RQ1. For this purpose, single melt tracks and hatched multi-track segments are manufactured and evaluated. From the generated data, the potential of the investigated beam profiles for accelerating the process and improving reliability through an enlarged process window is derived.

Figure 19 illustrates the process maps derived from single-track experiments for a Gaussian and three ring-shaped beam profiles processing AISI 316L. To determine the process windows, the produced single melt tracks are divided into five process categories:

- Stable melting: The melt track surface is smooth. The aspect ratio of the melt track depth to the melt track width is below 0.8.
- Protrusion and depression: The melt tracks exhibit a central protrusion and depressions on the sides. This cross-sectional melt track shape is created by firm backward melt flows due to the Marangoni effect and recoil pressure [136].
- Balling: The melt track surface is rough with single melt accumulations along the track.
- Undermelting: The melt track depth is below 40 μm . A lack of fusion is expected to occur in three-dimensional test specimens for the corresponding parameter sets.
- Keyholing: The aspect ratio of the melt track depth to the width is above 0.8 [23]. In a three-dimensional test specimen, keyhole porosity is expected for the corresponding parameter sets [18].

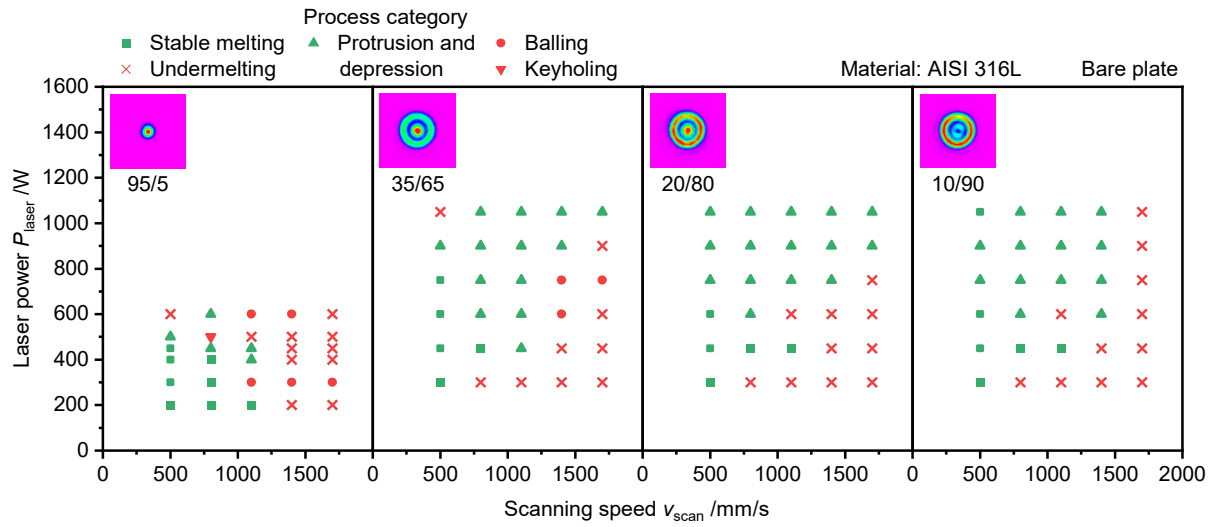


Figure 19: $P_{\text{laser}}-v_{\text{scan}}$ process map with categorized process regimes of single-track experiments on bare plates. When using the ring-shaped beam profiles 35/65, 20/80, and 10/90, the process windows enlarge and shift to higher power levels compared to using the almost Gaussian beam profile 95/5. Figure adapted from [11].

When using the almost Gaussian beam profile 95/5, the upper limit of the process window is at a laser power of 450 W and a scanning speed of 1100 mm/s. Balling effects and undermelting occur at higher scanning speeds. At higher applied powers, there is a risk of entering the keyhole mode. The use of the ring-shaped beam profiles 35/65, 20/80, and 10/90 enables higher laser powers of up to 1050 W and higher scanning speeds of up to 1500 mm/s, with single-track results promising to be transferable to a reliable PBF-LB/M process. No keyholing occurs at the highest tested power level of 1050 W. Furthermore, no balling phenomena are detected for the beam profiles 20/80 and 10/90. However, at low laser powers of 200 W, increasing undermelting occurs when using beam profiles 35/65, 20/80, and 10/90. Overall, the process windows shift to higher laser powers and enlarge significantly when ring-shaped beam profiles are used.

Additionally, the results show that ring-shaped beam profiles produce much wider melt tracks. Figure 20a shows the melt track widths obtained after exposure of hatched areas. For this purpose, the half-width of the melt track was measured in cross-section as the distance from the melt track edge to the deepest point of the melt track cross-section. At similar area energies, the melt tracks are approximately twice as wide when using ring-shaped beam profiles as compared to Gaussian profiles. (see Figure 20a). This enables similar relative overlap values between the individual melt tracks of a hatched area to be achieved using larger hatch distances with similar area energies (see Figure 20b).

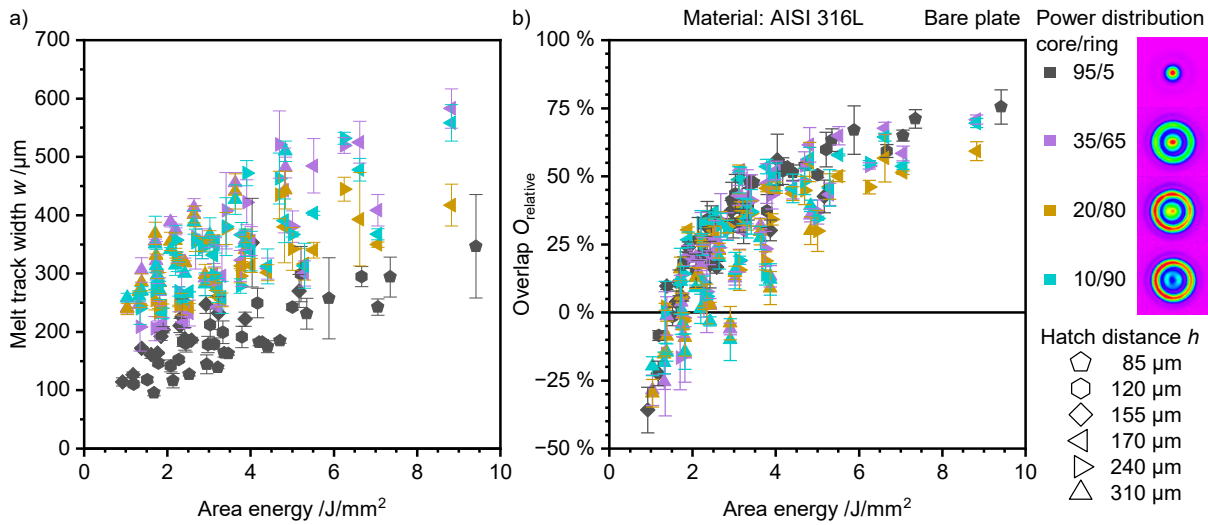


Figure 20: a) Melt track width w and b) relative overlap, defined as $O_{\text{relative}} = (h-w)/w$, as a function of the area energy applied during the exposure of hatched areas. The ring-shaped beam profiles 35/65, 20/80, and 10/90 result in melt tracks approximately twice as wide as those resulting from the Gaussian beam profile 95/5, which generates similar overlap values O_{relative} with twice as large hatch distances h . Data from [11].

Furthermore, using the ring-shaped beam profiles 35/65, 20/80, and 10/90 tends to result in slightly increased roughness at similar area energies (see Figure 21). Notably, despite the hatch distances being twice as large, there is only a slight increase in top surface roughness.

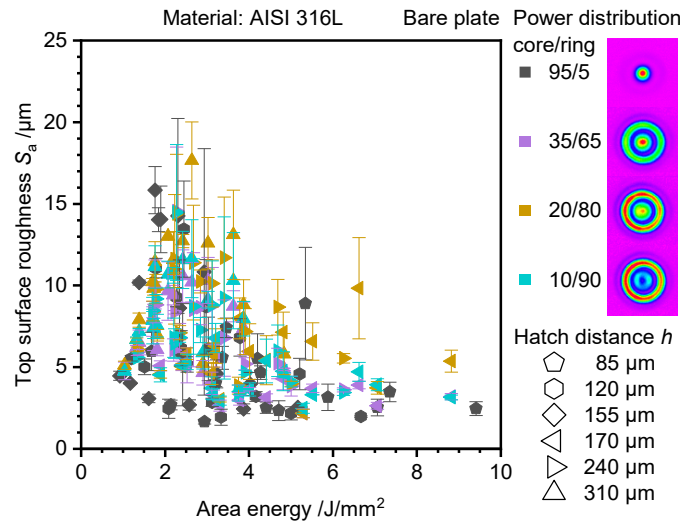


Figure 21: Top surface roughness S_a as a function of the area energy applied during the exposure of hatched areas. When using the ring-shaped beam profiles 35/65, 20/80, and 10/90, the hatched areas show slightly increased roughness compared to using the almost Gaussian beam profile 95/5. Data from [11].

In summary, the observations demonstrate that using ring-shaped beam profiles, PBF-LB/M becomes potentially more reliable with larger process windows and more productive with increased hatch distance and scanning speed, while maintaining similar top surface roughness. In conclusion, ring-shaped beam profiles can influence the melt track geometry and topology, resulting in wider melt tracks with similar top surface roughness in PBF-LB/M.

Publication [I1], published in 2021, was the first paper to investigate the influence of different beam shapes consisting of a Gaussian core and a surrounding ring on melt track dimensions using PBF-LB/M as the targeted process. At the same time, the study examined for the first time the use of a fiber laser with the corresponding multi-core feeding fiber for PBF-LB/M. This gave the study a pioneering character. The key findings of this work on the influence of ring-shaped beam profiles on melt track dimensions and the tendency to balling or humping were later confirmed on other materials, such as the high-strength aluminum alloys EN AW-5083 [87] and AlSi7Mg0.6 [115].

However, [I1] employed a black box approach, using the area energy as a comparative variable, which does not account for any characteristics of the beam profile, such as shape or spot size. Both influence the process result. Therefore, there is a gap in comparability, as the ring-shaped beam profiles have a larger beam parameter product than the diffraction-limited Gaussian laser beam, resulting in larger spot sizes. The following section addresses this comparability issue by investigating suitable scaling laws and methods for determining the spot size of ring-shaped beam profiles.

5.2 Scaling laws for single-track laser-based powder bed fusion of metals with ring-shaped beam profiles

[I2] J. Grünwald, M. Wittemer, K. Wudy, *Scaling laws for ring-shaped beam profiles in laser-based powder bed fusion of metals*, *Advances in Industrial and Manufacturing Engineering* 10 (2025) 100164. <https://doi.org/10.1016/j.aime.2025.100164>.

This section contains key results of publication [I2]. For details, please refer to the reprinted version in the Appendix.

A comparison of the dimensioned individual process parameters (laser power, scanning speed, hatch distance, etc.) or their combinations (line energies, area energy densities, volume energy densities, etc.) is of limited applicability for comparing processes and process outcomes generated with different beam profiles and spot sizes [58]. Therefore, solution module SM2 demonstrates an approach to modeling the single melt track dimensions and the melting mode resulting from PBF-LB/M with ring-shaped beam profiles. The approach utilizes a semi-analytical heat conduction model (SM2.1) combined with the second moment method for determining spot size (SM2.2), thereby addressing research question RQ2.

For preparing solution module SM2.1, the adaptation and extension of various models were investigated, such as the approaches of Fabbro [137] and Gan et al. [138], focusing on the keyholing melting mode, or the approach of Patel and Vlasea [21], based on the work of Ion et al. [126] focusing on the melting mode detection. The Eagar and Tsai [134] approach,

introduced in the state of the art, proved to be the most suitable for modeling the PBF-LB/M melt pool dimensions when using ring-shaped beam profiles. The model was adapted by Rubenchik et al. [133] to calculate a dimensionless temperature field (see Equation 2-14) using the Péclet number (see Equation 2-3) and the dimensionless coordinates x^* , y^* , and z^* (see Equation 2-13). The dimensionless melt pool depth d^* , length l^* , and width w^* (see Figure 22) can be determined from the dimensionless temperature field using the dimensionless enthalpy (see Equation 2-12) by means of the condition in Equation 2-15.

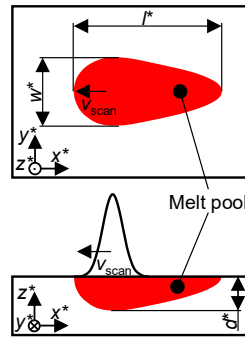


Figure 22: Schematic representation of the dimensionless melt pool depth d^* , length l^* , and width w^* in the dimensionless coordinate system consisting of x^* , y^* , and z^* .

The model is based on heat conduction with a Gaussian heat source, which is used as a representative of the beam profile. The spot size of the respective beam profile is considered when calculating the Péclet number (see Equation 2-3) and the dimensionless enthalpy (see Equation 2-12). To estimate the melt pool width and depth, the actual intensity distribution of the respective beam profile is only considered via the spot size. To maintain the semi-analytical nature of the equations used, the shape of the heat source in the model remains Gaussian. Therefore, the spot size is particularly important for appropriate modeling. However, various methods have been established for determining the spot size, such as the full width at half maximum (FWHM) or further threshold-based methods (e.g., $1/e$ and $1/e^2$), Gaussian fit, knife edge, moving slit, 86 % method, or the statistical second moment method. The investigation of which spot size determination method best suits the heat conduction model for modeling the melt pool dimensions is part of solution module SM2.2.

Figure 23 illustrates the exemplary influence of the respective method on the determined spot size of different beam profiles. When determining the spot sizes of Gaussian-like beam profiles, such as beam profile 95/5 (see Figure 23a), the differences in the results are small. When determining the spot sizes of ring-shaped beam profiles, the determined spot sizes vary significantly depending on the method used (see Figure 23b and Figure 23c). The differences in the determined spot sizes directly influence the quality of the model applied, as the calculation of the Péclet number and the dimensionless enthalpy are affected.

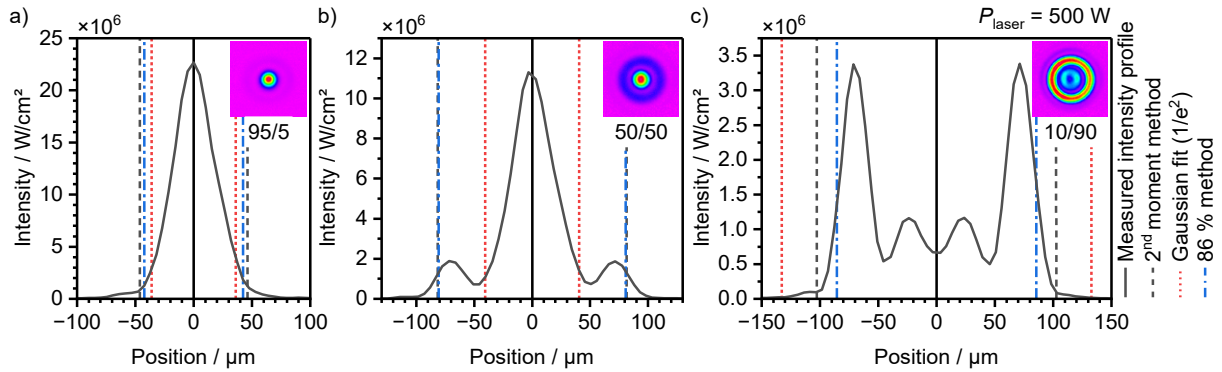


Figure 23: Comparison of the results of the various spot size determination methods: Second moment method, Gaussian fit (with the 1/e² value), and 86 % method using the example of the a) almost Gaussian beam profile 95/5, b) beam profile 50/50, and c) beam profile 10/90. Depending on the method used, the determined spot sizes differ significantly from one another, especially for the beam profiles 50/50 and 10/90. Figure adapted from [12].

Figure 24 presents the modeled melt track depths (see Figure 24a) and widths (see Figure 24b), comparing them with the dimensions of experimentally produced single melt tracks. The experimental data are represented as data points with dimensionless enthalpy calculated according to Equation 2-12 using the second moment method for spot size determination. The modeled data are plotted as line graphs. Since the melt track widths and depths depend not only on the dimensionless enthalpy plotted on the abscissa but also on the Péclet number, three different modeled Péclet numbers are shown. These are the minimum and maximum Péclet numbers examined experimentally, Pe_{min} and Pe_{max} , as well as the mean Péclet number from the experimental database, Pe_{mean} .

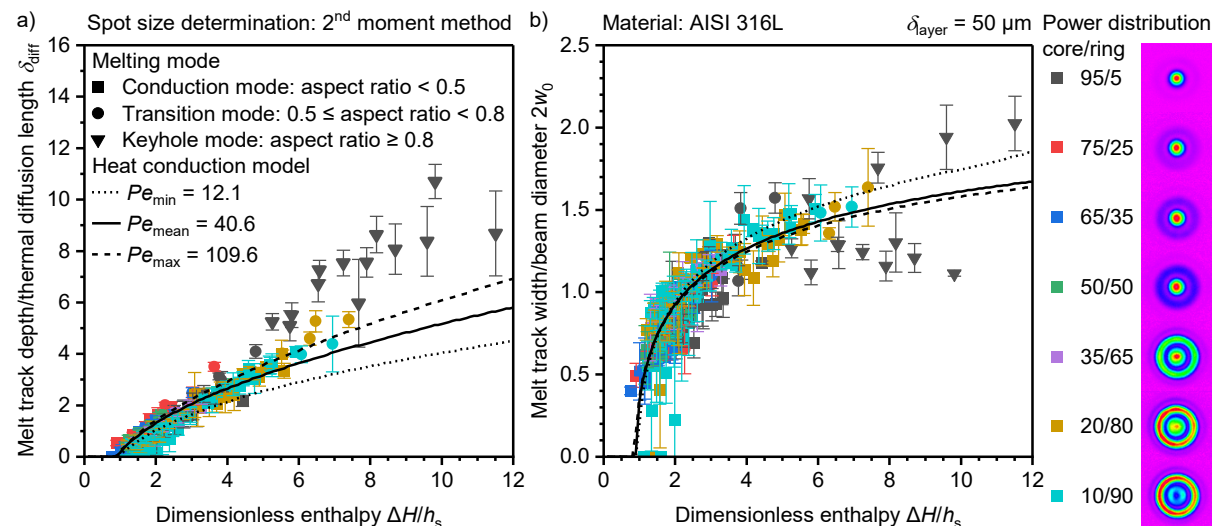


Figure 24: a) Dimensionless melt pool depth and b) dimensionless melt pool width as a function of the dimensionless enthalpy for different beam profiles. The line graphs represent semi-analytically modeled data for different Péclet numbers, while the data points represent experimentally generated values. When using the second moment method to determine the spot size, the data points converge to a curve and match with the applied heat conduction model. Only when entering the keyhole mode, larger deviations between the model and experimental data occur. Figure adapted from [12].

Two key findings can be deduced from the results. On the one hand, the data points of the experimental data for the different beam profiles converge to a single curve. This indicates that the second moment method is suitable for determining the spot size to compare processes and process results (SM2.2). On the other hand, the line graphs and data points match. Larger deviations between the model and experiment become apparent as soon as the detected melting mode of the experimental data enters the keyhole mode (see triangles in Figure 24). This is expected since the model is based on heat conduction. The observations demonstrate that the model approach of Rubenchik et al. [133] is suitable for modeling the melt track depth and width when applying ring-shaped beam profiles (SM2.1), provided the process regime is heat conduction-driven and the second moment method is used for spot size determination. For further details regarding the suitability of further spot size determination methods, please refer to [12].

However, although the model works satisfactorily for predicting the depth and width of a melt track, the melting mode cannot be estimated directly from the value of the dimensionless enthalpy. When using the beam profiles 75/25, 65/35, and 50/50, the melt tracks are slightly deeper and narrower than predicted by the model. When using the beam profiles 20/80 and 10/90, the melt tracks become slightly shallower and wider than in the model. In combination, the use of beam profiles with a low intensity peak (beam profiles 20/80 and 10/90) results in a lower probability of keyholing than the use of beam profiles with a high intensity peak (beam profiles 75/25, 65/35, and 50/50). For this reason, the maximum dimensionless enthalpy $\Delta H_{\max}/h_s$, which includes the peak-to-mean ratio ψ of the beam profile, is introduced as:

$$\frac{\Delta H_{\max}}{h_s} = \frac{\psi \Delta H}{2 h_s} \quad 5-16$$

The peak-to-mean ratio describes the ratio of the highest intensity peak measured I_{peak} of a beam profile to its mean intensity I_{mean} , which is determined by the laser power and the spot size according to:

$$\psi = \frac{I_{\text{peak}}}{I_{\text{mean}}} = \frac{I_{\text{peak}} \cdot \pi \cdot w_0^2}{P_{\text{laser}}} \quad 5-17$$

Figure 25 illustrates exemplary intensity distributions modeled for calculating the dimensionless enthalpy (yellow dashed line in Figure 25) and the maximum dimensionless enthalpy (green dotted line in Figure 25) for three beam profiles with varying peak-to-mean intensity ratios. Using the peak-to-mean ratio, the modeled peak intensity and the total power are adjusted by a factor of $\psi/2$.

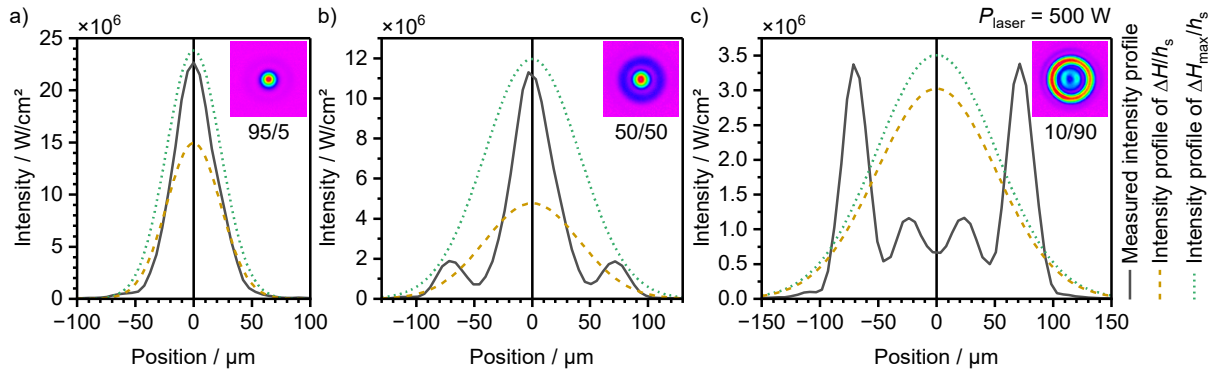


Figure 25: Comparison of the modeled intensity distributions for calculating the dimensionless enthalpy $\Delta H/h_s$ and the maximum dimensionless enthalpy $\Delta H_{\max}/h_s$ when using the a) almost Gaussian beam profile 95/5, b) beam profile 50/50, and c) beam profile 10/90. By considering the peak-to-mean intensity ratio in $\Delta H_{\max}/h_s$, the modeled peak intensity is adjusted to the level of the measured peak intensity. Figure adapted from [12].

Figure 26 demonstrates the effect of the introduced maximum dimensionless enthalpy, as defined in Equation 5-14, on the aspect ratio of the melt track cross-sections (melt track depth to melt track width). Depending on the ratio of melt depth to melt width, the melt modes are defined as in the “State of the art” section:

- Conduction mode for aspect ratios lower than 0.5
- Transition mode for aspect ratios between 0.5 and 0.8 [21]
- Keyhole mode for aspect ratios above 0.8 [23]

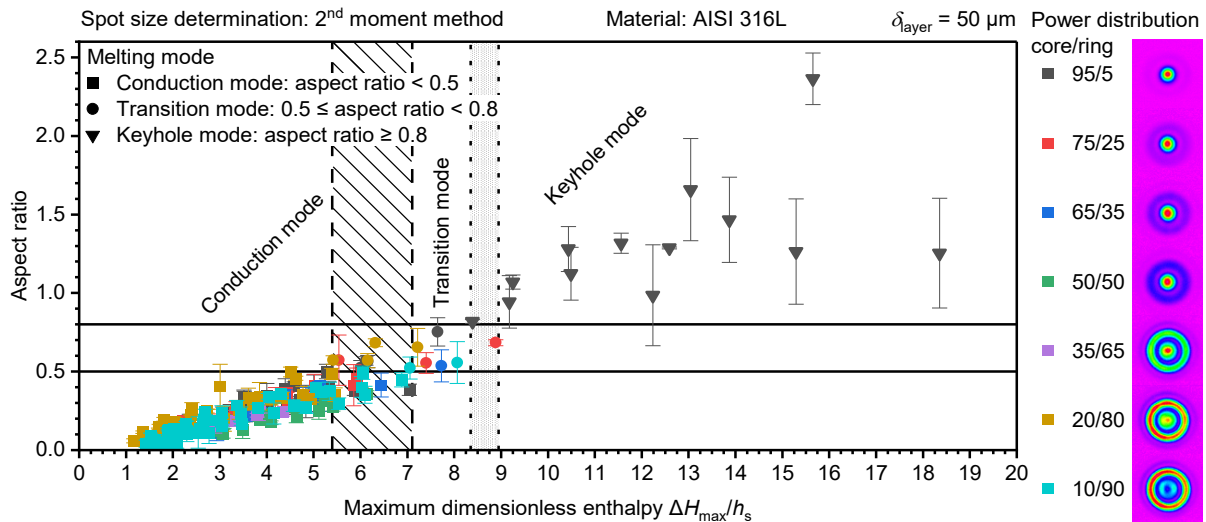


Figure 26: Aspect ratio as a function of the maximum dimensionless enthalpy. The crossover from conduction to transition mode is at a maximum dimensionless enthalpy of 6.25 ± 0.85 (see the dashed area). At a maximum dimensionless enthalpy of 8.65 ± 0.30 , the crossover to keyhole mode occurs (see dotted area). Figure adapted from [12].

By using the maximum dimensionless enthalpy instead of the dimensionless enthalpy, the data points are shifted on the abscissa depending on the peak-to-mean ratio, so that the melting mode changes in similar value ranges for different beam profiles. The melting mode crossover from conduction mode to transition mode is at a maximum dimensionless enthalpy

of 6.25 ± 0.85 (see shaded area in Figure 26). The transition-to-keyhole mode crossover occurs at a maximum dimensionless enthalpy of 8.65 ± 0.30 (see dotted area in Figure 26). Consequently, considering the peak-to-mean intensity ratio enables the estimation of the melting mode for different beam profiles with known process parameters.

In conclusion, SM2 includes two novelties. First, it is demonstrated that the approach for calculating a dimensionless temperature field, as proposed by Rubenchik et al. [133], is suitable to model the melt track width and melt track depth when using ring-shaped beam profiles. The prerequisite for the modeling approach is to characterize the ring-shaped beam profiles using the second moment method. The literature available at the time of publication contained simple approaches that could estimate melt track widths and depths based on dimensioned parameters, such as [139]. However, these lacked a dimensionless or scaling metric and therefore allowed only limited comparability between different processes. Second, the extension of the proposed approach to include a peak-to-mean intensity ratio enables the estimation of the melting mode. Until publication of [12], melting mode estimation was only possible for Gaussian-like beam profiles, as in [20, 21, 130]. This means that considering the influence of beam profile characteristics beyond the spot size in this context is a novel approach.

In summary, new approaches for estimating the resulting melt pool dimensions and melting modes when using beam shaping were introduced. The application of the presented approaches can significantly reduce the experimental effort or, in the case of a simulation-based study, the computational costs by an initial evaluation of the melt pool geometries and melting mode when investigating new beam profiles. However, in the presented stage of solution SM2, the findings are limited to single-track analyses. An extension to the three-dimensional PBF-LB/M process is the subject of solution module SM3.

5.3 Dimensionless process maps for laser-based powder bed fusion of metals with ring-shaped beam profiles

[13] J. Grünewald, K. Wudy, *Dimensionless process windows in laser-based powder bed fusion of AISI 316L using ring-shaped beam profiles*, *Additive Manufacturing Letters* 14 (2025) 100284. <https://doi.org/10.1016/j.addlet.2025.100284>.

This section presents the key findings of [13]. Detailed information on the included study can be found in the Appendix.

SM2 demonstrates the feasibility of modeling the melt pool size in PBF-LB/M on a single-track basis using the dimensionless enthalpy and the Péclet number combined with a semi-analytical heat conduction model. The model is valid as long as the process regime is heat

conduction-driven. However, the temperature field and the process result of a single-track exposure cannot be directly transferred to a three-dimensional PBF-LB/M process. Since the individual melt tracks adjacent to each other are exposed in rapid succession during hatching, the resulting temperature field and thus the process result are influenced. Therefore, the objective of solution module SM3 is, on the one hand, to extend the existing single-track model to cover hatching. On the other hand, a new process map representation independent of spot size and beam profile is introduced based on the parameters used in the model. Within these process maps, the process window limits are predicted based on geometric characteristics of the melt pools derived from the modeled temperature fields.

To consider the preheating effect of the process zone by adjacent previously exposed melt tracks, Equation 2-14 is extended by two terms to calculate the dimensionless temperature field. A bidirectional pattern with a mean scan vector length $\bar{x}$ is assumed for the exposure. Since the x and y coordinates of the dimensionless temperature model are normalized by the spot radius, the dimensionless hatch distance h^* (introduced by Thomas et al. [129]) and the dimensionless scan vector length $\bar{x}^*$ are normalized according to:

$$h^* = \frac{h}{w_0} \quad 5-18$$

$$\bar{x}^* = \frac{\bar{x}}{w_0} + h^* \quad 5-19$$

Considering the adjacent melt tracks, the dimensionless temperature field is calculated as:

$$g = \int_0^\infty \frac{\left(\exp\left(-\frac{z^{*2}}{4t} - \frac{y^{*2} + (x^* - t)^2}{\frac{4\sqrt{2}t}{Pe} + 1}\right) + \exp\left(-\frac{z^{*2}}{4t} - \frac{(y^* + h^*)^2 + (-x^* + \bar{x}^* - t)^2}{\frac{4\sqrt{2}t}{Pe} + 1}\right) + \exp\left(-\frac{z^{*2}}{4t} - \frac{(y^* + 2h^*)^2 + (x^* + 2\bar{x}^* - t)^2}{\frac{4\sqrt{2}t}{Pe} + 1}\right) \right)}{(\frac{4\sqrt{2}t}{Pe} + 1)\sqrt{t}} dt \quad 5-20$$

Figure 27 shows an example of a dimensionless temperature field resulting from Equation 5-20 in the zero plane of the z^* coordinate. The melt pool is marked by a dark red color. The directly adjacent melt track (second term) has a noticeable influence on the temperature distribution. The contribution of the third term is significantly smaller, indicating that it is feasible to not consider any further adjacent scan tracks.

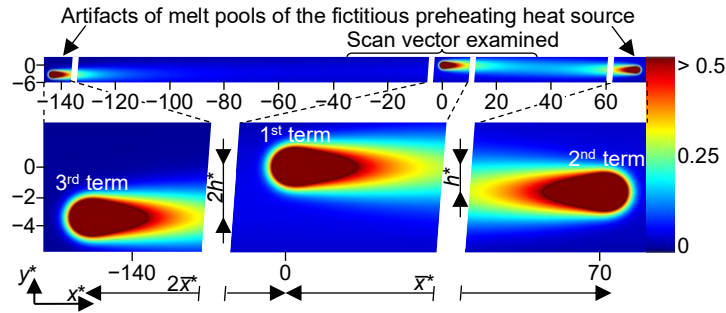


Figure 27: Example of a calculated dimensionless temperature field according to Equation 5-18. For the calculation, a Péclet number Pe of 15, a dimensionless hatch distance h^* of 1.7, and a dimensionless mean scan vector length $\bar{x}^*$ of 71.7 were used, which would result in a scan vector length $\bar{x}$ of 7 mm and a spot diameter $2w_0$ of 200 μm . The melt pools are indicated by the dark red color, corresponding to a dimensionless enthalpy $\Delta H/h_s$ of 3.5. Figure adapted from [13].

To estimate the process window limits, the dimensions of the melt pool at the coordinate origin of the calculated temperature field are analyzed in detail. For this purpose, the modeled melt pool width, depth, and length are derived using the condition following Equation 2-15. A minimum depth of 5 μm at the overlap between two adjacent melt tracks (see d_h^* in Figure 28) is defined as the criterion for lack of fusion porosity. Expressed by the dimensionless melt pool characteristics, this can be written as:

$$\frac{d^* \cdot w_0}{\sqrt{Pe}} \cdot \sqrt{1 - \frac{h^{*2}}{w^{*2}}} \geq 5 \mu\text{m} \quad 5-21$$

For the process window limit regarding bulging top surfaces, the balling criterion is specified as defined by Kruth et al. [29]. According to this criterion, the surface tension of a melt pool with an approximate half-cylinder shape breaks up the melt pool into spherical accumulations for the length-to-width ratios above 2.1. Expressed in dimensionless parameters, the processes without balling result for the condition:

$$\frac{l^*}{w^*} \leq 2.1 \quad 5-22$$

The parameters used to calculate the dimensionless temperature field serve as coordinates for dimensionless process maps. As in conventional $P_{\text{laser}} \cdot v_{\text{scan}}$ process maps, a two-dimensional representation is achieved by utilizing an energy dimension on the ordinate and a velocity dimension on the abscissa. For the energy dimension, the dimensionless enthalpy is divided by the dimensionless hatch. In this way, the amount of energy introduced into a specific volume is considered, as well as the accumulation of energy depending on the hatch distance used. For the velocity dimension, a dimensionless area rate is generated by considering the product of the Péclet number and the dimensionless hatch distance.

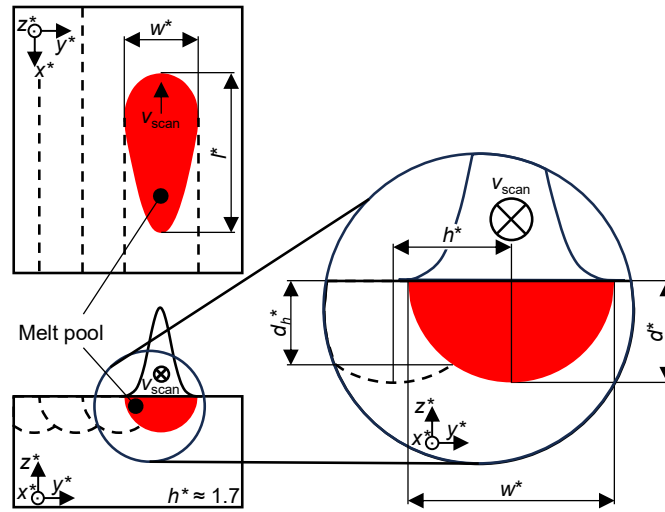


Figure 28: Schematic representation of the PBF-LB/M process zone during exposure with indicated previously exposed melt tracks in a dimensionless coordinate system with the marked dimensionless melt pool characteristics melt pool depth d^* , depth at the half hatch distance d_h^* , length l^* , and width w^* , as well as the dimensionless hatch distance h^* . Figure adapted from [13].

Figure 29 and Figure 30, respectively, show the resulting dimensionless process maps from a PBF-LB/M process of AISI 316L using seven different beam profiles. Figure 29 shows the optically determined relative densities, while Figure 30 shows the measured arithmetic mean height values S_a of the top surface of the specimens. The plotted data points represent the measured values obtained from manufactured cuboidal test specimens with dimensions of $7 \times 7 \times 10 \text{ mm}^3$. In the experiments, the laser power, scanning speed, and hatch distance were varied. For details, please refer to publication [13].

The modeled process window limits according to Equations 5-21 and 5-22 are plotted as black line graphs for different hatch distances. The levels cover the experimentally investigated value range of the dimensionless hatch distance.

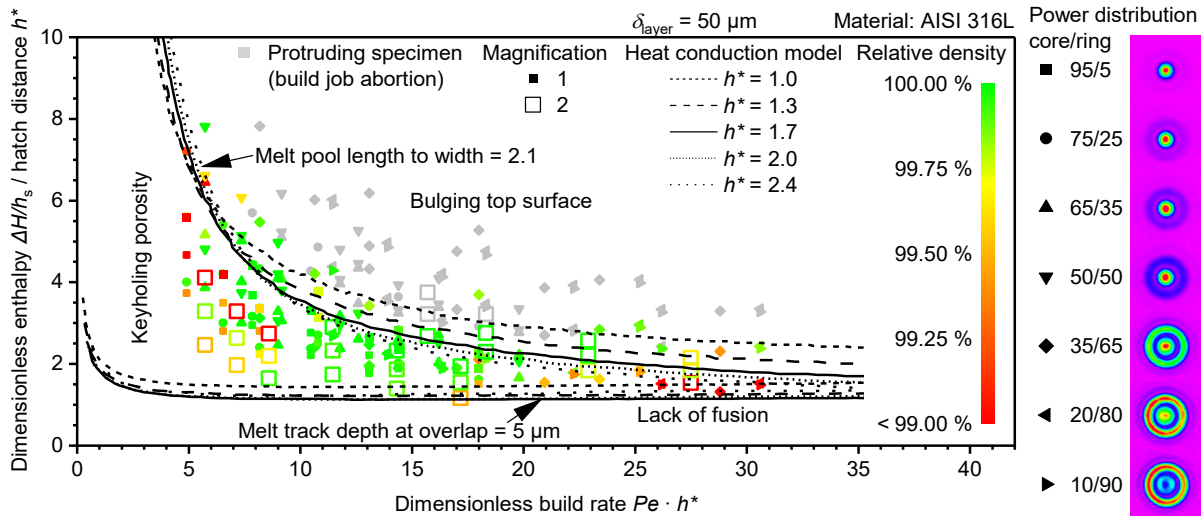


Figure 29: Dimensionless process map based on the dimensionless enthalpy $\Delta H/h_s$ per dimensionless hatch distance h^* and the dimensionless build rate $Pe \cdot h^*$ for relative specimen density. The process window limits predicted by Equations 5-21 and 5-22 are shown as line graphs for different dimensionless hatch distances h^* . Figure adapted from [13].

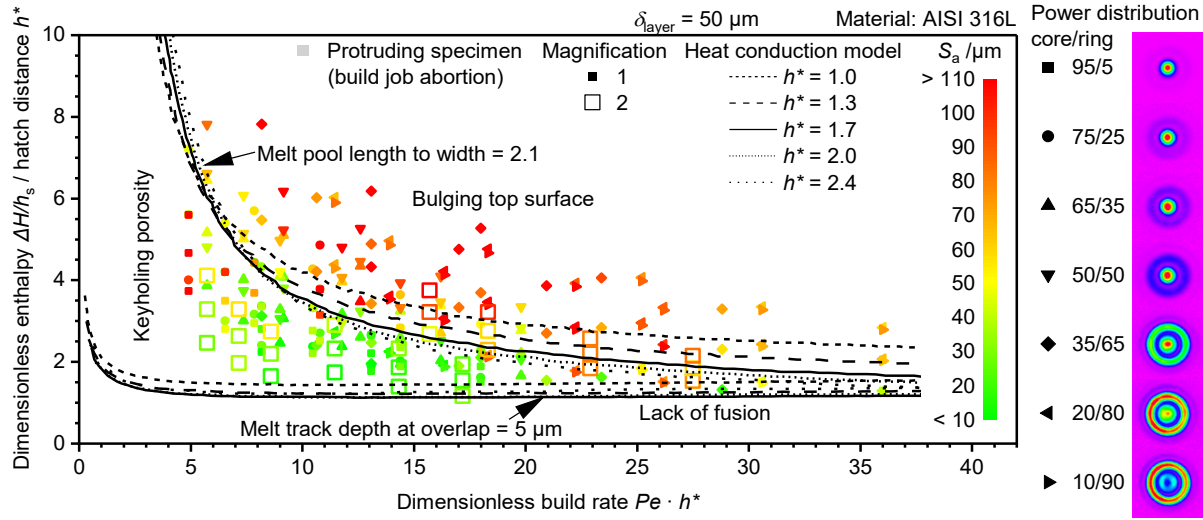


Figure 30: Dimensionless process map based on the dimensionless enthalpy $\Delta H/h_s$ per dimensionless hatch distance h^* and the dimensionless build rate $Pe \cdot h^*$ for the arithmetic mean height S_a of the top surface. The process window limits predicted by Equations 5-21 and 5-22 are shown as line graphs for different dimensionless hatch distances h^* . Figure adapted from [13].

The process window limit in the direction of lack of fusion lies within a range of a dimensionless enthalpy per dimensionless hatch of approximately 1.5 for both the modeled melt pools according to Equation 5-21 and the experimental data (see Figure 29). Above a dimensionless build rate of approximately 25, the corresponding specimens exhibit increased roughness. Below this value, the specimens with a lack of fusion porosity are smooth (see Figure 30).

In the direction of bulging specimens, the process window limit of the modeled processes according to Equation 5-22 overlaps well with the experimental data. Above the predicted limit, some specimens with high densities are found (see Figure 29). However, these exhibit increased roughness values of $S_a > 70 \mu\text{m}$ (see Figure 30). Due to surface tension-driven effects, components with very wavy surfaces and high density occur from a certain threshold until the waviness reaches a level where the build jobs of the respective parts must be aborted due to bulging. The possibility of predicting this limit via the shape of the melt pool in the heat conduction model using the condition from Equation 5-22 shows that the effect is independent of the beam profile in the parameter space investigated.

However, another interesting effect can be observed. When using smaller dimensionless hatch distances, the process window limit shifts upwards, enlarging the overall process window to higher energy inputs and build rates (see process window limits for h^* of 1.0). In many studies, the dimensionless hatch distance is selected to be smaller when ring-shaped beam profiles are used, although this is not explicitly emphasized. In [35], it is summarized that the hatch distance remains within a range of values typical for state-of-the-art processes (such as $70 \mu\text{m}$ [115, 116] or $90 \mu\text{m}$ [121]) when using larger ring-shaped beam profiles. This

means there is potential to increase productivity by using small dimensionless hatch distances combined with high Péclet numbers. In contrast to the state of the art, the dimensionless approach presented in [13] enables the isolation of the shown influences of beam shape and spot size from each other, allowing for their consideration separately, even when both influencing factors are varied simultaneously.

The process limit for keyhole porosity depends on the beam profile. When using the almost Gaussian beam profile 95/5, keyhole porosity occurs more frequently below a dimensionless build rate of 9. With the beam profiles 75/25, 65/35, and 50/50, keyhole porosity exclusively appears above a dimensionless enthalpy per dimensionless hatch of 6. No keyhole porosity is detected for the beam profiles 35/65, 20/80, and 10/90. Due to the dependence of the occurring keyhole porosity on the beam profile, the process window limit can neither be predicted by the model nor represented in the process map. Consequently, it can be concluded that ring-shaped beam profiles reduce keyhole porosity. Yuan et al. [105] utilized X-ray imaging to show that ring-shaped beam profiles can stabilize the keyhole by preventing collapses. Since keyhole collapses are the primary source of keyhole porosity, this stabilization can be assumed to be the origin of the reduced keyhole porosity in PBF-LB/M. Especially since the beam profiles 75/25, 65/35, and 50/50 are more likely to enter the keyhole melting mode than Gaussian beam profiles, according to the model from SM2. However, this keyhole is assumed to be stabilized by the ring-shaped beam profiles.

Concluded, SM3 demonstrates the feasibility of creating dimensionless process maps using dimensionless enthalpy, dimensionless hatch distance, and the Péclet number. Within these maps, the process window limits for lack of fusion and bulging top surfaces can be predicted based on the geometric properties of the modeled melt pools using a semi-analytical heat conduction model. The process window limit to keyhole porosity is influenced by the beam profile and, consequently, cannot be predicted by the model. Ring-shaped beam profiles typically result in lower keyhole porosity due to the stabilization of the keyhole by the ring-shaped beam profiles.

Key novelties of this solution module include the implementation of adjacent melt tracks in a semi-analytical heat conduction model, as well as the resulting prediction of process limits based on the geometric characteristics of the modeled melt pools. The prediction of bulging top surfaces represents a particularly significant innovation of this approach. Furthermore, the presentation format of the process maps is newly introduced. Previous approaches to deriving similar dimensionless process windows were either limited to single tracks [21, 126] or isolated a single influencing variable, such as the dimensionless hatch distance [129].

Furthermore, these approaches neglected the melt pool length and therefore made no statements about balling, humping, or bulging.

In conjunction with SM2, SM3 forms the basis for predicting the process window in PBF-LB/M with various beam profiles, utilizing the models presented and the new process map representation. The approach based on the modeled process signature enables, on the one hand, the acceleration of process parameter optimization in both science and industry. On the other hand, it provides the basis for comparing processes with different beam shapes and process parameters. Consequently, the following sections on SM4 and SM5 examine the influence of ring-shaped beam profiles for comparable processes, focusing on spatter formation and the resulting microstructure.

5.4 Spatter characteristics during laser-based powder bed fusion of metals with ring-shaped beam profiles

[I4] J. Grünwald, J. Reimann, K. Wudy, *Influence of ring-shaped beam profiles on spatter characteristics in laser-based powder bed fusion of metals. Journal of Laser Applications* 35 (2023) 042009. <https://doi.org/10.2351/7.0001153>.

This section contains the key results of publication [I4]. Detailed information is available in the reprint in the Appendix.

Solution modules SM1 to SM3 are focused on the mesoscopic process result. In order to answer research question RQ4 regarding the influence of ring-shaped intensity distributions on the process dynamics during PBF-LB/M, solution module SM4 investigates the impact on the process dynamics in the form of hot spatter formation. To ensure comparability between the examined processes, the dimensionless parameters used in solution modules SM2 and SM3 serve as a basis for comparison. In [I4], an alternative proportional equation for calculating the dimensionless enthalpy was used according to [20]. For consistency in this doctoral thesis, the graphs shown in this section were adjusted to use equation 2-12 for calculating the dimensionless enthalpy, deviating from [I4].

High-speed camera images of the process are captured and analyzed using image processing to investigate hot spatter formation. For this purpose, a high-speed camera (CP90-3-M-540, Optronis GmbH, Kehl, Germany) is integrated into the scanning system (see Section 4.2) and read out using a frame grabber (Coaxlink Quad G3, Euresys SA, Seraing, Belgium). In the experiments, images are acquired during the exposure of single tracks processed on an AISI 316L substrate with a 50 μm applied powder layer. The recorded raw data is evaluated according to the post-processing shown in Figure 31.

The acquired images initially show the overexposed process zone and the existing spatters (Step 1). The overexposed process zone is excluded in the next step (Step 2). For this purpose, a mask for the process zone is calculated for each track according to [140]. Based on binarization with an Otsu threshold [141], the existing spatters are detected in a subsequent step, and their projected size is determined (Step 3). Finally, the detected spatters from different frames are linked in a way that allows them to be combined into tracks, enabling the determination of spatter velocity using a Laplacian of Gaussian (LoG) filter in combination with a Kalman tracker (Step 4). This approach to spatter tracking has been introduced by Jäger et al. [142] in laser welding and by Slodczyk et al. [106] in PBF-LB/M, thus providing an established foundation.

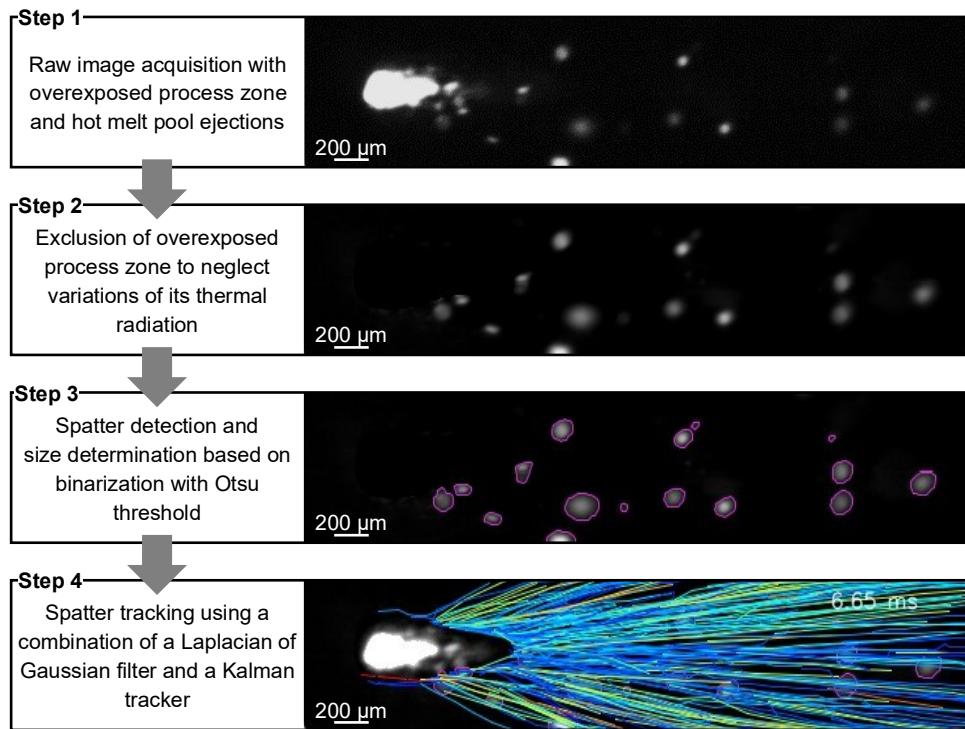


Figure 31: Schematic representation of the four steps of image processing for evaluating the characteristics of hot spatters in PBF-LB/M using on-axis imaging. Figure adapted from [14].

Figure 32a shows the resulting number of tracked spatters per fused area. The fused area is defined as the product of the melt track width measured at cross-sections and the scan vector length. The beam profile greatly influences the number of spatters per fused area. Processes using beam profiles with high peak-to-mean intensity ratios (especially beam profiles 75/25 and 65/35) show an increased number of detected spatters. Using beam profiles with low peak-to-mean aspect ratios (beam profiles 35/65, 20/80, and 10/90) results in a significantly lower number of spatters per fused area in similar ranges of the dimensionless enthalpy. The beam profiles with lower peak intensities result in lower peak

temperatures [143]. This presumably reduces the evaporation rate in the process zone, resulting in less powder entrainment effects, which can reduce the number of spatters [36].

In addition, Figure 32b shows the spatter velocity as a function of the dimensionless enthalpy for processes with different beam profiles. The main factor influencing the spatter velocity is the dimensionless enthalpy introduced. With higher dimensionless enthalpy, spatters become faster. This indicates an increased recoil pressure [66] resulting in high process dynamics.

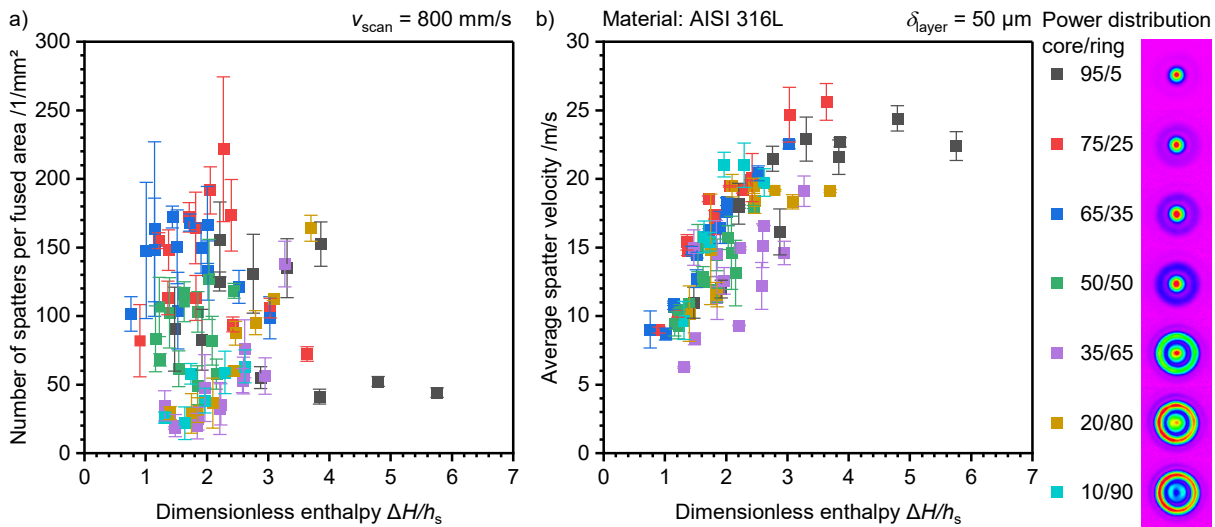


Figure 32: a) Number of spatters per fused area and b) average spatter velocity as a function of the dimensionless enthalpy during PBF-LB/M using ring-shaped beam profiles. Beam profiles with a high peak-to-mean intensity ratio (beam profiles 95/5, 75/25, 65/35, and 50/50) tend to generate more spatters than beam profiles with a low peak-to-mean intensity ratio (beam profiles 35/65, 20/80, and 10/90). The spatter velocity correlates with the dimensionless enthalpy introduced. Here, the beam profile is less decisive. Figure adapted from [14].

Figure 33 shows the spatter size as a function of the spot size used. Larger spots produce larger spatter. Spatters resulting from processing with beam profiles having high peak-to-mean intensity ratios (beam profiles 75/25, 65/35, and 50/50) tend to be larger than those produced by processes using beam profiles with lower peak-to-mean intensity ratios.

The hypothesis to explain the difference in spatter size is that the spatter origin changes due to the different peak-to-mean intensity ratios. Beam profiles characterized by less power in the center and more power in the ring (beam profiles 35/65, 20/80, and 10/90) result in a more stable melt pool because of the lower peak intensity. With a stable melt pool, processing results in a reduced amount of melt pool ejecta and mainly produces powder spatters [41]. The powder particles maintain their initial size when not agglomerating. Melt pool ejecta are typically larger than the powder spatter particles [38]. This results in smaller measured projected spatter sizes when using beam profiles with low peak-to-mean intensity profiles.

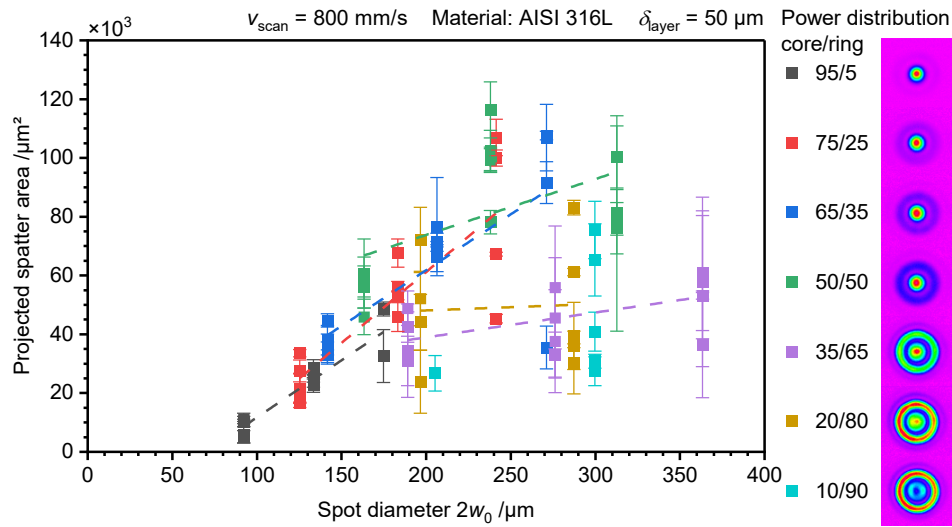


Figure 33: Projected mean spatter area as a function of the spot size of the beam profile used. Processes with larger beam profiles tend to produce larger spatter. Beam profiles with high peak-to-mean intensity ratios (beam profiles 75/25, 65/35, and 50/50) result in larger spatters than beam profiles with low peak-to-mean intensity ratios (beam profiles 35/65, 20/80, and 10/90). Figure adapted from [I4].

In summary, processes in which beam profiles with a high peak-to-mean intensity ratio (beam profiles 95/5, 75/25, 65/35, and 50/50) tend to produce more and larger spatters than processes using beam profiles with a low peak-to-mean intensity ratio (beam profiles 35/65, 20/80, and 10/90). The spatter velocity depends primarily on the dimensionless enthalpy introduced.

In conclusion, publication [I4] demonstrated that beam shaping can reduce spatter formation to smaller and fewer spatters, resulting in a more reliable process due to less dynamic process behavior. The evaluation was based on the dimensionless enthalpy, which was presented in SM2 and SM3 as a suitable comparison variable.

Publication [I4] was the first to use high-speed on-axis imaging for the quantitative evaluation of spatter quantity, size, and velocity during the use of ring-shaped beam profiles in PBF-LB/M. The results confirmed previous qualitative findings [101, 104] and characteristic values determined by off-axis imaging [108]. However, after the publication of [I4], a study was presented in which Kippels et al. [107] demonstrated a higher amount of spatter with smaller spatter particles when using ring-shaped beam profiles detected with an off-axis camera setup. The publication [107] does not propose any hypotheses regarding possible causes for the higher spatter amount, while observing the smaller spatter particles for ring-shaped beam profiles aligns with [I4].

5.5 Tailored microstructures in laser-based powder bed fusion of metals using ring-shaped beam profiles

[I5] J. Grünewald, R. Pfund, D. Röhrer, L. Hitzler, J. Torgersen, K. Wudy, *Tailored microstructures in laser-based powder bed fusion of 316L using beam shaping: From microstructure formation in individual specimens to their transition in segmented parts*, *Additive Manufacturing* (2025) 105026. <https://doi.org/10.1016/j.addma.2025.105026>.

This section presents the key results of [I5]. For detailed information, please refer to the reproduction of the publication in the Appendix.

Solution module SM5 investigates the ability to modify the as-built microstructure of AISI 316L in PBF-LB/M using ring-shaped beam profiles. In a first step, the influence of different beam profiles on the resulting microstructure in a larger volume is investigated. The feasibility of locally tailoring the microstructure within one component using different beam profiles is examined in a second step. As in SM4, the investigations utilize the dimensionless parameters derived from SM2 and SM3 for comparison between processes with different beam shapes. The study answers research question RQ5, which concerns the influence of beam profiles used on the as-built microstructure of PBF-LB/M-manufactured parts.

For the microstructure investigations in a larger volume, specimens manufactured with constant comparable dimensionless parameters (according to solution module SM3) are examined (“single-parameter specimens”, see Figure 34a). To investigate the feasibility of locally tailoring microstructure, two comparable parameter sets with different beam profiles and distinct microstructures are combined to manufacture different segments within a single component (“segmented specimens”, see Figure 34b). The resulting microstructure of these specimens is analyzed using electron backscatter diffraction (EBSD) measurements, evaluating the crystallographic texture and grain size. Additionally, the size of the microstructural transition zone is evaluated for the segmented specimens.

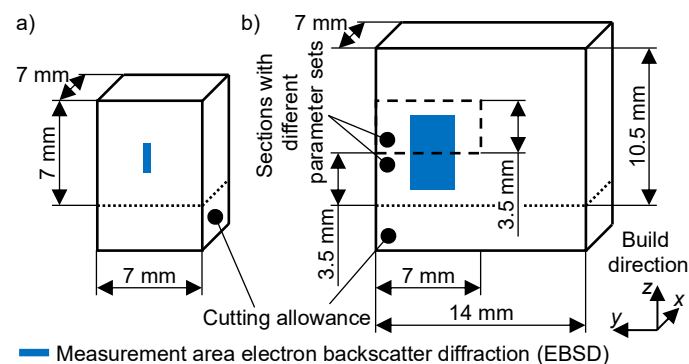


Figure 34: Drawings of the sample geometries used: a) single-parameter specimens for the investigations in a larger volume, and b) segmented specimens for the investigations of local microstructure tailoring. Figure adapted from [I5].

For this transition zone evaluation, the grains are reconstructed from the EBSD data in sections of varying sizes with different distances from the plane, where the beam profile was changed. The grain size within these sections is calculated for each grain, and it is determined whether the respective grain is intersected by the beam profile switching plane. The methodology is shown schematically in Figure 35. From the calculated grain sizes, the area fraction of the grains intersected by the plane of the beam profile change $A_{\text{intersected},\%}$ is determined by relating the area of the intersected grains $A_{\text{intersected}}$ to the sum of the areas of the newly emerged grains A_{new} and the intersected grains $A_{\text{intersected}}$ corresponding to:

$$A_{\text{intersected},\%} = \frac{A_{\text{intersected}}}{A_{\text{new}} + A_{\text{intersected}}} \quad 5-23$$

The height of the transition zone is defined for an area fraction $A_{\text{intersected},\%}$ of 50 %.

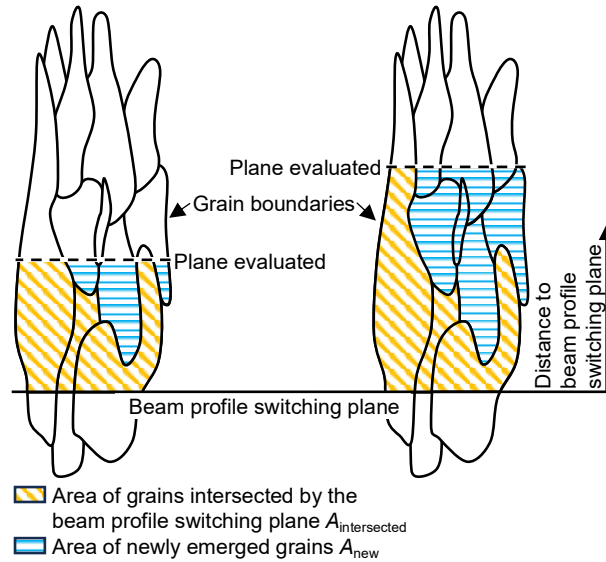


Figure 35: Schematic representation of the procedure for evaluating the size of the transition zone of segments manufactured with different beam profiles. Figure adapted from [15].

Figure 36 shows excerpts from the inverse pole figure (IPF) maps and the corresponding IPFs parallel to the build direction (z-direction) of the single-parameter specimens, which have a comparable dimensionless enthalpy and a dimensionless hatch h^* in the range from 1.3 to 1.4. The color code of the IPF maps represents the orientation in the z-direction. Additionally, Figure 38a shows the fiber volume fractions of the high-symmetry directions $\langle 100 \rangle$, $\langle 110 \rangle$, and $\langle 111 \rangle$ in the build direction with a tolerance angle of 10° . The preferred crystallographic orientation differs depending on the beam profile and parameter set used. The specimen manufactured with beam profile 95/5 exhibits a firm $\langle 100 \rangle \parallel \text{BD}$ texture with a fiber volume fraction of approximately 20 %, as expected from the literature [50]. There is no preferred orientation when using beam profiles 65/35 and 50/50. All fiber volume fractions in high-symmetry directions are below 10 %. When using beam profile 35/65, the sample

exhibits two preferred orientations. The majority of the grains have a preferred $\langle 100 \rangle \parallel$ BD orientation with a volume fraction of almost 25 %, while a minor fraction of approximately 15 % is oriented in $\langle 111 \rangle \parallel$ BD direction. When using the beam profiles 20/80 and 10/90, a preferred direction forms in $\langle 111 \rangle \parallel$ BD with a volume fraction of approximately 25 %.

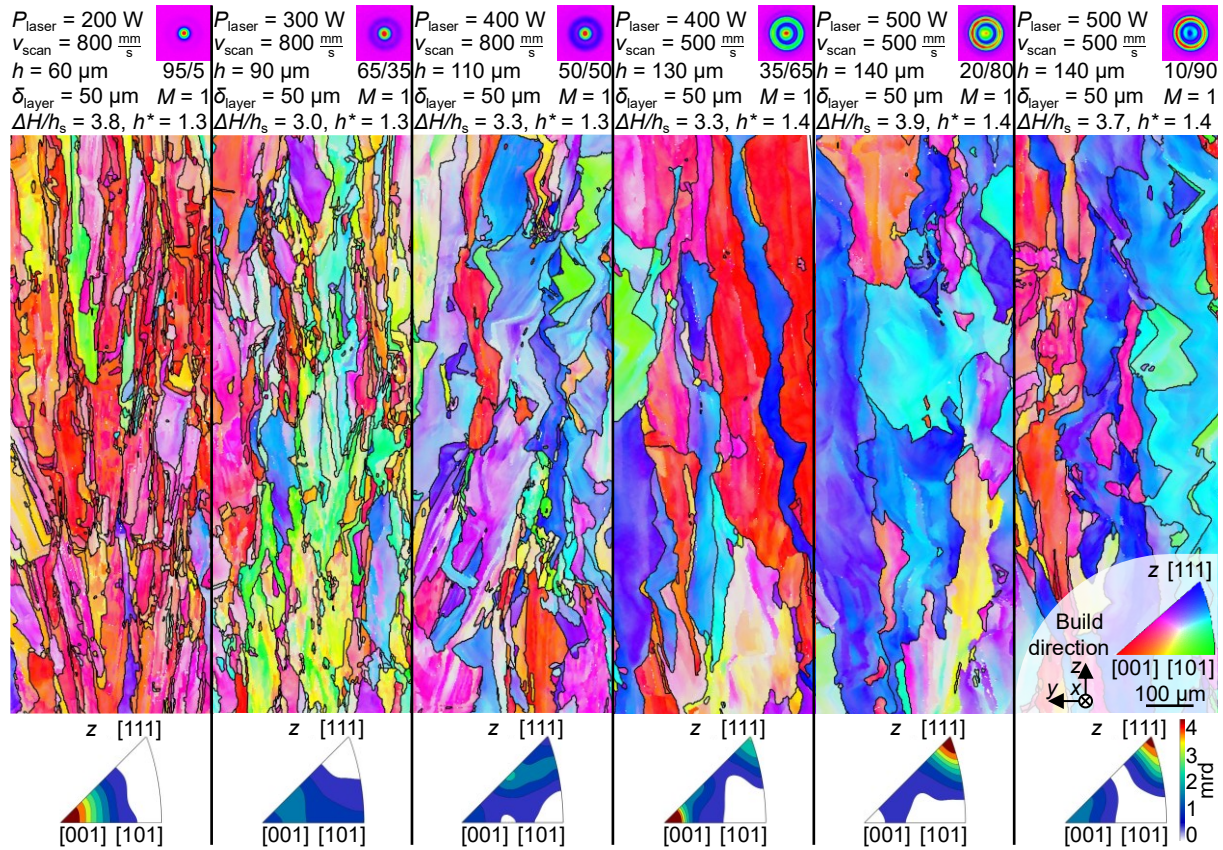


Figure 36: Excerpts of the IPF maps and IPFs of specimens manufactured using different beam profiles, a comparable dimensionless enthalpy, and a dimensionless hatch between 1.3 and 1.4. The specimen manufactured using beam profile 95/5 exhibits a firm $\langle 100 \rangle \parallel$ BD texture. Beam profiles 65/35 and 50/50 result in no preferred crystallographic orientation. Using beam profile 35/65 results in the two preferred orientations $\langle 100 \rangle$ and $\langle 111 \rangle$ parallel to the z-direction. A firm $\langle 111 \rangle \parallel$ BD texture is formed when beam profiles 20/80 and 10/90 are used. Figure adapted from [15].

The different textures observed can be attributed to the shape of the remaining parts of the melt tracks. As shown in Figure 37a, when using the almost Gaussian beam profile 95/5, the melt pool cross-sections have steep side walls almost parallel to the build direction. This enables the grains to grow along their preferred $\langle 100 \rangle$ growth direction parallel and perpendicular to the build direction, both from the sides and the bottom of the melt pool. As a result, a preferred $\langle 100 \rangle \parallel$ BD orientation is created. When using the ring-shaped beam profile 10/90, the remaining part of the melt track has a different shape. The steepest side angle is approximately 45° (see Figure 37b), which would result in a preferred $\langle 110 \rangle$ or $\langle 111 \rangle$ direction relative to the build direction. It is assumed that the lateral solidification fronts in the longitudinal middle section of the melt pool (“transition section” according to [13]) have rotated to the x-z and y-z planes in such a way that the $\langle 111 \rangle$ direction of the solidifying

material is parallel to the build direction. This three-dimensional melt pool shape aligns with the hypothesis of Pérez-Ruiz et al. [121]. Moreover, a preliminary study [144] using a high-fidelity multi-physics simulation for similar process parameters proved the emergence of this melt pool shape. Furthermore, using a layer rotation angle φ_{rot} of 67° in the experiments could favor the $\langle 111 \rangle$ direction, as it is close to the symmetry angle of the $\langle 111 \rangle$ orientation, which is 60° .

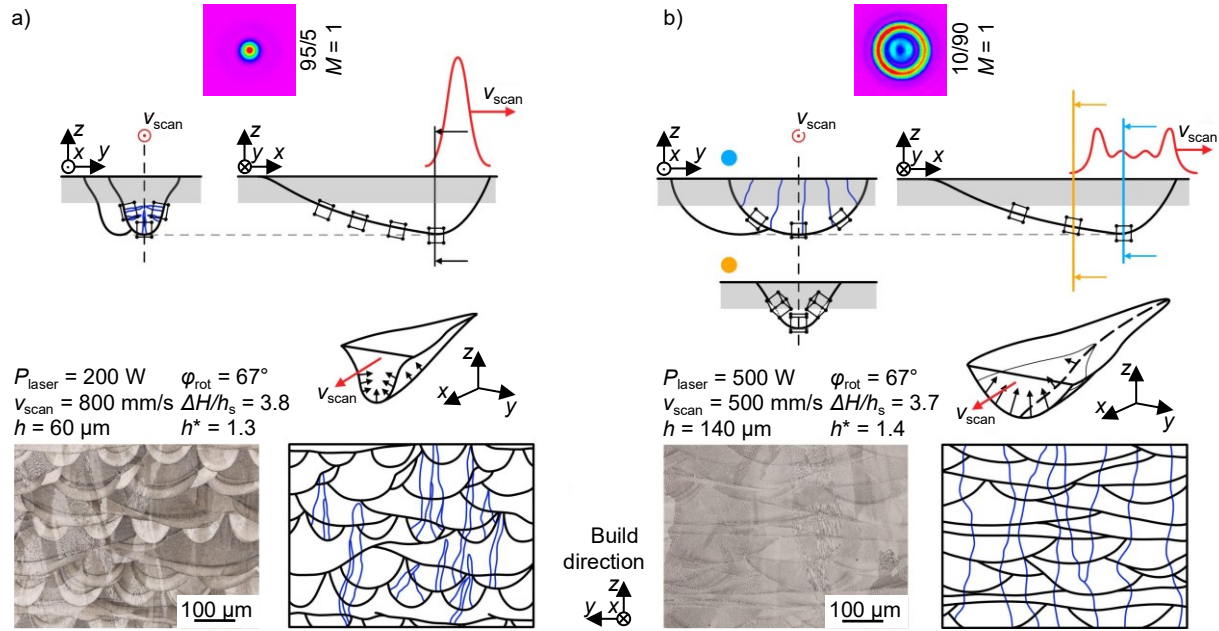


Figure 37: Illustration of the texture forming mechanism for a) beam profile 95/5 and b) beam profile 10/90. The use of beam profile 95/5 yields melt track cross-sections with steep side walls, enabling grains to grow from the bottom of the melt pool and the side walls with the preferred $\langle 100 \rangle$ orientation. Beam profile 10/90 produces melt-track cross-sections with side walls with a maximum angle of approximately 45° to the build direction, resulting in a rotated preferred growth direction. The schematic melt pool representations are adopted from high-fidelity multi-physics simulations carried out in [144]. Figure adapted from [15].

Additionally, Figure 38b shows the influence of the beam profiles on the area-weighted equivalent grain radius $r_{50,2}$. The beam profiles are sorted by size. As the beam profiles become larger, the resulting microstructure tends to become coarser.

A general explanation for a coarser microstructure is the reduction in the average cooling rate [42]. In the present case, the cooling rate is assumed to be reduced due to the use of larger beam profiles. Since the samples were manufactured with similar dimensionless parameters, a reduced temperature gradient can be assumed for the larger beam profiles. As derived and discussed in detail in solution modules SM2 and SM3, similar dimensionless parameters result in a similar dimensionless temperature field. The corresponding coordinates of this field are normalized to the spot size (x^* and y^* coordinates) and the thermal diffusion depth (z^* coordinate). This means that the dimensioned temperature field is spatially expanded for larger beam sizes. Consequently, the temperature gradients are reduced for larger beam profiles.

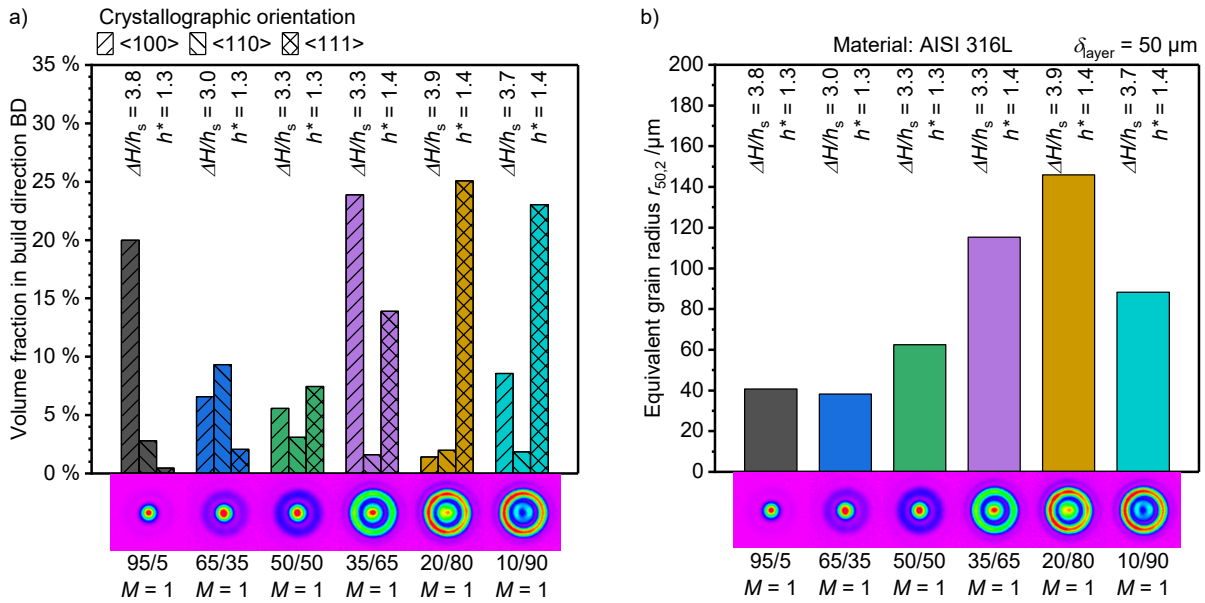


Figure 38: a) Fiber volume fractions of the crystallographic high-symmetry directions <100>, <110>, and <111> in the build direction with a tolerance angle of 10° and b) equivalent grain radius $r_{50,2}$. Depending on the beam profile used, the volume fractions in the high-symmetry direction change. The composition varies from a high volume fraction of 20 % in <100> || BD using beam profile 95/5 to a <111> || BD texture with a volume fraction of approximately 25 % when beam profiles 20/80 and 10/90 are used. The average equivalent grain radius depends primarily on the spot size. Larger beam shapes tend to produce larger grains. Figure adapted from [I5].

Due to the considerable differences in texture and grain size, the parameter sets of the beam profiles 95/5 and 10/90 are examined for the segmented samples. A parameter set with an enlarged beam profile 95/5 is additionally examined to isolate the influences of beam profile and size. Figure 39 shows cross-sections of the combined samples etched with V2A etchant.

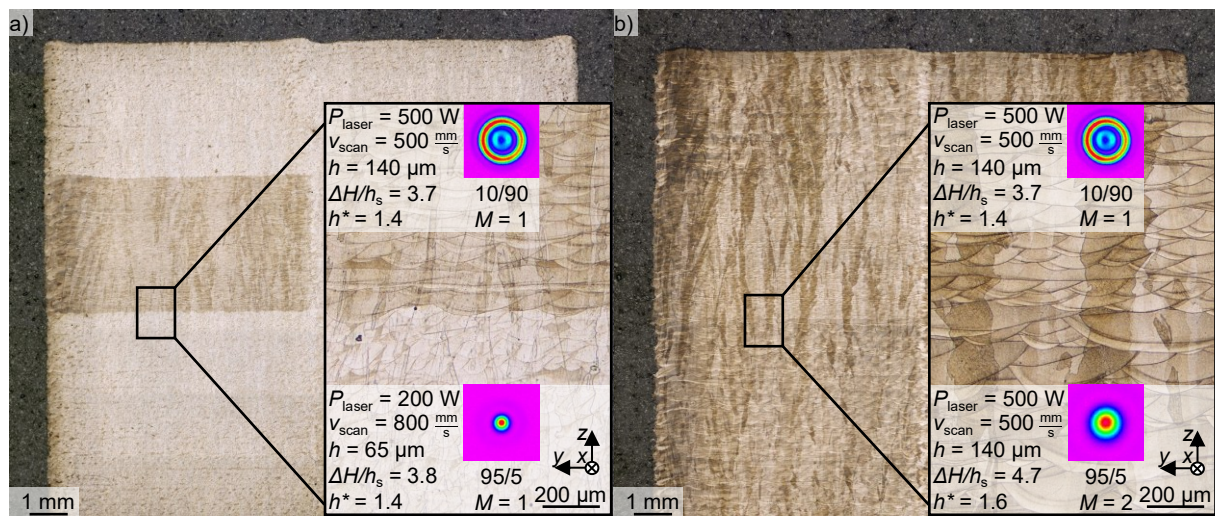


Figure 39: Etched cross-sections of combined specimens using a) small beam profile 95/5 and beam profile 10/90, and b) an enlarged beam profile 95/5 and beam profile 10/90. When using the smaller beam profile 95/5 and beam profile 10/90, a high contrast between the segments is apparent. When using similarly sized beam profiles (enlarged beam profile 95/5 and similarly sized beam profile 10/90), the contrast between the segments is considerably lower. Figure adapted from [I5].

When using the smaller, almost Gaussian beam profile 95/5 in combination with the ring-shaped beam profile 10/90, a clear difference between the various segments is visually apparent (see Figure 39a). The close-up section shows the expected differences in the melt pool cross-sectional shapes. When parameter sets with the larger, almost Gaussian beam profile 95/5 and a similarly large ring-shaped beam profile 10/90 are combined, the contrast between the segments is lower (see Figure 39b). However, differences in the cross-sectional shapes of the individual melt tracks are visible in the close-up section. In addition, it can be observed that large grains grow beyond the beam profile switching plane.

Figure 40 shows the IPF maps of the EBSD measurements in the transition zone between two segments of the segmented samples. In addition, the IPFs of the individual segments and the curve of the area fraction $A_{\text{intersected},\%}$ along the distance to the beam profile switching plane are shown. Furthermore, Figure 41 shows the fiber volume fractions of the high-symmetry directions $\langle 100 \rangle$, $\langle 110 \rangle$, and $\langle 111 \rangle$ in the build direction with a tolerance angle of 10° (see Figure 41a) and the quantitative evaluations of the grain sizes (see Figure 41b) of all segments examined.

Figure 40a shows the data from the segmented sample combining the ring-shaped beam profile 10/90 with the small, almost Gaussian beam profile 95/5. In the IPF map, the beam profile switching level is marked by a continuous white line. A clear difference in grain size is visible between the two segments. The grains in the initially manufactured layers using the small ($M = 1$) beam profile 95/5 are significantly smaller than those in the subsequently manufactured layers using beam profile 10/90 (see Figure 40a and Figure 41b). The height of the transition area is approximately $175 \mu\text{m}$, corresponding to 3.5 layer thicknesses. Both segments have a weak $\langle 100 \rangle \parallel \text{BD}$ texture with volume fractions slightly above 10 % respectively (see Figure 40a and Figure 41a).

Figure 40b shows the data from the segmented sample, which combines the ring-shaped beam profile 10/90 with the enlarged, almost Gaussian beam profile 95/5. The grains of the initially manufactured layer with the expanded ($M = 2$) beam profile 95/5 are slightly smaller than the grain size of the layers subsequently manufactured with beam profile 10/90 (see Figure 41b). The grains are significantly larger than for the segmented sample in Figure 40a across the entire measurement range. This is reflected in the elongated curve of the area fraction $A_{\text{intersected},\%}$. With a height of $1475 \mu\text{m}$ (29.5 layers), the transition area is significantly larger than in Figure 40a. Another contrast to the segmented sample in Figure 40a is the texture change between the two segments. The initially manufactured layers with the expanded ($M = 2$) beam profile 95/5 exhibit two moderately pronounced preferred directions with $\langle 100 \rangle \parallel \text{BD}$ and $\langle 111 \rangle \parallel \text{BD}$ orientations, each with volume fractions of slightly above

10 % (see Figure 41a). In the layers subsequently manufactured with beam profile 10/90, the texture rotates towards $\langle 111 \rangle \parallel \text{BD}$ with a volume fraction of 20 % (see Figure 41a).

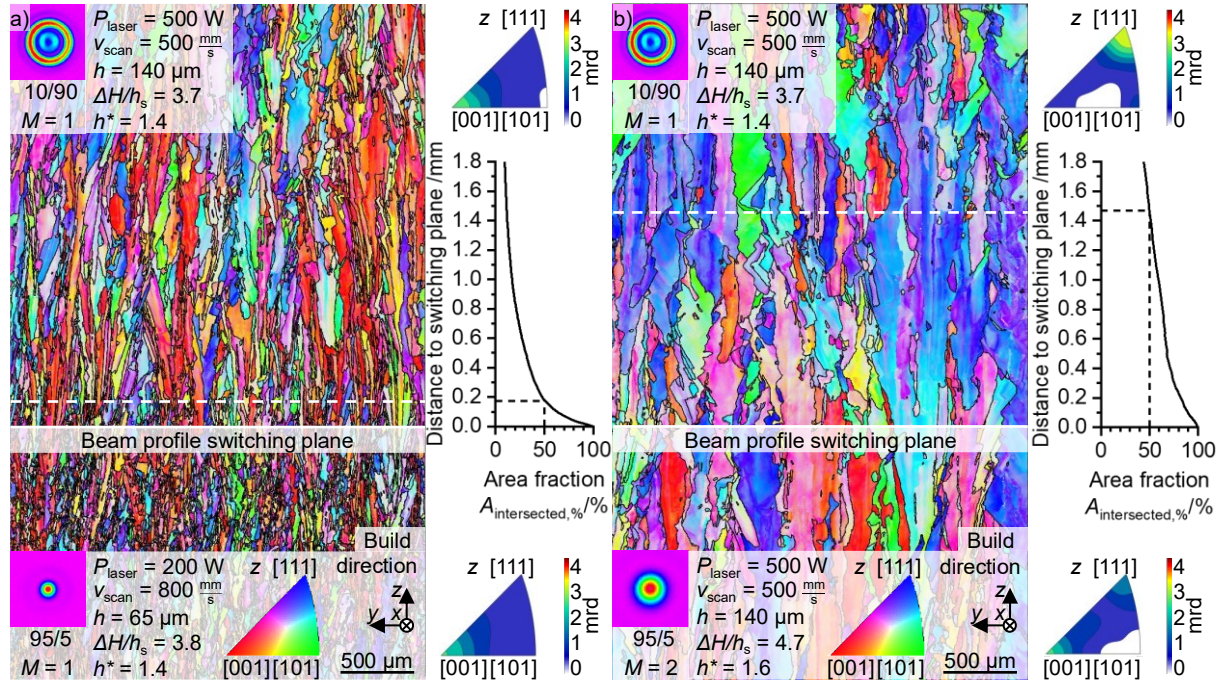


Figure 40: IPF maps of the EBSD measurements in the transition zone between two segments, IPFs of the individual segments, and the progression of the area fraction $A_{\text{intersected, \%}}$ of the segmented samples that combine a) beam profile 10/90 with a small, almost Gaussian beam profile 95/5 and b) beam profile 10/90 with a widened, almost Gaussian beam profile 95/5. Both samples differ significantly. The combination with the underlying small, almost Gaussian beam profile 95/5 has a small transition zone of approximately 175 μm . The combination based on the widened, almost Gaussian beam profile 95/5 has a significantly larger transition zone of approximately 1475 μm . Figure adapted from [15].

A direct comparison of the specimen and segments, which were produced under different boundary conditions with beam profiles 10/90 and the corresponding parameter set ($P_{\text{laser}} = 500 \text{ W}$, $v_{\text{scan}} = 500 \text{ mm/s}$, $h = 140 \mu\text{m}$), reveals that the specimen and segments exhibit significantly different microstructures. Depending on the underlying microstructure, different grain sizes and texture compositions are formed (see Figure 41). The finest grain structure is present in the segments built on the segment produced with the small beam profile 95/5. This is followed by the specimen built directly on the build platform. The largest grains are formed in the sample built on a segment manufactured with the large beam profile 95/5 ($M = 2$). Regarding preferred crystal orientations, the sample built on the “neutral” build platform exhibits the strongest $\langle 111 \rangle \parallel \text{BD}$ preference, with a volume fraction of approximately 25 %. The segmented sample, including the coarse microstructure generated with the enlarged beam profile 95/5 ($M = 2$), comes closest to this texture. This segment exhibits a volume fraction of 20 % in $\langle 111 \rangle \parallel \text{BD}$ direction. The preferred $\langle 111 \rangle \parallel \text{BD}$ orientation could not be reproduced when building on the fine-grained microstructure, manufactured with the small beam profile 95/5 ($M = 1$). Here, the corresponding parameter

set yields a slightly preferred $\langle 100 \rangle \parallel$ BD orientation with a volume fraction of approximately 10 %.

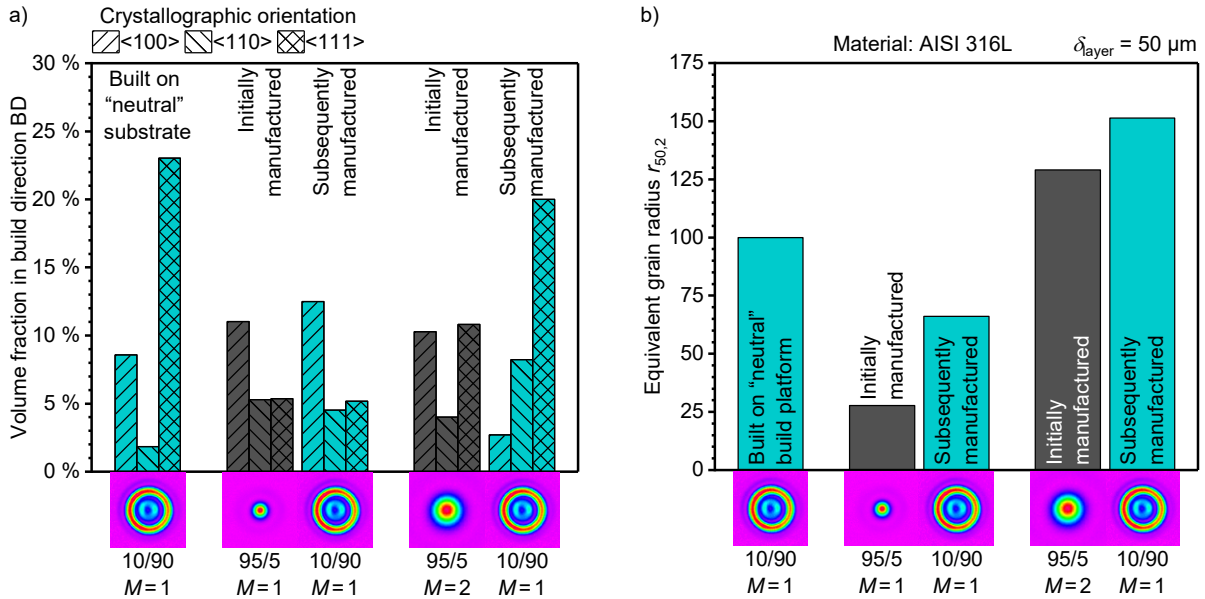


Figure 41: Quantitative comparison of the microstructure of samples manufactured using the ring-shaped beam profile 10/90 under different boundary conditions with regard to a) the resulting textures in the form of the volume fraction in the high-symmetry directions $\langle 100 \rangle$, $\langle 110 \rangle$, and $\langle 111 \rangle$ in the building direction, data from [I5], and b) the resulting equivalent grain radius $r_{50,2}$, adapted from [I5]. Both the resulting texture and the grain sizes differ in the three samples and depend on the microstructure of the initially manufactured layers or substrate.

By using beam profile 10/90 with identical process parameters in the observed sections, the solidification conditions are similar. Consequently, the different microstructures are assumed to form due to epitaxial grain growth, discussed in [145]. Hence, the various grain sizes could emerge due to the non-necessity of nucleation in AISI 316L [10], which allows grains present at the solidification front in initial layers to further grow. This results in a grain size in the subsequently produced layers that corresponds to or at least scales with the grain size of the initial layers.

Additionally, the texture in the initially manufactured layers influences the microstructure in the subsequently produced layers. The parameter set using beam profile 10/90 in the single-parameter cuboid sample also has a volume fraction of approximately 10 % in $\langle 100 \rangle \parallel$ BD orientation. Therefore, the continuous growth of the grains within the sample manufactured with the small Gaussian beam profile in $\langle 100 \rangle \parallel$ BD could be energetically more favorable. The ability of the beam profile 10/90 to produce $\langle 100 \rangle \parallel$ BD-oriented and $\langle 111 \rangle \parallel$ BD-oriented grains may extend the transition zone of the crystallographic texture.

In conclusion, the results indicate that the initial grain size and crystallographic orientation of the initially manufactured layers or the substrate have a significant influence on the final texture and grain size of the subsequently manufactured segment. This means that when

locally adjusting microstructures, the microstructure of previously manufactured layers must be considered, as it significantly influences the result. However, the results demonstrate the possibility of locally tailoring microstructures and, thus, potentially adjusting mechanical properties by utilizing different beam profiles with corresponding parameter sets for manufacturing different segments within a component. Exploiting this potential enables PBF-LB/M-manufactured components to possess an additional degree of design freedom that cannot be achieved in any other manufacturing process.

At the time of publication of [15], there were a limited number of studies on the influence of ring-shaped beam profiles on the resulting microstructure during the processing of IN718. Since IN718 is an fcc material like AISI 316L, the findings are transferable to a certain extent. Compared to the prior state of the art, [15] contains two key novelties. On the one hand, previous beam-shaping studies [110, 121, 122] exclusively considered the influence of the beam profile in test specimens manufactured using a single parameter set. The transition zone between different microstructures, and thus the potential for microstructure tailoring, had not been evaluated before. On the other hand, in contrast to the existing literature, [15] used the dimensionless approach from SM2 and SM3 as a comparison variable. Cozzolino et al. [110] use the volume energy density as a comparison variable between processes with different beam profiles. According to Bakhtari et al. [58], the volume energy density is limited as a comparison parameter between processes with different beam profiles. Mirzabeigi et al. [122] and Pérez-Ruiz et al. [121] used individual dimensioned process parameters, such as laser power, scanning speed, and hatch distance, as comparison parameters. Consequently, [15] employs a new comparison basis and offers fundamental insights into how existing knowledge of microstructure modification through beam shaping can be utilized to tailor the microstructure at the part level, with corresponding transition areas.

6 Critical discussion and conclusion

Despite the expanded manufacturing possibilities, the industrial implementation of the PBF-LB/M process faces certain challenges. These include long throughput times in production, small process windows, and the limited use of potential process freedoms, such as microstructure tailoring, partly attributed to the common use of Gaussian beam profiles in PBF-LB/M. Driven by these challenges, this doctoral thesis demonstrates the fundamental potential and influence of applying ring-shaped beam profiles in PBF-LB/M. The overall aim of the work is to gain *a fundamental understanding of the relationship between beam profile characteristics, a modeled process signature, and the dynamic process behavior, as well as the resulting component properties*. To achieve this, a hierarchical approach consisting of five solution modules (SM) is pursued:

SM1 Investigation of the basic potential of ring-shaped beam profiles to influence the melt track dimensions and top surface topology.

SM2 Modeling the melt pool dimensions and the melting modes when using different beam profiles in various sizes. An important side branch for this is the Identification of an adequate characterization metric for ring-shaped beam profiles.

SM3 Creation of a beam size and profile independent process map representation with predicted process limits.

SM4 Demonstration of the impact of ring-shaped beam profiles on the process dynamics.

SM5 Modification of the as-built microstructure using various beam profiles.

The contents of the solution modules are based on peer-reviewed publications and consolidated in the “Research findings” section. The respective publications are reprinted in the Appendix.

In solution module SM1, the basic influence of ring-shaped beam profiles on the melt track dimensions (width and depth) and the top surface topology in PBF-LB/M is examined. From the collected data, the size of the process windows is deduced. Ring-shaped beam profiles result in shallower and wider melt tracks. As a result, laser power can be increased without the process entering the keyhole melting mode, avoided with PBF-LB/M. Moreover, fewer balling effects are observed when using ring-shaped beam profiles. Consequently, the ring-shaped beam profiles enlarge the process window. The accompanying paper [I1] examined for the first time the potential of using different beam shapes consisting of a Gaussian core and a surrounding ring for PBF-LB/M. It also investigated for the first time the use of a corresponding multi-core fiber laser for PBF-LB/M. Subsequently, the key findings of the paper were confirmed by other research groups. Additionally, the findings can be applied to

other related processes, such as heat conduction welding. However, the paper employed a black box approach to the investigation, observing input-output correlations, as is commonly used in the state of the art, which limits the transferability of the findings to other system technologies, spot size, or materials.

To establish a basis for comparing processes with different beam shapes, solution module SM2 proposes a method for modeling melt track widths and depths on a single-track basis using a semi-analytical heat conduction model. The model is based on the parameters dimensionless enthalpy and Péclet number. These parameters include thermophysical properties of the processed material and the spot radius of the beam profile, thereby scaling the modeled temperature field accordingly. The beam shape is assumed to be Gaussian in the model to maintain its semi-analytical character. A prerequisite for reducing model errors when using ring-shaped beam profiles in experiments is the use of the second moment method to determine the spot size. Additionally, the model is extended to include the peak-to-mean intensity ratio. This way, it enables the melting mode to be estimated. Beam profiles with a high peak-to-mean intensity ratio exhibit a higher tendency to result in keyholing than those with a low peak-to-mean intensity ratio. Thus, the corresponding publication [12] demonstrated new semi-analytical approaches for estimating the resulting melt pool dimensions and melting modes when using ring-shaped beam profiles. The application of these approaches provided a crucial foundation for comparing processes with different beam shapes. However, at this stage, the model is limited to single tracks.

Extending the dimensionless heat conduction model to include the dimensionless hatch distance and scan vector length enables the simulation of a hatched process within solution module SM3. A prediction of the process limits, in terms of bulging top surfaces and lack of fusion porosity, is enabled by evaluating the geometric characteristics of the modeled melt pools. Modeled melt pools with a length-to-width ratio above 2.1 exhibit bulging top surfaces in the experiments. When the modeled minimum melt track depth of a layer is below 5 μm , lack of fusion porosity is present in the experimental specimen. Furthermore, using the parameters specified in the model (dimensionless enthalpy, Péclet number, and dimensionless hatch distance), new dimensionless process maps can be created. These maps facilitate the comparison of processes with different beam profiles across various beam sizes. This approach of [13] has the potential to significantly reduce the experimental effort required for investigating the PBF-LB/M process, since the models and comparative parameters applied are freely parameterizable and include beam and material characteristics. However, in the corresponding publications, both approaches of SM2 and SM3 have been validated exclusively for ring-shaped beam profiles and the material AISI

316L with powder layer thicknesses of 50 μm . The transfer to other beam profiles, materials, or powder layer thicknesses is the subject of future work. Nevertheless, in combination, solution modules SM2 and SM3 provide dimensionless parameters, enabling the comparison of processes with different beam shapes. These provide the basis for exploring the influence of ring-shaped beam profiles on process dynamics and the resulting component properties in solution modules SM4 and SM5.

Solution module SM4 examines the influence of ring-shaped beam profiles on process dynamics by investigating hot spatter formation. With comparable dimensionless parameters according to SM2 and SM3, the beam profile affects the amount and size of spatters occurring during the process. Processes in which beam profiles with a high peak-to-mean intensity ratio tend to produce more and larger spatters than processes using beam profiles with a low peak-to-mean intensity ratio. The spatter velocity depends primarily on the dimensionless enthalpy introduced. The corresponding paper [14] presented for the first time data from high-speed on-axis imaging for the quantitative evaluation of spatter formation during PBF-LB/M using ring-shaped beam profiles. Thereby, findings from other investigation methods regarding the influence of ring-shaped beam profiles on spatter formation during the processing of different materials were confirmed. The use of on-axis imaging in SM4 enabled the generation of high-resolution images with a high frame rate. However, this approach entails that only the top view of the process zone is imaged. As a result, only the projected size and velocity of the spatter could be evaluated. This means that information about the spatter height and, accordingly, the three-dimensional spatter velocity was neglected. The use of an off-axis stereo camera setup [146] could provide additional valuable three-dimensional information in future investigations.

Finally, solution module SM5 shows the influence of ring-shaped beam profiles on the resulting microstructure, representing the component properties. As in SM4, the dimensionless parameters from SM2 and SM3 were used as comparison variables. At comparable parameters, the application of ring-shaped beam profiles alters the resulting texture of the microstructure. When using state-of-the-art Gaussian beam profiles, a pronounced $\langle 100 \rangle \parallel \text{BD}$ texture can be created. Ring-shaped beam profiles allow the preferred direction to be rotated towards a $\langle 111 \rangle \parallel \text{BD}$ texture. Moreover, the resulting grain size scales with the spot size of the beam profiles used. The microstructural differences of various beam profiles are used to investigate the potential for locally tailoring microstructure. In this application, the results reveal a strong influence of the microstructure of the substrate or the initially manufactured layers on the emerging microstructure. Among other effects, larger grains in the initially produced layers lead to a larger transition zone. This means that

the microstructure of previously manufactured layers must be considered, as it significantly influences the result. On the one hand, the study of [15] confirmed observations previously made in studies conducted on the IN718 material. On the other hand, the work included a detailed investigation of the transition zone in local microstructure adaptation, which represented a significant degree of novelty. This demonstrated for the first time the influence of the microstructure present in initially manufactured layers on the emerging microstructure during microstructure tailoring by beam shaping. However, the study was conducted on simple specimens. Therefore, the implementation of the findings in the manufacturing process of a real-world application component with customized microstructure sections would be an exciting prospect for future investigations. Furthermore, the microstructure formation was examined as representative of the component properties. The influence of the microstructure on mechanical component properties, such as static or dynamic strength and Young's modulus, remains to be investigated in the future.

In summary, this doctoral thesis offers a new understanding of the relationship between the characteristics of ring-shaped beam profiles in the PBF-LB/M process and the resulting component properties, i.e., the microstructure. To enable a consistent linkage between the processes with different beam profiles, a semi-analytical approach is proposed to provide a transferable process signature. The insights gained contribute significantly to the highly topical research area of applying beam shaping techniques in laser material processing.

The specific impact of the overall findings might be improved process transferability between different PBF-LB/M systems when using beam shaping through the proposed model approach. Moreover, the findings can be used to adapt processes through beam shaping, which can lead to reduced spatter or a modified microstructure. This, in turn, opens the potential to locally tailor the microstructure of different component areas according to their macroscopic load. The findings support the further industrialization of additive manufacturing and demonstrate an easy-to-implement solution for exploiting previously untapped potential, such as local microstructure tailoring.

7 Publications of the author

As part of the research work during the doctoral period, various approaches of spatial beam shaping [I1–I5, J1–J3, J6–J8, J10, C2–C4], temporal beam shaping (such as power modulation [J4, J5]) and combination of these approaches (such as very rapid redistribution of the laser power [C1, P1–P15]) were investigated. This thesis focuses on the use of ring-shaped beam profiles created with spatial beam shaping and their influence on the resulting process and component properties, utilizing the exemplary material AISI 316L. Consequently, the marked references [I1–I5] are included in this doctoral thesis. All publications included were primarily written by the first author and published as open-access articles under the CC BY license. The included publications are reprinted in the Appendix.

7.1 Integrated peer-reviewed journal publications

- [I1] J. Grünewald, F. Gehringer, M. Schmöller, K. Wudy, Influence of Ring-Shaped Beam Profiles on Process Stability and Productivity in Laser-Based Powder Bed Fusion of AISI 316L, *Metals* 11 (2021) 1989. <https://doi.org/10.3390/met11121989>.
- [I2] J. Grünewald, M. Wittemer, K. Wudy, Scaling laws for ring-shaped beam profiles in laser-based powder bed fusion of metals, *Advances in Industrial and Manufacturing Engineering* 10 (2025) 100164. <https://doi.org/10.1016/j.aime.2025.100164>.
- [I3] J. Grünewald, K. Wudy, Dimensionless process windows in laser-based powder bed fusion of AISI 316L using ring-shaped beam profiles, *Additive Manufacturing Letters* 14 (2025) 100284. <https://doi.org/10.1016/j.addlet.2025.100284>.
- [I4] J. Grünewald, J. Reimann, K. Wudy, Influence of ring-shaped beam profiles on spatter characteristics in laser-based powder bed fusion of metals. *Journal of Laser Applications* 35 (2023) 042009. <https://doi.org/10.2351/7.0001153>.
- [I5] J. Grünewald, R. Pfund, D. Röhrer, L. Hitzler, J. Torgersen, K. Wudy, Tailored microstructures in laser-based powder bed fusion of 316L using beam shaping: From microstructure formation in individual specimens to their transition in segmented parts, *Additive Manufacturing* (2025) 105026. <https://doi.org/10.1016/j.addma.2025.105026>.

7.2 Additional peer-reviewed journal publications

- [J1] B. Burchard, J. Hofmann, J. Grünewald, K. Wudy, Increasing productivity in laser-based powder bed fusion of plastics using defocused Gaussian beams, *Rapid Prototyping Journal* 31 (2025) 247–256. <https://doi.org/10.1108/RPJ-01-2025-0020>.

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- [J3] M. Godec, N. Mirzabeigi, Č. Donik, D.A. Skobir Balantič, S. Malej, I. Paulin, B. Podgornik, J. Grünewald, T. Novotny, F. Bayer, F. Ruiz-Zepeda, K. Wudy, Influence of laser beam shape and post heat-treatment on the microstructure and mechanical properties of nickel-based alloy IN718 manufactured by laser powder bed fusion, *Journal of Materials Research and Technology* 37 (2025) 1970–1986. <https://doi.org/10.1016/j.jmrt.2025.06.119>.
- [J4] J. Grünewald, P. Clarkson, R. Salvesson, G. Fey, K. Wudy, Influence of Pulsed Exposure Strategies on Overhang Structures in Powder Bed Fusion of Ti6Al4V Using Laser Beam, *Metals* 11 (2021) 1125. <https://doi.org/10.3390/met11071125>.
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- [J6] V. Holla, J. Grünewald, P. Kopp, P.M. Praegla, C. Meier, K. Wudy, S. Kollmannsberger, Validity of Thermal Simulation Models for Different Laser Beam Shapes in Bead-on-Plate Melting, *Integrating Materials and Manufacturing Innovation* 13 (2024) 969–985. <https://doi.org/10.1007/s40192-024-00382-2>.
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- [J10] S. Yan, R. Off, A.B. Yayak, K. Wudy, A. Aghajani-Talesh, M. Birg, J. Grünewald, M. Holenderski, N. Meratnia, Deep learning based phase retrieval with complex beam shapes for beam shape correction, *Optics Express* 33 (2025) 10806. <https://doi.org/10.1364/OE.547138>.

7.3 Peer-reviewed conference papers

- [C1] J. Grünewald, V. Blickle, M. Allenberg-Rabe, P. Wagenblast, K. Wudy, Flexible and highly dynamic beam shaping for Laser-Based Powder Bed Fusion of metals, *Procedia CIRP* 111 (2022) 65–70. <https://doi.org/10.1016/j.procir.2022.08.124>.
- [C2] J. Grünewald, R. Prudlik, V. Holla, P. Kopp, R. Off, T. Stoll, S. Kollmannsberger, K. Wudy, Generating brick-like melt pools for Laser-Based Powder Bed Fusion using flexible beam shaping: a proof of concept on bead-on-plate single tracks, *Procedia CIRP* 124 (2024) 746–750. <https://doi.org/10.1016/j.procir.2024.08.216>.
- [C3] V. Holla, P. Kopp, J. Grünewald, P.M. Praegla, C. Meier, K. Wudy, S. Kollmannsberger, Laser beam shape optimization: Exploring alternative profiles to Gaussian-shaped laser beams in powder bed fusion of metals, University of Texas at Austin (2023). <https://doi.org/10.26153/tsw/50986>.
- [C4] K. Wudy, R. Prudlik, J. Grünewald, Melt pool stability criterion for beam shaping in laser-based powder bed fusion of metals, *Procedia CIRP* 137 (2025) 397–402. <https://doi.org/10.1016/j.procir.2025.02.274>.

7.4 Patents

- [P1] M. Allenberg-Rabe, V. Blickle, J. Grünewald, W. Meiners, P. Wagenblast (TRUMPF Laser- und Systemtechnik GmbH) EP000004185427A1, 2021.
- [P2] M. Allenberg-Rabe, V. Blickle, J. Grünewald, W. Meiners, P. Wagenblast (TRUMPF Laser- und Systemtechnik GmbH) EP000004185428A1, 2021.
- [P3] M. Allenberg-Rabe, V. Blickle, J. Grünewald, W. Meiners, P. Wagenblast (TRUMPF Laser- und Systemtechnik GmbH) EP000004185429A1, 2021.
- [P4] M. Allenberg-Rabe, V. Blickle, J. Grünewald, W. Meiners, P. Wagenblast (TRUMPF Laser- und Systemtechnik GmbH) EP000004185430A1, 2021.
- [P5] M. Allenberg-Rabe, V. Blickle, J. Grünewald, P. Wagenblast (TRUMPF Laser- und Systemtechnik GmbH) WO002022018148A1, 2021.
- [P6] M. Allenberg-Rabe, V. Blickle, J. Grünewald, P. Wagenblast (TRUMPF Laser- und Systemtechnik GmbH) WO002022018149A1, 2021.
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- [P8] M. Allenberg-Rabe, V. Blickle, J. Grünewald, P. Wagenblast, W. Meiners (TRUMPF Laser- und Systemtechnik GmbH) US2023143334A1, 2023.
- [P9] J. Grünewald, P. Wagenblast, M. Allenberg-Rabe, V. Blickle (TRUMPF Laser- und Systemtechnik GmbH) DE102020209173A1, 2020.

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- [P10] J. Grünwald, P. Wagenblast, M. Allenberg-Rabe, V. Blickle (TRUMPF Laser- und Systemtechnik GmbH) DE102020209172A1, 2020.
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- [P15] W. Meiners, P. Wagenblast, J. Grünwald, M. Allenberg-Rabe, V. Blickle (TRUMPF Laser- und Systemtechnik GmbH) WO002022018146A1, 2021.

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Appendix

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Article

Influence of Ring-Shaped Beam Profiles on Process Stability and Productivity in Laser-Based Powder Bed Fusion of AISI 316L

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Abstract: A major factor slowing down the establishment of additive manufacturing processes as production processes is insufficient reproducibility and productivity. Therefore, this work investigates the influence of ring-shaped beam profiles on process stability and productivity in laser-based powder bed fusion of AISI 316L. For this purpose, the weld track geometries of single tracks and multi-track segments with varying laser power, scan speed, hatch distance, and beam profile (Gaussian profile and three different ring-shaped profiles) are analyzed. To evaluate the process robustness, process windows are identified by classifying the generated single tracks into different process categories. The influence of the beam profiles on productivity is studied by analyzing the molten cross-sectional areas and volumes per time. When using ring-shaped beam profiles, the process windows are significantly larger (up to a laser power of 1050 W and a scanning speed of 1700 mm/s) than those of Gaussian beams (laser power up to 450 W and scanning speed up to 1100 mm/s), which suggests a higher process robustness and stability. With ring-shaped beam profiles, larger volumes can be stably melted per track and time. The weld tracks created with ring-shaped profiles are significantly wider than those generated with Gaussian profiles (up to factor 2 within the process window), allowing enlargement of the hatch distances. Due to the higher scanning speeds and the enlarged hatch distances for ring-shaped beam profiles, the process can be accelerated by a factor of approximately 2 in the parameter range investigated.

Keywords: additive manufacturing; laser-based powder bed fusion of metals; PBF-LB/M; stainless steel AISI 316L; beam shaping; donut-shaped profile; ring-shaped profile



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1. Introduction

Currently, continuous wave, vector-based exposure using Gaussian spots with spot diameters in the range from 40 to 120 µm represents the standard approach to laser-based powder bed fusion of metals (PBF-LB/M) [1]. The selected spot size is a compromise between achievable resolution and productivity. By enlarging the spot size and the scanning speed, productivity can be increased. The potential for increased productivity resulting from increases in laser power and scanning speed has been studied for several years [2]. To obtain sufficient melting and bonding of the individual melt tracks, the laser power must be increased to match the scanning speed. However, the two parameters cannot be increased ad infinitum because of various physical phenomena in the melt pool. The use of small Gaussian laser spots results in high intensities and temperature gradients when using PBF-LB/M, which cause different effects such as overheating or distinct Marangoni flows. Such effects can result in process defects such as keyholing and balling. The range of process parameters in which the process operates stably (no process defects such as keyholing or balling) and non-porous parts are generated can be termed the process window. With

larger process windows, more process parameter combinations over a larger range result in a stable process. This reduces the probability of process defects occurring. Thus, the process becomes more stable and robust.

Investigations by Yadroitsev et al. [2] have shown that for high laser powers and scanning speeds, process defects such as overheating of the melt pool, powder denudation, and balling occur. Overheating occurs due to the Gaussian energy input into the center of the melt pool and causes spatter [3]. When using Gaussian beam profiles, powder denudation is primarily caused by the Bernoulli effect induced by material vaporization in the melt tracks center [4]. Balling effects are described in the literature by Marangoni flows and the Plateau–Rayleigh capillary instability of a liquid cylinder, which represents the melt pool [5]. Balling occurs at a critical length to diameter ratio of the melt pool [6]. In addition to reducing the scanning speed or laser power and thus shortening the melt pool, the width can also be increased to lower the ratio. This can be implemented by increasing the power and enlarging the spot. However, centric overheating remains if the Gaussian shape of the intensity distribution is maintained. Modern methods of beam shaping make it possible to use not only Gaussian and top-hat (also known as flat top) profiles but also other beam shapes. The advantages of alternative (non-Gaussian) beam profiles have been discussed in the literature in respect to welding and additive manufacturing. In the literature, the beam shapes have been achieved using diffractive optical elements (DOEs) [7,8] and lenses [9] among others. Until now, knowledge in respect of the influence of alternative beam profiles on PBF-LB/M remains limited.

Rasch et al. [7] achieve a wider process window and reduced surface roughness when using a ring profile instead of a Gaussian profile (similar diameters) for the heat conduction welding of aluminum–copper alloys. Ayoola et al. [10] show that top-hat profiles produce flatter weld geometries in comparison with Gaussian profiles of the same diameters. However, the intensity distribution has no significant influence on the width of the weld track [10]. Studies on the influence of different beam profiles on PBF-LB/M have been carried out at Moscow State University of Technology (STANKIN). Zhirnov et al. [11], Metel et al. [12], and Okunkova et al. [13,14] generated weld tracks using a Gaussian profile, a top-hat profile, and a donut profile, varying laser power and varying scanning speed. They conducted their experiments with CoCrMo powder. Zhirnov et al. [11] achieve significantly larger process windows by using alternative beam profiles (top-hat and ring profile) instead of a Gaussian beam profile. Based on this, Metel et al. [12] show that the process can be extended with other beam shapes. For the top-hat profile used, power and speed can be increased by 14% with similar geometric parameters of the melt tracks, and, by using a donut profile, power and speed can be increased by 43% [12]. Okunkova et al. [13,14] show experimentally that the use of top-hat and donut-shaped profiles can reduce spatter formation and the width of the denudation zone. Wischeropp et al. [8] investigate the stability and productivity of PBF-LB of AlSi10Mg with a ring-shaped beam profile. They achieve larger melt pools, fewer defects (cracks, balling, and porosity) and thus a better process robustness due to the donut-shaped beam profile [8]. Schleifenbaum et al. [15] and Loh et al. [16] show the potential of increased build rates by using a top-hat profile. When processing steel, Schleifenbaum et al. [15] achieve build rates of 5 mm³/s with powder layer thicknesses of 100 µm and build rates of 20 mm³/s with powder layer thicknesses of 400 µm and relative densities of more than 99%.

In laser deep-penetration welding, the use of ring-shaped beam profiles in combination with centric top-hat or Gaussian profiles to increase process stabilization and reduce spatter formation is already established and commercially available [17–19].

All the presented prior research shows the existing potential of ring-shaped beam profiles in powder bed fusion. The alternative beam shapes usually have a larger spot size and produce larger melt pools, resulting in reduced resolution [15]. New programmable laser systems are now available that can be switched on the fly between Gaussian and ring-shaped beam profiles with different intensity distributions [20]. Thus, potential high resolutions can be combined with high productivities in the PBF-LB/M process. Systematic

investigations with the possible beam profiles have not yet been published. Therefore, this paper presents a study of the potential of three alternative ring-shaped beam profiles to increase the process robustness and productivity in PBF-LB/M of AISI 316L compared to a state-of-the-art Gaussian beam profile.

2. Materials and Methods

2.1. Feedstock Material

AISI 316L substrates are used for the experiments. This alloy is well-established in additive manufacturing using stainless steel. The substrate plates (see Figure 1) are laser cut from a cold-rolled sheet. To ensure reproducible surface properties, the substrates are sandblasted with corundum, resulting in a matte, rough surface ($S_a \approx 3 \mu\text{m}$, $S_z \approx 42 \mu\text{m}$).

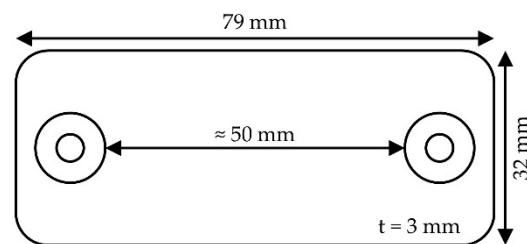


Figure 1. Schematic substrate plates laser cut from cold-rolled 3 mm sheet.

The experiments are carried out without powder. He et al. [21] show that experiments without powder are transferable to PBF-LB/M. Adding powder leads to a small shift of the process window due to a slightly more pronounced balling tendency [21].

2.2. Job Design

Single tracks and multi-track segments are exposed to analyze the shape and dimensions of the weld tracks with different parameter sets. The design and orientation of the tracks and segments on the substrate are shown in Figure 2. The single tracks had a length of 15 mm. The segments were manufactured with an edge length of 5 mm and different hatch distances h . To avoid the interaction of laser radiation with process emissions, the single tracks are exposed in the opposite direction to the shielding gas flow. When exposing the multi-track segments, the tracks are scanned perpendicularly to the shielding gas flow and one after the other in the opposite direction to the shielding gas flow.

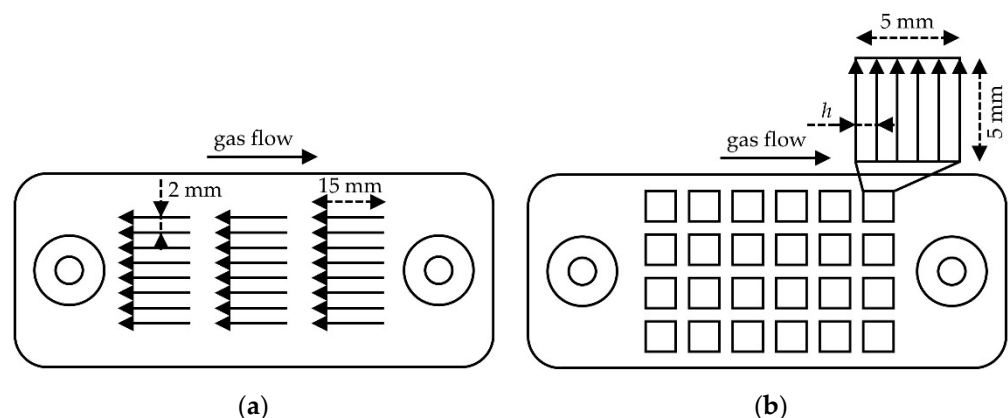


Figure 2. Schematic design of single tracks (a) and multi-track segments (b).

2.3. Experimental System

The experiments are carried out on an in-house developed test rig and process chamber (shown in Figure 3). The chamber was flooded with argon to a residual oxygen content of 0.2%. Argon flows over the substrate plate at a volume flow rate of 18 l/min. An AFX-1000 (nLIGHT Inc., Vancouver WA, USA) with seven switchable beam profiles is utilized

as the beam source. The switchable beam profiles are termed as “Variable Beam Parameter Product (VBPP)”. The applied beam profiles are shown in Figure 4 as 2D distributions in (a) and as cross-sections in (b). As shown in Figure 4, the ring-shaped beam profiles differ in power ratio between the central peak and the surrounding ring. Collimation is performed using a 60 mm collimator with an aperture (nLIGHT Inc., Vancouver WA, USA). The beam deflection was realized with an Elephant 50 (ARGES GmbH, Wackersdorf, Germany) scan system. The resulting beam diameters and power ratios of the central peak and surrounded ring of the used beam profiles are listed in Table 1.

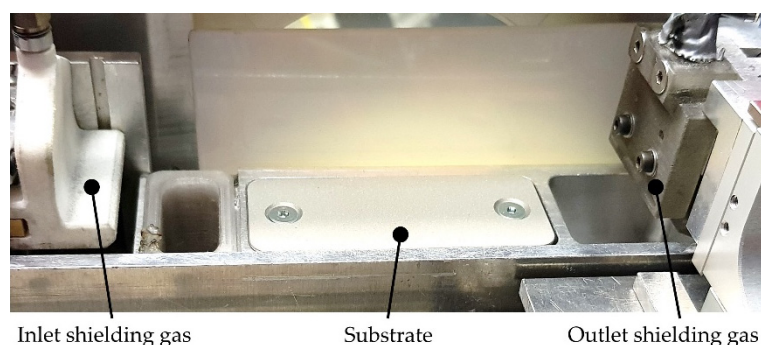


Figure 3. Photo of the process zone inside the process chamber.

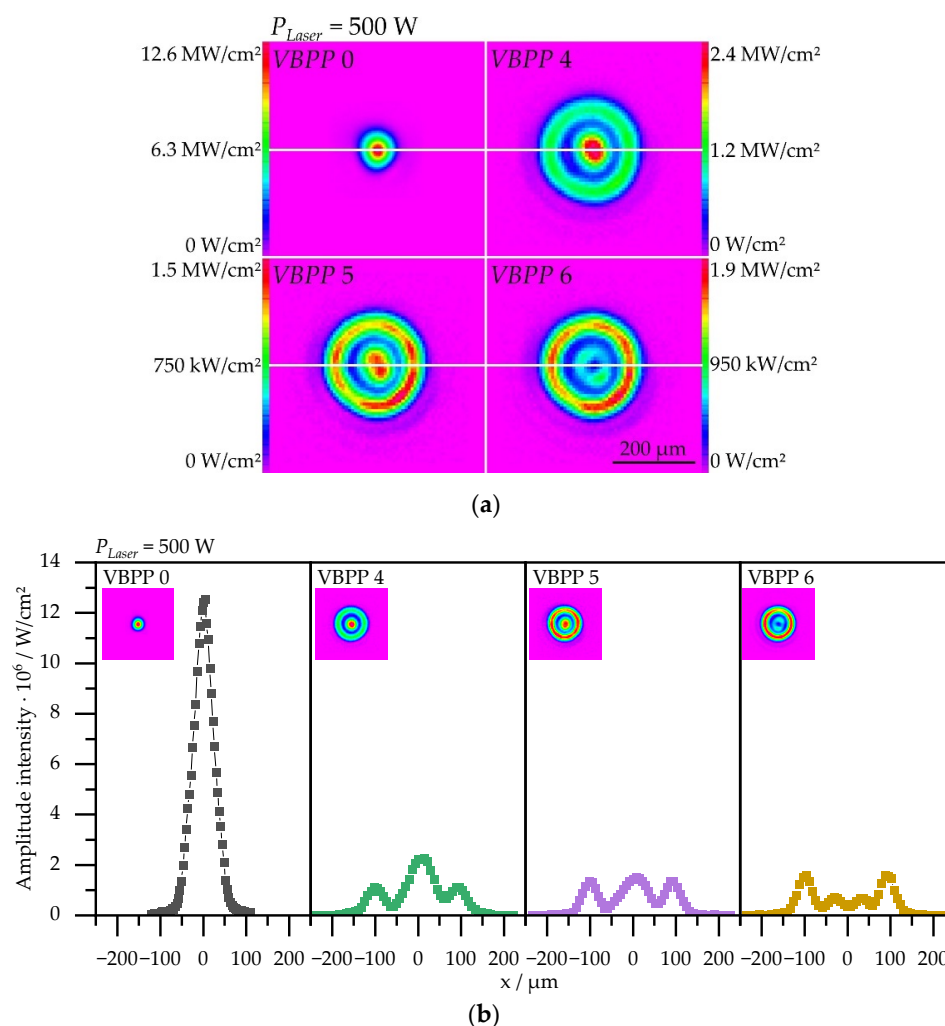


Figure 4. Intensity distributions of the beam profiles used at $P_{Laser} = 500\text{ W}$ as 2D distributions (a) and cross-sections (b).

Table 1. Power distribution according to manufacturer adapted from [22] and beam diameter at $P_{Laser} = 500$ W calculated using the 2nd moment method according to [23] for used beam profiles.

Profile Name VBPP	Peak Power	Ring Power	Spot Diameter $2w_0$
0	100%	0%	107 μm
4	40%	60%	253 μm
5	20%	80%	262 μm
6	10%	90%	271 μm

2.4. Process Parameters

The process parameter sets for the single-track and multi-track experiments with the different beam profiles are summarized in Tables 2 and 3. An overview of all used parameters, symbols and abbreviations is listed in Appendix A. The laser power P_{Laser} , the scanning speed v_{scan} , and the hatch distance h are varied for the different beam profiles. A delay time of 2 s between the single tracks and 1 min between the various multi-track segments is implemented to avoid beam-fume interactions. The individual tracks of a segment are manufactured one immediately after the other to create realistic process environments.

Table 2. Summarized process parameter sets of the single tracks.

Gaussian Profile (VBPP 0)		Ring-Shaped Profiles (VBPP 4, 5, 6)	
Laser Power P_{Laser} in W	Scanning Speed v_{scan} in mm/s	Laser Power P_{Laser} in W	Scanning Speed v_{scan} in mm/s
200	500, 800, 1100, 1400, 1700	300	500, 800, 1100, 1400, 1700
300	500, 800, 1100, 1400, 1700	450	500, 800, 1100, 1400, 1700
400	500, 800, 1100, 1400, 1700	600	500, 800, 1100, 1400, 1700
450	500, 800, 1100, 1400, 1700	750	500, 800, 1100, 1400, 1700
500	500, 800, 1100, 1400, 1700	900	500, 800, 1100, 1400, 1700
600	500, 800, 1100, 1400, 1700	1050	500, 800, 1100, 1400, 1700

Table 3. Summarized process parameter sets of the multi-track segments.

Gaussian Profile (VBPP 0)			Ring-Shaped Profiles (VBPP 4, 5, 6)		
Laser Power P_{Laser} in W	Scanning Speed v_{scan} in mm/s	Hatch Distance h in μm	Laser Power P_{Laser} in W	Scanning Speed v_{scan} in mm/s	Hatch Distance h in μm
200	500, 800, 1100, 1400	85, 120, 155	450	500, 800, 1100, 1400	170, 240, 310
300	500, 800, 1100, 1400	85, 120, 155	600	500, 800, 1100, 1400	170, 240, 310
400	500, 800, 1100, 1400	85, 120, 155	750	500, 800, 1100, 1400	170, 240, 310
500	800, 1100, 1400, 1700	85, 120, 155	900	800, 1100, 1400, 1700	170, 240, 310

The single-track experiments (Table 2) are carried out to detect suitable process windows and hatch distances for the multi-track segments. A total of 120 parameter sets are studied. Only one track per parameter set is generated and analyzed in two cross-sections of the track. For this reason, no standard deviations are listed in the evaluations of the single-track experiments. The mean values shown result from the evaluated cross-sections of one track. The multi-track segments (Table 3) are statistically validated. For each parameter set, three segments are exposed with 5 weld tracks analyzed in the center of each segment. The mean value and standard deviation are calculated from the 15 tracks analyzed per parameter set. In total, 192 parameter sets are tested with 3 repetitions each.

2.5. Component Properties

In order to investigate the influence of the various parameters on the welding result, the cross-sectional geometries of the generated weld tracks are analyzed. In addition, the roughness of the multi-track segments is evaluated.

2.5.1. Cross-Sectional Geometries

To enable the analysis of the weld track geometries, cross-sectional micrographs are prepared. For this purpose, the specimens are cut centrally and embedded in a transparent epoxy. Afterwards, the specimens are wet ground in three stages using silicon carbide paper (grit 180, 360 and 1200) and polished in two stages using polishing pastes (3 μm and 1 μm). Finally, the specimens are etched with an Adler etchant. The cross-sections of the weld tracks are documented using an optical microscope. Based on the microscope images, the weld width w_w , welding penetration depth δ_w , and the molten cross-sectional area A_m are determined. A schematic illustration of the process and the resulting measured values is shown in Figure 5. In addition, it is qualitatively determined whether balling effects occur. Depending on the weld track geometries, the specimens are classified into one or more of the following categories adapted from [8,21] (see Figure 6):

1. Regular: The weld track topology is homogenous and smooth.
2. Keyholing $\rightarrow A = \delta_w / w_w > 0.8$: The aspect ratio of the welding penetration depth to width of the weld track is greater than 0.8 [24]. Increased energy input can cause a vapor capillary (“keyhole”) to form in the melt pool, which causes the high aspect ratios.
3. Balling: The weld track is characterized by an irregular shaped weld track, with single melt “balls” along the track [25,26]. Plateau–Rayleigh instability causes very long melt pools to break up and individual melt spheres to form [26,27].
4. Protrusion and depression: The topology of the corresponding weld tracks is characterized by protrusions in the center and depressions on the sides of the weld tracks. This weld track shape is created by long melt pools with increased backward flow of the molten material due to the Marangoni effect and recoil pressure [28]. Plateau–Rayleigh instability [25,27] causes the melt pool to form a protrusion that solidifies before it can break into individual melt balls.
5. Undermelting $\rightarrow \delta_w < 40 \mu\text{m}$: The welding penetration depth is less than the powder layer thickness (40 μm in this study). In such weld tracks, the energy applied is usually too low.

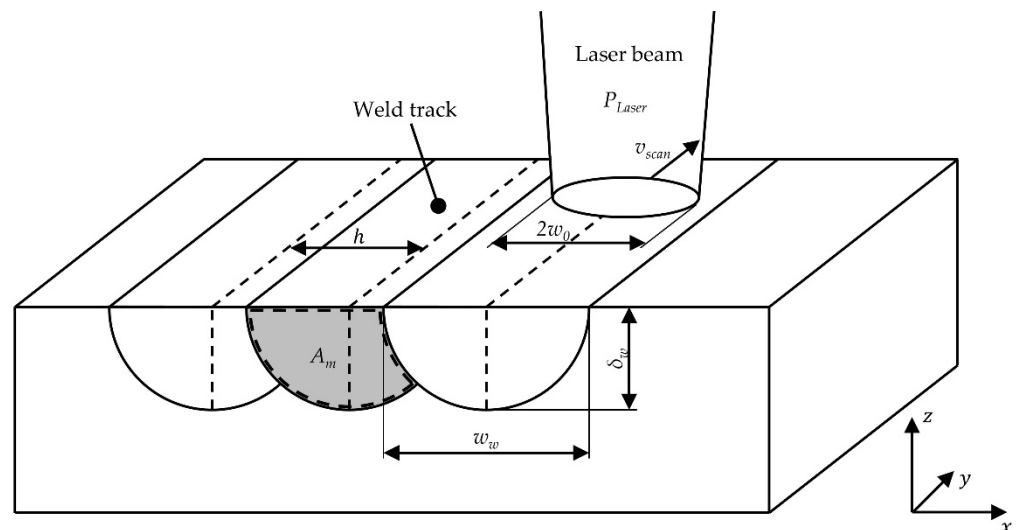


Figure 5. Schematic illustration of the process and the resulting measured values.

2.5.2. Roughness Measurement

A 3D profilometer (Keyence VR-3100, Ōsaka, Japan) is used to measure the roughness of the multi-track segments via fringe projection. The optical measurement enables a local description of the whole topography and thus the evaluation of the associated roughness values S_i [29]. In this study, the areal arithmetic average roughness S_a (according to DIN EN ISO 25178-2 [30]) is used for roughness determination.

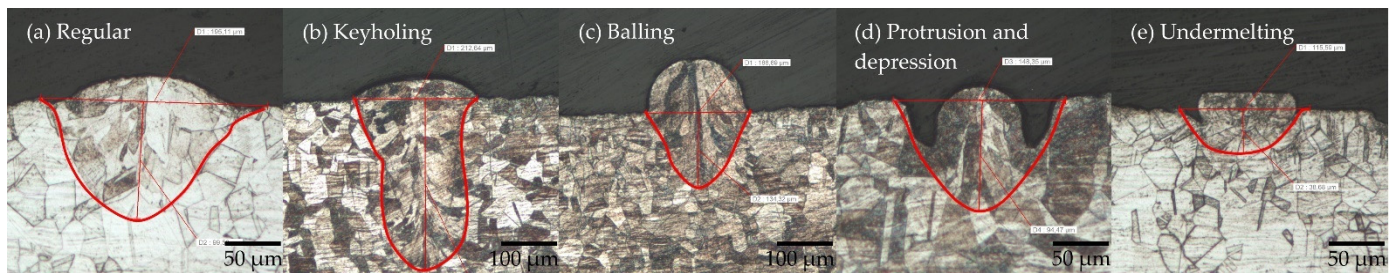


Figure 6. Typical microscopy images and their classification into the different categories: (a) Regular, (b) Keyholing, (c) Balling, (d) Protrusion and depression, and (e) Undermelting.

3. Results and Discussion

3.1. Influence of Laser Power and Scanning Speed on Melt Pool Geometry for Single Tracks

The single-track experiments aim at detecting process windows for the different beam profiles. This is based on the categorization of the weld tracks described in Section 2.4. The categories “Regular” and “Protrusion and depression” are considered as functional processes, because continuous weld tracks with suitable weld penetration depths are produced. “Protrusion and depression” does not necessarily tend to an unstable process and thus to component defects such as porosity. However, if the protrusion and depression are very distinct, weld tracks in this category can also lead to non-qualification of the corresponding parameter set in subsequent investigations. In addition to the detection of the process windows, values for suitable hatch distances of the multi-track segments are detected via the weld width of the single tracks, and the degree of laser energy coupling for the different beam profiles is compared. Figure 7 shows the weld widths w_w in (a), welding penetration depths δ_w in (b), and molten cross-sectional area A_m in (c) measured on the cross-sections of the welds. To prevent the graphs from becoming overly complicated, only the power levels at 300, 450, and 600 W are shown for beam profile VBPP 0.

For all the investigated beam profiles, the weld width w_w tends to decrease for increasing scanning speeds v_{scan} and decreasing laser powers P_{Laser} (see Figure 7a). By using a Gaussian beam profile (VBPP 0), the welding penetration depths δ_w decrease for increasing scanning speeds v_{scan} and laser powers P_{Laser} (Figure 7b). Due to the exposure with beam profiles VBPP 4, 5, and 6, the welding penetration depths δ_w decrease for increasing scan speeds v_{scan} . For the laser power variances, the welding penetration depths δ_w increase up to around 600–750 W. With further increasing laser powers P_{Laser} , the welding penetration depths δ_w tend to decrease. Likewise, molten cross-sectional areas A_m tend to the weld width w_w (see Figure 7c). With increasing scan speeds v_{scan} and decreasing laser powers P_{Laser} , the molten cross-sectional areas A_m of the single tracks become smaller. At low scanning speeds (500 mm/s), the molten cross-sectional areas A_m decrease for the very high laser powers (600 W at VBPP 0 and 1050 W at VBPP 4, 5, 6). The tendencies of the weld widths w_w and the molten cross-sectional areas A_m can be explained by the energy input of the laser, which can be estimated via the line energy E_l . The line energy is calculated from the laser power P_{Laser} and the scanning speed v_{scan} according to

$$E_l = \frac{P_{Laser}}{v_{scan}}. \quad (1)$$

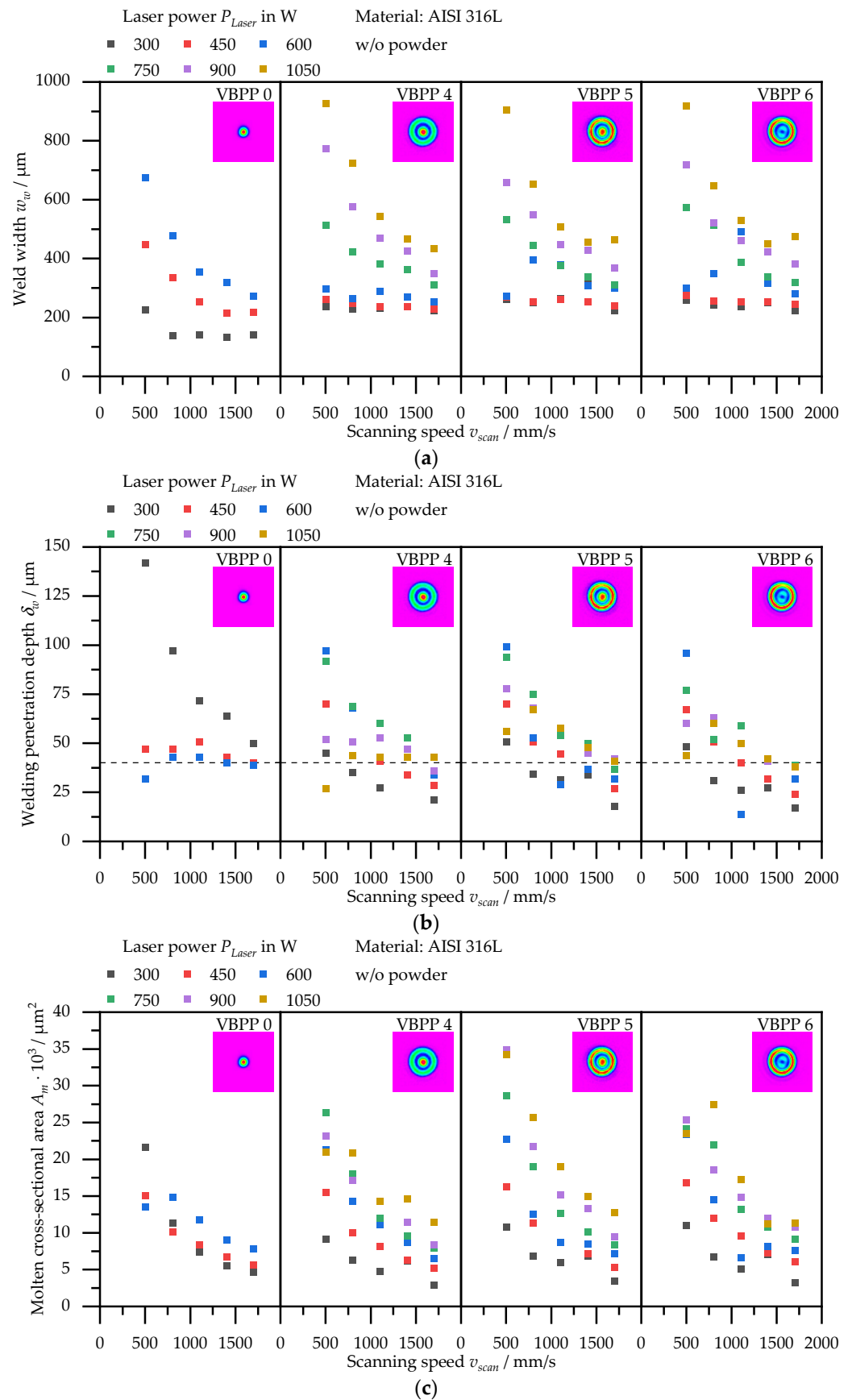


Figure 7. Weld width w_w (a), welding penetration depth δ_w (b), and molten cross-sectional area A_m (c) of single tracks as a function of scanning speed v_{scan} and laser power P_{Laser} .

Lower scanning speeds and/or higher laser powers result in higher line energies. In theory, the resulting higher energy input results in a larger melt pool, which can be expressed in larger weld widths w_w , welding penetration depths δ_w , and larger molten cross-sectional areas A_m . Where the weld widths w_w and molten cross-sectional areas A_m are concerned, this reasoning explains the measured values. However, the assumption could not be confirmed for the welding penetration depths. Up to a certain power/energy, the penetration depths increase. For laser powers higher than 300 W for the Gaussian beam profile and 750 W for the ring-shaped profiles, the penetration depth is reduced. Reasons for the low welding penetration depths could be laser–fume interactions. Interaction of the laser with the outflowing process emissions can widen the laser beam [31]. This results in an enlargement of the spot, which on the one hand creates wider melt pools and on the other hand leads to shallower welding penetration depth due to the reduced intensity. Another possible reason is thermal lensing. With thermal lenses, the focal length of the system changes as a function of laser power because the refractive index changes due to temperature gradients in the optical material [32]. In this way, at high laser powers, the focus usually shifts toward the scanning system. However, the focus position was found to be independent of power via burn-in tests. This rules out the thermal lens as a cause, and hence, the theory of laser–fume interactions as the cause of the reduced welding penetration depths seems most probable. As the measured values are not statistically validated, no valid statement can be made.

According to Wischeropp et al. [33], statements about the degree of energy coupling of the laser beam can be made by analyzing the molten cross-sectional areas of the weld tracks. The larger the cross-sectional area for a constant line energy, the higher the energy coupling [33]. Figure 8 shows the molten cross-sectional area A_m as a function of the line energy. Since the cross-sectional areas are similar for all beam profiles, any influence of the beam shape on the energy coupling cannot be identified. Although the molten cross-sectional areas are in similar ranges at similar line energies, the weld track geometries can be quite diverse between the Gaussian and the ring-shaped beam profiles (see Figure 9).

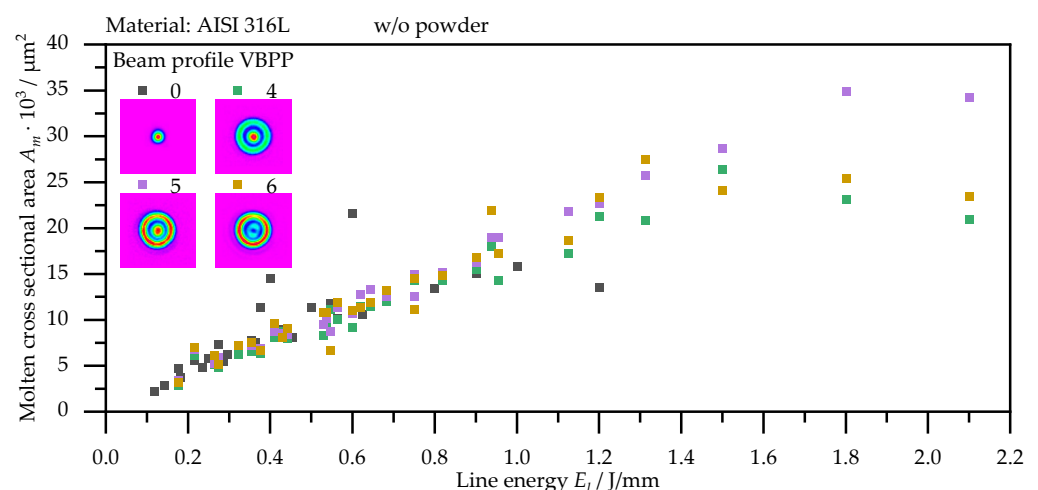


Figure 8. Molten cross-sectional area A_m of single tracks as a function of line energy E_l .

Figure 10 summarizes the classification of all investigated parameter sets. As “Protrusion and depression” or “Balling” and “Undermelting” partly take place simultaneously, some points on the P_{Laser} – v_{scan} process map overlap. As can be seen in Figure 10, the process windows between the Gaussian beam profile VBPP 0 and the ring-shaped beam profiles VBPP 4, 5, 6 differ significantly. It is not possible to run a stable process with the Gaussian beam profile in these investigations if laser powers P_{Laser} are 500 W or greater, or scanning speeds v_{scan} greater than 1100 mm/s are used. The ring-shaped beam profiles significantly enlarge the process windows. At very high laser powers of 1050 W, single tracks without keyholing and balling effects can be generated with these profiles. However, the protrusion

and depression tendency at very high laser powers and scanning speeds lead to significant protrusions in these single tracks. One reason for the significantly larger process windows when using the ring-shaped profiles is the larger area of the spots. This reduces the intensities compared to the Gaussian profile at the same laser powers (see Figure 4b), which avoids process defects such as keyholing. However, these reduced intensities also result in insufficient welding penetration depth being produced at low powers P_{Laser} below 750 W and high scan speeds v_{scan} above 1400 mm/s (see Figure 7b).

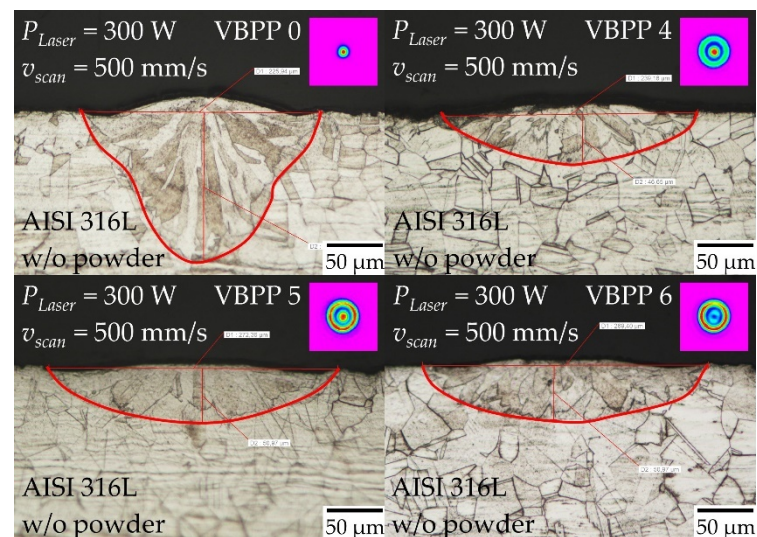


Figure 9. Direct comparison of the weld track geometries at constant process parameters laser power $P_{Laser} = 300$ W and scanning speed $v_{scan} = 500$ mm/s with different beam profiles.

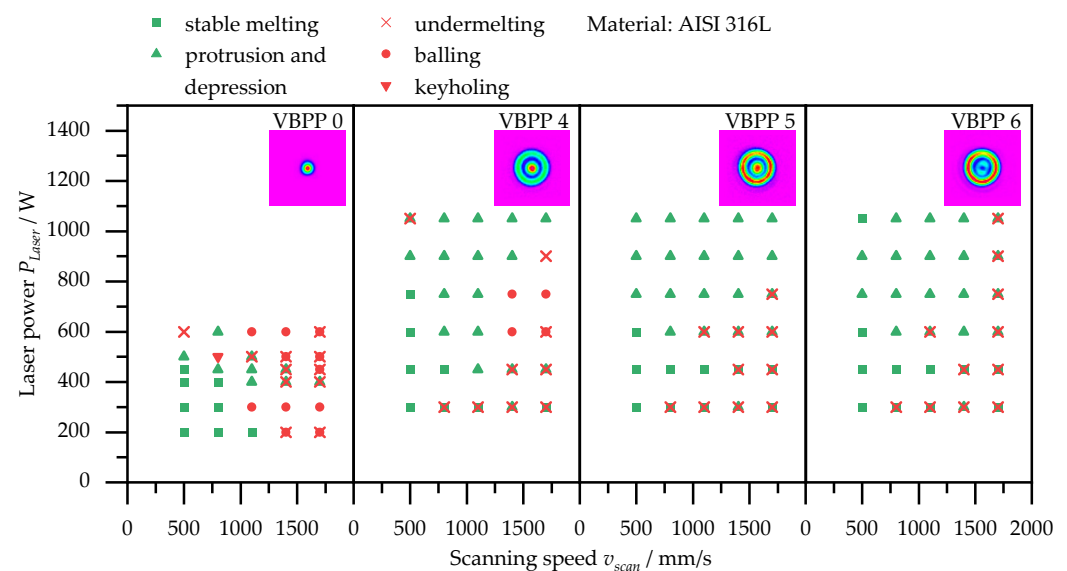


Figure 10. P_{Laser} - v_{scan} process map of single-track experiments with AISI 316L.

The enlarged process windows of the ring-shaped beam profiles confirm the findings of Rasch et al. [7], Wischeropp et al. [8], Zhirnov et al. [11], and Metel et al. [12], who also showed significantly larger process windows and a reduction of balling and keyholing when using ring-shaped beam profiles. In the referred previous research, beam profiles of similar diameters are used in the interaction zone. Contrary to this, this investigation was performed with different spot sizes between the Gaussian and the ring-shaped profiles (see Table 1). Compared to the Gaussian beam profile, the diameters of the ring-shaped beam profiles extend by approximately 250%. Due to the different spot sizes, the direct

comparability of the beam profiles is limited due to the different peak intensities at the same powers (see Figure 6b). However, the potential of combining the beam profiles becomes visible. In the long term, the Gaussian beam profile can be used for area with high required resolution, and the ring-shaped profiles can be used to fill in the hatch areas, which often represent a majority of the areas to be exposed. Whether defocusing or widening a Gaussian beam profile achieves similar results was not investigated in this study, as the potential of enlarging the process window for similar beam diameters has already been demonstrated in [7,8,11,12].

Due to the process behavior mentioned, a significant potential for increasing productivity while maintaining a stable process is indicated. With the ring-shaped beam profiles, the speed can be increased by a factor of approximately 1.3 at more than double the laser power. The weld widths increase on average by a factor of about 1.5 with stable process behavior. By increasing the hatch distance, this can be used to increase productivity. Based on these data, productivity can be increased by a factor of about 2 when combining the two options.

3.2. Influence of Laser Power, Scanning Speed, and Hatch Distance on Relative Overlap, Molten Cross-Sectional Area, and Molten Volume per Time

Based on the results of the single tracks, the multi-track segment experiments are carried out with the parameter sets from Table 3. The boundary areas of the process window are still investigated. Compared to the single tracks, the following adjustments are made. For the Gaussian profile VBPP 0, the maximum and intermediate power levels at 600 W and 450 W are not considered further. As a result of the very high protrusions at the maximum power levels at 1050 W and the insufficient welding penetration depth of the lowest power levels 300 W for the ring-shaped profiles, these levels are not investigated further. The speed levels at 1700 mm/s are not considered for the three lower power levels of all beam profiles because of the insufficient welding penetration depths. For the maximum power levels, the scan speeds are shifted, the 500 mm/s speed is not investigated and is replaced by the 1700 mm/s speed.

Due to the hatch of several weld tracks with hatch distances h less than the weld width w_w , an overlap occurs in the multi-track segments. As a result of the missing reference edge, the welding penetration depth can no longer be measured. The weld width w_w is determined from twice the distance from the center of the weld track to the weld track edge (twice half the width of the weld track). From this dimension, the relative overlap of the tracks O_r is determined with the hatch distance h according to

$$O_r = \frac{w_w - h}{w_w}. \quad (2)$$

According to Di et al. [34], an overlap of 30% is ideal for a productive and stable processing of AISI 316L using PBF-LB/M. Dong et al. [35] indicate an optimum overlap at 54% for the highest possible densities of AISI 316L components. Since the ring-shaped beam profiles result in much wider, shallower weld tracks, it may be possible that even a smaller overlap is sufficient to provide a stable process for a specimen with a high density.

The molten volume per time V_m is determined, a characteristic productivity value, and this can be calculated as a product of the molten cross-sectional area A_m and the scanning speed v_{scan} according to

$$V_m = A_m \times v_{scan}. \quad (3)$$

By measuring the visible cross-sectional area per track at the multi-track segments, the influence of the hatch distance is also considered (see Figure 5). Figure 11a shows the relative overlap O_r , (b) shows the molten cross-sectional area A_m , and (c) shows the molten volume per time V_m of the multi-track segments as a function of scanning speed v_{scan} , laser power P_{Laser} , and hatch distance h . To clarify the graphs, only two power levels and two levels of hatch distance per beam profile are shown in Figure 11. All parameter sets are shown in Figure 12. The relative overlap O_r decreases for higher scanning speeds v_{scan} ,

lower laser powers P_{Laser} , and higher hatch distances h . A negative overlap O_r means that the weld tracks do not overlap but are spaced apart. If the hatch distance h is varied at constant laser powers and scanning speeds, the overlap also decreases at higher hatch distances. This behavior can be explained by the smaller melt pools and thus weld widths at higher scan speeds and lower laser powers. With constant hatch distances, narrower welds result in lower relative overlaps according to Equation (2). At constant laser powers and scanning speeds, similar weld widths are produced. The decrease in relative overlap at larger hatch distances h is expected in accordance with Equation (2). The molten cross-sectional area of the multi-track segments exhibits a similar tendency to that of the single tracks for constant hatch distances. With increasing scan speeds and decreasing laser powers, the molten cross-sectional area of a track decreases. If these parameters are kept constant and the hatch distance is varied, the molten cross-sectional areas become larger for larger hatch distances. The cross-sectional areas are measured excluding the section of the track that is remelted by the following track. This means that the larger the overlap or the smaller the hatch distance, the smaller the molten cross-sectional area per track because a larger part of the track is remelted.

Figure 11c combines the molten cross-sectional area A_m and the scan speed v_{scan} to form the molten volume per time V_m in accordance with Equation (3). Analogously to the molten cross-sectional area A_m , the molten volume per time V_m increases for increased hatch distances and laser powers. It is noticeable that this characteristic value for the productivity tends to increase with scanning speed for constant powers and hatch distances. Firstly, this results from the reduced melt pool size and thus reduced relative overlap for increasing scanning speeds. However, the trend is also present for parameter sets that have negative relative overlap, i.e., where a change in this parameter results in no change in the molten cross-sectional area due to overlap. Secondly, this means that a change in the scanning speed influences the melting process itself. Higher scanning speeds tend to result in a higher molten volume per unit of time. One reason for this may be a change in laser energy coupling due to the shorter laser–material interaction at higher scanning speeds. According to this theory, relatively more energy would be absorbed in a shorter time, resulting in a larger molten volume per time. Another reason may be lower overheating of the melt pool due to shortened laser–material interaction at higher speeds. This means that less energy is used to superheat the material, and proportionately more is used for phase transformation. This results in a more “efficient” use of energy.

The process parameters laser power P_{Laser} , scanning speed v_{scan} , and hatch distance h can be combined to form the area energy E_A according to

$$E_A = \frac{P_{Laser}}{v_{scan} \cdot h}. \quad (4)$$

Figure 12a shows the relative overlap O_r , (b) shows the molten cross-sectional area A_m , and (c) shows the molten volume per time V_m of the multi-track segments as a function of the area energy E_A . The colors and symbols are the same as in Figure 11.

For increasing applied area energies E_A , the relative overlap O_r becomes larger. For constant area energies, the overlap increases for greater laser powers and reduced hatch distances (see Figure 12a). As described above, the reasons for this are the larger melt pools and thus weld widths at higher laser powers and smaller hatch distances. The comparison of Figures 11a and 12a indicates that it is not the variation of individual process parameters but rather their combination to form the area energy E_A that influences the overlap. If a suitable overlap is assumed based on Di et al. [34] and Dong et al. [35], a range of O_r between 30% and 54% is suitable. This range is reached for many of the parameter sets with area energies E_A between 2 and 4 J/mm³.

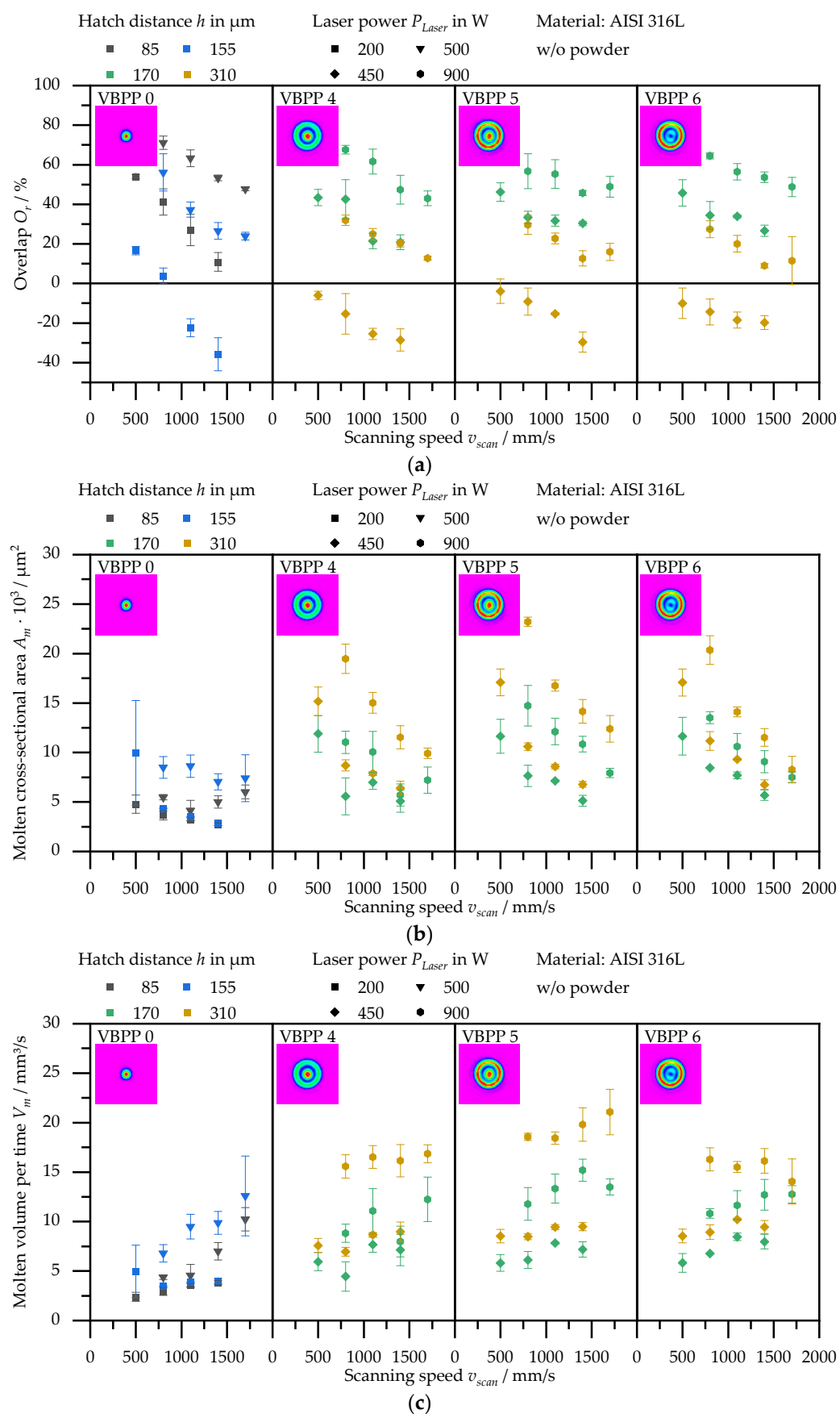


Figure 11. Overlap O_r (a), molten cross-sectional area A_m (b), and molten volume per time V_m (c) of multi-track segments as a function of scanning speed v_{scan} , laser power P_{Laser} , and hatch distance h .

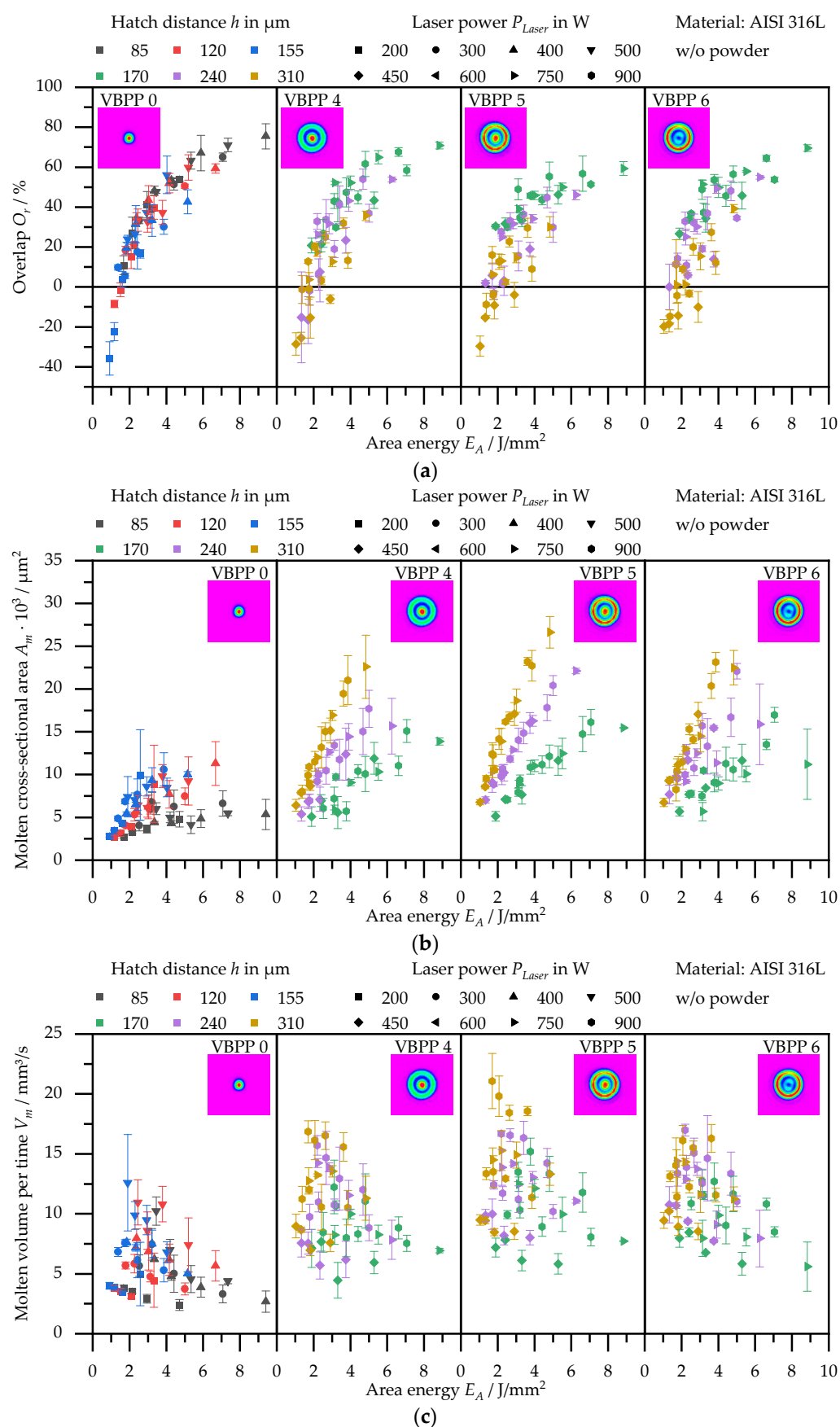


Figure 12. Overlap O_r (a), molten cross-sectional area A_m (b), and molten volume per time V_m (c) of multi-track segments as a function of area energy E_A , laser power P_{Laser} , and hatch distance h .

The molten cross-sectional area A_m increases approximately linearly with increasing area energy E_A for constant hatch distances h (see Figure 12b). This indicates a dependency of the molten cross-sectional area A_m per track on the hatch distance h and the quotient of laser power P_{Laser} and scan speed v_{scan} (line energy E_l according to Equation (1)). For constant area energies, the molten cross-sectional areas increase in size with greater hatch distances. Reasons for this are the dependencies between hatch distance and relative overlap discussed above.

Figure 12c illustrates that the molten volume per time does not reach a maximum at high area energy. Rather, the molten volumes per time reach their maxima in a range of area energies E_A between 2 and 4 J/mm³ for all four beam profiles. One reason for this is the impact of the scan velocity and hatch distance on the molten volume per time, while also forming the divisor in the calculation of the area energy. Figure 12c also shows that the molten volume per time reaches a maximum at constant area energy with high powers and hatch distances. Accordingly, the scanning speed must be selected in such a way that suitable area energies and overlaps result, depending on the power and hatching. Compared to Gaussian beam profiles, significantly higher volumes can be molten per time with suitable overlaps when using ring-shaped profiles (VBPP 4, 5, 6) at equal area energies ($E_A \approx 2 \dots 4$ J/mm³). This indicates that there is a significant potential for increasing productivity. It should be noted that the power levels and hatch distances used with the Gaussian beam profile were significantly lower than the parameters of the ring-shaped profiles. Higher values for laser power and hatch distances are not considered for the Gaussian profile due to the lack of process stability and the reduced weld widths w_w .

Figure 13 shows the roughness S_a as a function of the area energy E_A . Similarly to the molten volume per time, the roughness reaches a maximum in the range of the area energy $E_A \approx 2 \dots 4$ J/mm³. It is likely that parameter sets with high hatch distances and laser powers in particular will reach high roughness values. Despite the lower values for power and hatch distance for the Gaussian beam profile (VBPP 0), the roughnesses are at a slightly higher level than the roughnesses when using ring-shaped beam profiles. One reason for this could be the significantly larger weld widths and relatively lower protrusions of the weld tracks produced with ring-shaped beam profiles.

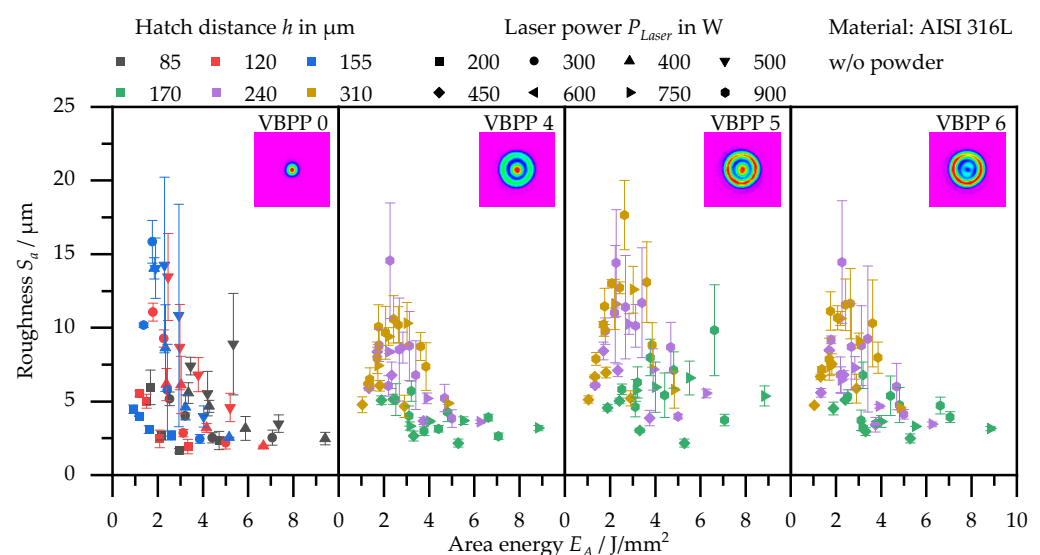


Figure 13. Roughness S_a of multi-track segments as a function of area energy E_A , laser power P_{Laser} , and hatch distance h .

In summary, a majority of the parameter sets with area energies E_A between 2 and 4 J/mm³ generate sufficient overlaps. In this area energy range, the highest values of molten volumes per time are achieved for high laser powers and hatch distances, which indicates high process productivity. By applying ring-shaped beam profiles, the molten

volumes per time V_m can be improved up to $20 \text{ mm}^3/\text{s}$ compared to the conventional Gaussian beam profiles (less than $12 \text{ mm}^3/\text{s}$). The roughness S_a is slightly reduced when using ring-shaped beam profiles than when using a Gaussian beam.

Similar values for the molten volume per time V_m are achieved by Schleifenbaum et al. [15] when processing steel and using a top-hat profile. However, the increase in productivity in [15] is at the expense of roughness and resolution. Gusarov et al. [3] and Loh et al. [16] have already described the potential of using alternative intensity distributions with larger beam diameters to increase build rates without decreasing part quality. The potential described can be confirmed by the shown increases in molten volume per time V_m with similar roughness values S_a of the multi-track segments.

4. Conclusions

This study investigated the influence of ring-shaped beam profiles on process stability and productivity in laser-based powder bed fusion of AISI 316L. The results of experiments with different parameter sets of Gaussian and ring-shaped beam profiles have been compared. Based on the classification of single tracks into different process categories, process windows were identified, whose size allows a deduction of the process stability and robustness. Furthermore, the influences of different beam shapes on productivity were investigated on the basis of multi-track segments. The data and relationships shown and discussed allow some generalized conclusions to be reached about the potential of the used ring-shaped beam shapes during processing AISI 316L in PBF-LB/M:

- The PBF-LB/M process itself can be stabilized at high laser powers and scanning speeds using ring-shaped beam profiles. Due to the reduction of process-limiting phenomena such as keyholing and balling, the process window for ring-shaped beam profiles increases significantly
- The process windows shift to higher laser powers when using ring-shaped profiles due to the larger beam diameters and thus decreasing intensities.
- An influence of the beam profiles on the degree of coupling of the laser energy is not found. For equal line energies, comparable cross-sectional areas are melted irrespective of the beam profile.
- When manufacturing multi-track segments, optimum area energies between 2 and 4 J/mm^3 are identified for processing AISI 316L, independent of the beam profile used.
- In the experiments, the molten volume per time can be increased from less than $12 \text{ mm}^3/\text{s}$ (for Gaussian beams) to $20 \text{ mm}^3/\text{s}$ (for ring-shaped profiles) with slightly reduced roughness S_a .

Thus, the use of ring-shaped beam profiles in PBF-LB/M offers enormous potential for stabilizing the process while simultaneously accelerating it. To be able to assess the potential of the alternative beam shapes in more detail, the maximum size of a stable melt pool has still to be investigated, because this limit has not yet been reached with the ring-shaped beam profiles shown. With even larger melt pools, productivity could potentially be increased even further. Additionally, further experiments should investigate how the alternative beam profiles influence the properties in resulting components (e.g., microstructure) due to the changed energy input.

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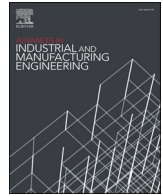
Appendix A

Symbol	Explanation	Unit
A	Aspect ratio	-
A_m	Molten cross-sectional area	μm^2
E_l	Line energy	J/mm
E_A	Area energy	J/mm ²
h	Hatch distance	μm
O_r	Relative overlap	%
P_{Laser}	Laser power	W
S_a	Roughness: Arithmetical mean height	μm
S_z	Roughness: Maximum height	μm
V_m	Molten volume per time	$\mu\text{m}^3/\text{s}$
v_{scan}	Scanning speed	mm/s
w_w	Width of the weld track	μm
w_0	Spot radius	μm
δ_w	Welding penetration depth	μm
Abbreviation	Explanation	
PBF-LB/M	Laser-based powder bed fusion of metals	
VBPP	Variable Beam Parameter Product	

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Scaling laws for ring-shaped beam profiles in laser-based powder bed fusion of metals

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ABSTRACT

The application of alternative beam shapes is a current research trend to stabilize and accelerate the laser-based powder bed fusion of metals process. Although many publications show a reduced process dynamic and an enlargement of the process window for dense components using non-Gaussian beam profiles, a generally valid correlation between the energy input - in terms of the beam shape and the process parameters - and the melting mode is lacking. Consequently, intensive experimental work is required to qualify process parameters for alternative beam profiles. The present work aims to reduce this experimental effort for the parameter qualification of alternative beam profiles by estimating the melting modes based on dimensionless parameters. For this purpose, a simple heat conduction model is applied to a new database of melt track widths and depths generated with various ring-shaped beam profiles with different spot sizes. The approach shows a correlation between the dimensionless enthalpy and the melt track depth and width if the 2nd moment method is used to determine the spot size of the laser beam profiles. Finally, introducing a maximum dimensionless enthalpy considering the peak intensity of the beam profile used enables the estimation of the melting mode. Regardless of the beam profile, the transition from conduction mode to keyhole mode occurs between maximum dimensionless enthalpies of 6.25 ± 0.85 and 8.65 ± 0.30 .

Nomenclature

A	Absorption
c_p	Specific heat capacity
h_s	Enthalpy at melting
I	Intensity
L_f	Latent heat of fusion
M	Magnification
Pe	Peclet number
P_{laser}	Laser power
T_0	Ambient temperature
T_m	Melting temperature
v_{scan}	Scanning speed
w_0	Spot radius
x, y, z	Spatial coordinates
x^*, y^*, z^*	Dimensionless spatial coordinates
A	Thermal diffusivity
δ_{diff}	Thermal diffusion depth
$\Delta H/h_s$	Dimensionless enthalpy
$\Delta H_{\text{max}}/h_s$	Maximum dimensionless enthalpy
ρ	Density
ψ	Peak-to-mean intensity ratio

1. Introduction

Laser-based powder bed fusion of metals (PBF-LB/M) is gaining popularity as an industrial manufacturing process. In PBF-LB/M, metallic components are built up by layer-by-layer selective exposure and fusing a powder bed with a high-power laser. Selective exposure is performed by the laser beam scanning the desired areas in adjacent hatched single tracks. In the state-of-the-art, single-mode beam sources with Gaussian beam profiles are used (Lee et al., 2017). Typically, beam diameters in the range of 30–115 μm are applied (Khorasani et al., 2020). Standard processes are limited in productivity and reproducibility due to different physical effects in the melt pool. The limitations appear in the form of process instabilities, such as keyhole collapses (Huang et al., 2022), spatter formation (Grünewald et al., 2021), and balling phenomena (Lindström et al., 2023). Such phenomena are mainly driven by existing recoil pressure and Marangoni flows, as well

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as the melt pool size and wetting behavior between the melt pool and substrate (Huang et al., 2022). Most of these physical effects are determined by the temperature field during processing. The temperature field results from the material properties and the energy input. Since the material properties of established materials can only be changed to a limited extent, a current research trend is to adapt the energy input by applying alternative beam shapes and thus influence the temperature field and the process behavior.

Several studies show the potential of ring-shaped beam profiles to enlarge the process window during the processing of AlSi7Mg0.6 (Galbusera et al., 2023), AlSi10Mg (Wischeropp et al., 2020), AlMg4.5Mn (Nahr et al., 2023), CoCrMo (Metel et al., 2019; Okunkova et al., 2016), pure zinc (Baldi et al., 2024), and 316L (Grünewald et al., 2021). In addition, the cross-sectional shape of the solidified melt pools is changed. Regardless of the material, the melt pools become wider when using ring-shaped beam profiles. Wischeropp et al. (2020) report an almost complete suppression of keyhole porosity during PBF-LB/M of AlSi10Mg with ring-shaped beam profiles. In addition, Baldi et al. (2024) show a reduced tendency to keyholing when using ring-shaped beam profiles to process pure zinc. Furthermore, alternative beam profiles have the potential to influence the amount (Grünewald et al., 2023; Rothfelder et al., 2022), size (Grünewald et al., 2023), and velocity (Tumkur et al., 2021) of resulting spatter. In addition to process optimization, such as spatter reduction or increased productivity, alternative beam profiles offer new processing possibilities, such as tailoring the microstructure (Galbusera et al., 2024). Mohebbi and Ploshikhin (2023) demonstrated a significant grain refinement during the processing of Scalmlloy® with a ring-shaped beam profile. Galbusera et al. (2024) prove how changing the beam shape influences the melt pool cross-section and affects the grain size and the texture of the microstructure.

The state of research shows that the melt pool shape, size, and dynamics can be influenced by beam shaping and optimized according to a desired parameter. The first step in experimentally evaluating the existing potential of a beam shape is typically to determine the process window in which components with a high relative density can be manufactured. With PBF-LB/M, it is common to create process maps in which the laser power and scanning speed are varied at a constant powder layer thickness and hatch distance. The design of experiments is often fully factorial. Each additional degree of freedom adds another dimension to the experimental matrix. This leads to an almost unmanageable number of experiments and process maps using different beam shapes and sizes. To start experimental screening with a small number of dimensions, it is common practice in additive manufacturing to begin with single-track experiments and determine melt track depth and melting mode. Examining single tracks is a fast and simplified way of obtaining a rough overview of the process result with varying process parameters (Wang et al., 2023). Although this simplifies the complexity of accumulated heat of adjacent scan tracks or multiple melting and solidification phenomena (Wang et al., 2023), these examinations are an excellent starting point for process investigations.

The melting modes are defined as follows.

1. Conduction mode: In conduction mode, a phase transition of the material occurs due to the temperature increase – a melt pool forms in the process zone. The heat transfer from the process zone into the solid substrate is predominantly via heat conduction. The resulting cross-sections of the solidified melt pool are approximately semi-circular (see Fig. 1a).
2. Keyhole mode: At high intensities, the material evaporates, forming a vapor capillary within the melt pool due to the recoil pressure (King et al., 2015). Consequently, the resulting penetration is significantly deeper than in conduction mode. The solidified melt pools in keyhole mode have an aspect ratio (melt track depth to width) above 0.8 (see Fig. 1c) (Tenbrock et al., 2020). During processing in keyhole mode, there is a risk of trapped gas pores, which can be caused by collapsing

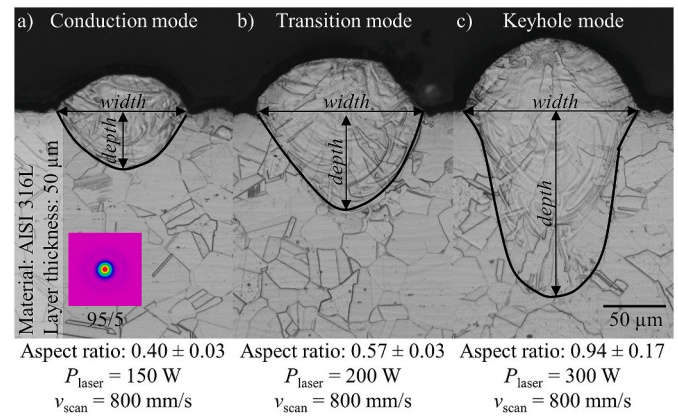


Fig. 1. Cross-section of single tracks resulting in different melting modes a) conduction mode, b) transition mode, and c) keyhole mode.

keyholes (Huang et al., 2022). Therefore, the keyhole melting mode is avoided with PBF-LB/M.

3. Transition mode: Even before the aspect ratio threshold of 0.8 is reached, evaporation of the material and convection effects driven by melt flows play a decisive role in laser material processing (Fabbro, 2020). The transition mode is defined for aspect ratios between 0.5 and 0.8 (Patel and Vlasea, 2020). An example cross-section is shown in Fig. 1b).

Studies published on ring-shaped beam profiles show that keyhole porosity is significantly reduced or avoided with alternative beam profiles (Galbusera et al., 2023; Wischeropp et al., 2020). However, there is no generally valid approach for determining a process parameter threshold for ring-shaped beam profiles above which keyholing occurs. Therefore, this work aims to provide a dimensionless parameter to estimate in which melting mode a beam shape with a known parameter set will result. To achieve this objective, a new database of melt track depths and widths produced with different ring-shaped beam profiles in the material AISI 316L is presented. The raw data are available as supplementary material. The data set is compared with the results of an analytical dimensionless heat conduction model. Since the spot size is a decisive factor for normalizing the heat conduction model, three established spot size determination methods are benchmarked against each other. Furthermore, the dimensionless enthalpy approach of Hann et al. (2011) and Rubenchik et al. (2018) is extended by a (peak-to-mean intensity ratio) factor derived from the laser beam shape, thus enabling an estimation of the melting mode.

2. State of the art

This chapter presents the state of the art for dimensionless process parameters and a dimensionless approach for an analytical heat conduction model. Due to the tremendous influence of spot size on the parameters presented, various established methods for determining spot sizes are described.

2.1. Dimensionless process parameters

To summarize process parameters during laser material processing, energy densities are typically used to estimate the process behavior. These relate the input power P_{laser} to the scanning speed v_{scan} and, if applicable, the hatch distance and layer thickness. This way, the established line, area, or volume energy densities are calculated. With PBF-LB/M, limits of the volume energy density are often specified for components with a high relative density. At volume energy densities above the specified range, gas pores occur due to unstable keyholing (Huang et al., 2022). Below this range, insufficient melting leads to lack

of fusion defects in the component (Cacace and Semeraro, 2022). However, additional parameters, such as the laser spot size, are decisive for estimating the melting mode. In addition, the material strongly influences the detection of threshold values due to its absorption behavior and thermophysical properties.

In order to make process parameters transferable, the relevant thermophysical material properties and laser beam properties, e.g., the spot size, must be included in the corresponding characteristic parameters. Various models have been developed to consider the influence of material and spot size on melting modes. In 1992, Ion et al. (1992) introduced dimensionless power and dimensionless traverse speed as parameters to classify laser material processes. The parameters include the thermal diffusivity and thermal conductivity of the processed material and the laser spot size. From the dimensionless power and the dimensionless traverse speed, dimensionless process maps are drawn in which the process behavior (heating, melting, vaporization) can be predicted based on an analytical estimate of the peak temperature occurring.

In 2011, Hann et al. (2011) extended the approach of Ion et al. (1992) by introducing the dimensionless enthalpy $\Delta H/h_s$ for predicting the weld geometry during laser welding. The dimensionless enthalpy estimates the enthalpy introduced and relates it to the enthalpy at melting h_s . The model is based on heat conduction within a semi-infinite solid with a moving Gaussian heat source. There are various definitions of dimensionless enthalpy. In this study, the definition according to (Coen et al., 2022) is used, which includes the latent heat of fusion in calculating the dimensionless enthalpy. In this way, dimensionless enthalpy can be written as

$$\frac{\Delta H}{h_s} = \frac{A \cdot P_{\text{laser}}}{\rho \cdot (c_p \cdot (T_m - T_0) + L_f) \cdot \sqrt{\pi \cdot \alpha \cdot v_{\text{scan}} \cdot (2w_0)^3}} \quad 1$$

with absorption A , laser power P_{laser} , density ρ , specific heat capacity c_p , temperature delta of the melting temperature to the ambient temperature ($T_m - T_0$), latent heat of fusion L_f thermal diffusivity α , scanning speed v_{scan} , and spot radius w_0 .

Several scaling laws were published based on the dimensionless enthalpy to predict the melt track depth and melting mode. King et al. (2014) showed that the dimensionless enthalpy can be applied to estimate the melting mode in PBF-LB/M for different-sized Gaussian beams. Gan et al. (2021) used a slightly modified form (factor $1/\sqrt{\pi}$) of dimensionless enthalpy as the “keyhole number” and showed a linear relationship with the keyhole aspect ratio. Rubenchik et al. (2018) showed the melt track depth normalized to the thermal diffusion length as a function of the dimensionless enthalpy to collapse on a well-defined curve. Most of the scaling laws use constant absorption values. This neglects the presence or shape of existing keyholes, which may influence the absorption behavior (Wittemer et al., 2023). Therefore, Ye et al. (2019) introduced a non-constant absorption determined by calorimetric laser absorption measurements using a Gaussian laser beam. However, this increases the complexity of the model.

2.2. Analytical heat conduction model

In addition to the finding that the dimensionless penetration depth as a function of the dimensionless enthalpy collapses into a well-defined curve for differing process environments and materials, Rubenchik et al. (2018) introduce an approach for calculating a dimensionless temperature field g . The approach is based on the Eagar-Tsai model (Eagar and Tsai, 1983) using a Gaussian heat source. To obtain a dimensionless solution for the temperature field, the variables are normalized. In this work, the spatial coordinates in the working plane x and y are normalized to the spot size w_0 , while the perpendicular z -coordinate is normalized to the thermal diffusion length δ_{diff} , which is calculated according to

$$\delta_{\text{diff}} = \sqrt{\frac{\alpha \cdot 2w_0}{v_{\text{scan}}}}, \quad 2$$

resulting in the dimensionless coordinates $x^* = x/w_0$, $y^* = y/w_0$, and $z^* = z/\delta_{\text{diff}}$. The velocity-dependent aspects are considered by the dimensionless Peclet number, which is calculated according to

$$Pe = \frac{2w_0 \cdot v_{\text{scan}}}{\alpha}. \quad 3$$

The dimensionless temperature field can be calculated as

$$g = \int_0^\infty \frac{\exp\left(-\frac{z^{*2}}{4t} - \frac{y^{*2} + (x^* - t)^2}{\frac{4\sqrt{2}t}{Pe} + 1}\right)}{\left(\frac{4\sqrt{2}t}{Pe} + 1\right)\sqrt{t}} dt. \quad 4$$

The factor $\sqrt{2}$ results from a definition of the spot size differing from Rubenchik et al. (2018) (see Section 2.2). The melt pool boundaries occur for

$$g = \frac{\pi}{\sqrt[3]{2^3 \cdot \frac{\Delta H}{h_s}}}. \quad 5$$

Applying this approach to Gaussian beam profiles, the normalized structure of a melt pool can be described by the Peclet number and the dimensionless enthalpy (Rubenchik et al., 2018). Due to the heat conduction character of the model, it is only valid if the process is predominantly driven by heat conduction and melt pool dynamics and the associated recoil pressure and convection are only of minor relevance. Since the aim is to find the boundary of the conduction mode melting, this limitation of the model has no drawbacks.

2.3. Spot size determination

The spot size is essential for normalizing the model and calculating the dimensionless parameters. However, most beam shapes do not have sharp edges due to diffraction phenomena, meaning the definition and the corresponding determination of the beam size is not trivial. Different methods have been established to determine the beam size without sharp edges. The simplest way of beam size determination works with threshold values. For this purpose, two diametrically opposed points of a beam profile, whose intensity is a specific part of the peak intensity of the beam, are chosen within the intensity distribution. The distance between the two points is defined as the size of the beam. One possible choice for the ratio of the threshold value and the peak value is 0.5. The obtained diameter is called the Full Width at Half Maximum (FWHM). Another possible choice of the fraction is $1/e$ ($\approx 36.8\%$) or $1/e^2$ ($\approx 13.5\%$). These values are essential for the analytical description of Gaussian beam profiles, where the intensity profile is described by

$$I(r) = \frac{2P_{\text{laser}}}{\pi w_0^2} e^{-2\frac{r^2}{w_0^2}}. \quad 6$$

In the case of the straight threshold value approach, the beam diameter depends only on 3 points. In order to take the spatial intensity distribution into account, the raw data are often approximated by a Gaussian curve whose $1/e$ or $1/e^2$ values are specified as radii. If the intensity distribution deviates significantly from a Gaussian distribution, the error of the Gaussian fit becomes large.

Some methods do not require a Gaussian distribution to prevent large fit errors. An alternative beam size determination method is the 86 % method. Here, the diameter of the beam profile, which contains 86 % of the total power, is given. The solution must be found numerically.

An analytical alternative is offered by the 2nd moment method standardized according to ISO 11146. The resulting diameter is sometimes referred to as $D4\sigma$. The diameter is given by four times the

standard deviation sigma of the intensity distribution. In this paper, the common spot radius w_0 is used. The radius along the x-axis is calculated according to

$$w_x = 2 \sqrt{\frac{\iint x^2 I(x, y) dx dy}{\iint I(x, y) dx dy}} \quad 7$$

For a Gaussian beam profile, the $1/e^2$ diameter of a Gaussian fit, the 86 % diameter, and the 2nd moment diameter give the same value. The resulting beam diameters differ if the beam profile deviates from the Gaussian distribution (see Fig. 4).

3. Materials and methods

This section describes the methods and materials used to generate experimental samples.

3.1. Experimental setup

All measurements and powder bed fusion experiments are conducted on an in-house developed test bench. A fiber laser with switchable beam shapes (AFX-1000, nLIGHT Inc., Vancouver, WA, USA) is used as the beam source. The beam shapes are composed of a central core and a surrounding ring, in between which the laser power can be distributed in seven discrete steps. The name of the beam profiles in this publication is based on the approximate power distribution between the Gaussian core and the surrounding ring. For beam deflection, focusing, and zooming, a four-axis beam deflection unit (AM-MODUL NEXT GEN, RAYLASE GmbH, Wessling, Germany) is used. Optical output power of up to 1200 W is available. Table 1 summarizes the spot sizes for different magnification levels determined using a Gaussian fit, the 86 % method, and the 2nd moment method. Additionally, a peak-to-mean intensity ratio ψ is specified for the 2nd moment method. This is the ratio of the maximum measured peak and the mean intensity (laser power P_{laser} per cross-sectional spot area πw_0^2) according to

$$\psi = \frac{I_{\text{peak}}}{I_{\text{mean}}} = \frac{I_{\text{peak}} \cdot \pi w_0^2}{P_{\text{laser}}} \quad 8$$

The peak-to-mean ratio values ψ given in Table 1 are the mean values of all three measured magnifications since the beam profile does not change qualitatively by zooming. Fig. 2 shows an intensity cross-section of the beam profile 50/50 with the resulting peak and mean intensity to visualize the beam profile dependence of the peak-to-mean ratio. All optical measurements were obtained with a camera-based beam profiler (FBP-2KF, CINOGY Technologies GmbH, Duderstadt, Germany).

The powder bed fusion experiments take place in a self-built miniaturized process chamber. During the experiments, a maximum oxygen content of 0.2 % and a protective gas flow of 16 l/min across the substrate plates are ensured in the chamber.

3.2. Design and evaluation of experiments

The investigations are conducted based on single-track experiments. An exemplary build job is shown in Fig. 3. Exposure takes place opposite to the shielding gas flow. Laser-cut substrates made of cold-rolled 316L sheet metal are used as the base material. A 50 μm powder layer is applied using a blade and gauge tape. The layer thickness is validated using a fringe projection measurement. The powder material used is MetcoAdd 316L-A (Oerlikon Metco Europe GmbH, Raunheim, Germany). The powder diameters are specified as $d_{10,3} = 19 \mu\text{m}$, $d_{50,3} = 30 \mu\text{m}$, and $d_{90,3} = 46 \mu\text{m}$. The substrate plates are sandblasted to increase the surface roughness ($Sa \approx 3 \mu\text{m}$, $Sz \approx 42 \mu\text{m}$). The process parameters laser power and scanning speed were varied to investigate a wide range of dimensionless enthalpy for the beam profiles 95/5, 20/80, and 10/90. For the beam profiles 75/25, 65/35, 50/50, and 35/65, the scanning speed was kept constant at 800 mm/s while the laser power varied. The tested parameter sets are shown in Table 2.

The single tracks produced are cut in the middle and embedded in an epoxy resin. Subsequently, the samples are ground with silicon carbide paper in four stages (#320, #800, #1200, #2500) and polished with lubricant and polishing suspension in two stages (3 μm and 1 μm). Before microscopy, the samples are etched with an etchant according to Kalling 2 to make the solidified melt pools visible. Three single tracks are created for each parameter set, from which the mean value and the standard deviation are calculated for the evaluation. The calculated and measured raw data are available in the supplementary material. For calculating the dimensionless parameter, the thermophysical properties of AISI 316L are taken from (Mills, 2002) following Coen et al. (2022) as listed in Table 3. According to King et al. (2015), the absorption A is assumed to be constant at 0.6. This neglects the influence of the melting mode on the absorption coefficient while keeping the model simpler.

Table 1
Evaluated spot size of the used beam profiles.

Power distribution core/ring	Beam size determination method	Spot diameter $2w_0$ in μm Magnification M			Peak-to- mean ratio ψ
		1	1.5	2	
 95/5	Gaussian Fit	72	103	134	3.19
	86 % Method	85	123	161	
	2 nd Moment Method	92	133	175	
 75/25	Gaussian Fit	72	103	134	4.89
	86 % Method	143	212	281	
	2 nd Moment Method	125	183	241	
 65/35	Gaussian Fit	75	106	138	5.11
	86 % Method	154	226	299	
	2 nd Moment Method	142	206	271	
 50/50	Gaussian Fit	81	113	144	5.02
	86 % Method	162	238	314	
	2 nd Moment Method	163	238	313	
 35/65	Gaussian Fit	227	335	444	2.95
	86 % Method	168	246	325	
	2 nd Moment Method	189	276	363	
 20/80	Gaussian Fit	247	365	484	1.95
	86 % Method	169	248	326	
	2 nd Moment Method	197	287	378	
 10/90	Gaussian Fit	266	395	524	2.32
	86 % Method	171	250	330	
	2 nd Moment Method	205	300	394	

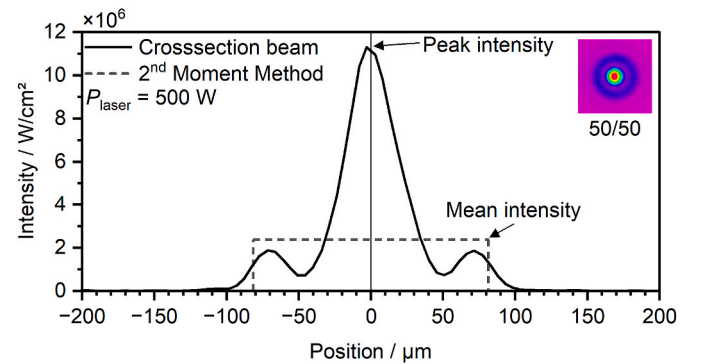


Fig. 2. Exemplary comparison of the measured peak and mean intensity for beam profile 50/50.

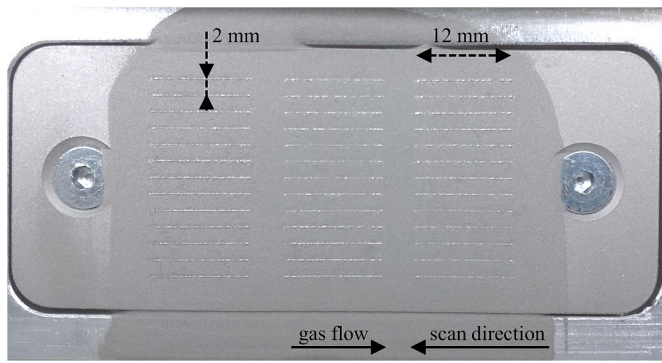


Fig. 3. Photo of an executed build job.

4. Results and discussion

First, the different methods of spot size determination are compared and briefly discussed. Next, the experimental single-track results are compared with the heat conduction model using different spot size determination methods. Subsequently, the peak-to-mean intensity ratio is used to introduce a maximum dimensionless enthalpy. Finally, the aspect ratio is correlated with the maximum dimensionless enthalpy. Consequently, the melting mode is detected based on the maximum dimensionless enthalpy.

4.1. Spot size determination

Fig. 4 shows four exemplary beam shapes used in this study and the corresponding beam sizes determined using a Gaussian fit with $1/e^2$ diameter, the 86 % method, and the 2nd moment method. For shapes that a Gaussian function can well approximate, the values of the beam radii are similar (see Fig. 4a). Beam profiles that deviate strongly from the Gaussian shape (see Fig. 4b, c, and d) show enormously different beam diameters depending on the method. Fig. 4b and c shows that the Gaussian fit is very much oriented to the central intensity peak. As a result, the power in the ring is less considered. Consequently, the

Gaussian fit becomes much smaller than the 86 % or the 2nd moment method. If the power fraction in the ring is increased, these conditions are inverted. Fig. 4d shows that a high power fraction in the ring causes the Gaussian fit to result in a substantially larger beam diameter. The high power fraction in the ring also leads to a more uniform illumination of the beam cross-sectional area, so the 86 % method results in the smallest beam diameter. The 2nd moment diameter is between the ones determined with Gaussian fit and 86 % method. The following section benchmarks the various determination methods based on the deviations between the experimental and modeled melt track depth and width data. In this way, the most suitable spot size determination method is identified.

4.2. Correlation between an analytical heat conduction model and experimental data

Figs. 5, 6, and 7 show the melt track depths normalized to the thermal diffusion depth in a and the melt track width normalized by the spot diameter in b as a function of the dimensionless enthalpy. The figures differ from each other in the method of spot size determination. The spot size is part of the calculation of the dimensionless enthalpy and of the normalization of the melt track depth and the melt track width. The experimental results of the different beam profiles are identified within the diagrams by a color code. In addition to the experimental data, the calculated melt pool depth and width from the heat conduction model are included according to equations (4) and (5). The melt track depths and widths depend besides the dimensionless enthalpy on the Peclet number. Accordingly, the representation with a 3D plot could be used. As the influence of the Peclet number in the investigated value range ($Pe > 9$) is small compared to the dimensionless enthalpy, three exemplary Peclet numbers are shown as line graphs for a clearer presentation. Pe_{min} (black dotted line) and Pe_{max} (black dashed line) are the smallest and largest Peclet numbers used in the data set. Pe_{mean} (black solid line) is the average Peclet number of all data points. In order to evaluate the fit of experimental data and the heat conduction model, Table 4 shows the deviations between model and experimental values resulting in conduction mode (squares in Figs. 5–7) as the square root of

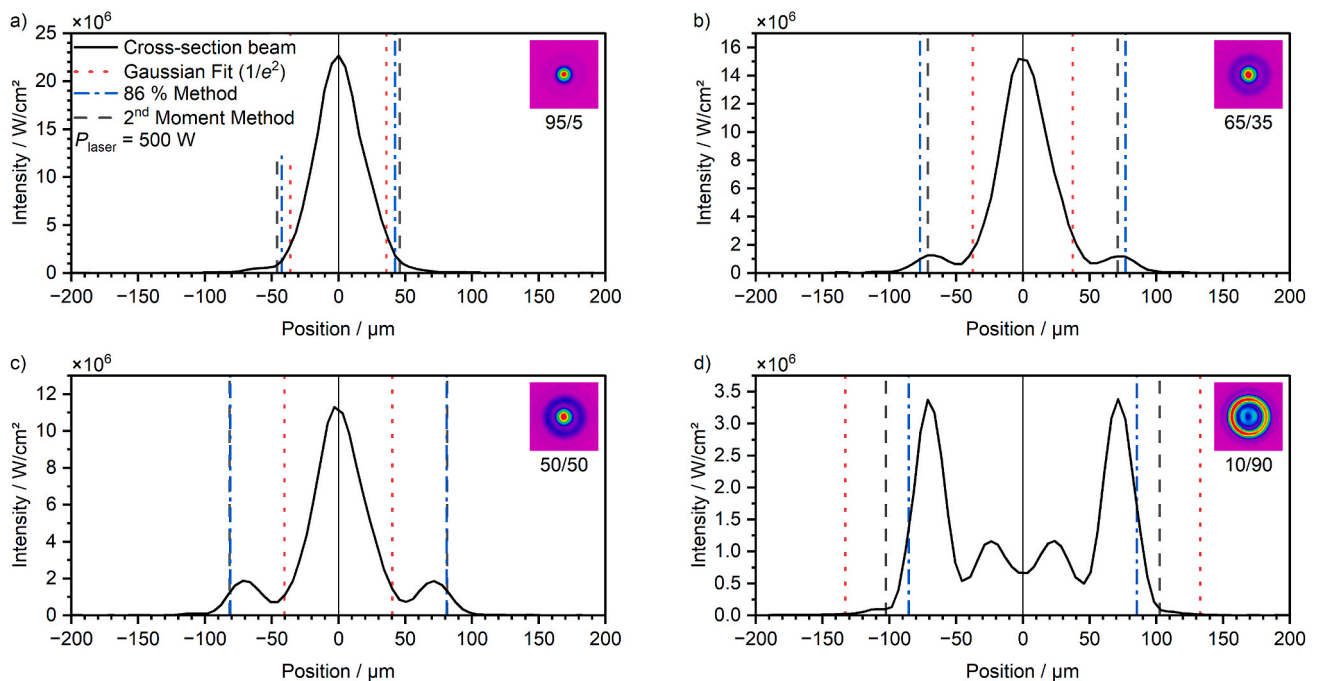


Fig. 4. Measured cross-section and determined beam size using a Gaussian fit, the 86 % method, and the 2nd moment method of the four beam profiles a) 95/5, b) 65/35, c) 50/50, and d) 10/90.

Table 2

Process parameters used for the proof of concept build jobs.

Power distribution core/ring	Magnification	Laser power P_{laser} in W	Scanning speed v_{scan} in mm/s
 95/5	1	100, 150, 200, 250, 300, 400, 500, 600	800
	1	400, 500, 600	1100
	1	400, 500, 600	1400
	1	400, 500, 600	1700
	1.5	200, 250, 300, 350	800
	2	200, 300, 400, 500, 600	800
	2	400, 500, 600	1100
	2	400, 500, 600	1400
 75/25	1	100, 150, 200, 250, 300	800
	1.5	200, 250, 300, 350	800
	2	200, 300, 400, 500	800
 65/35	1	100, 150, 200, 250, 300	800
	1.5	200, 250, 300, 350	800
	2	200, 300, 400, 500	800
 50/50	1	150, 200, 250, 300	800
	1.5	250, 300, 350, 400, 450	800
	2	400, 500, 600, 700	800
 35/65	1	200, 300, 400, 500	800
	1.5	400, 500, 600, 700	800
	2	600, 750, 900, 1050, 1200	800
 20/80	1	300, 400, 500, 600, 750	800
	1	750, 900, 1050, 1200	1100
	1	750, 900, 1050, 1200	1400
	1	750, 900, 1050, 1200	1700
	1.5	400, 500, 600, 700, 800	800
	2	750, 900, 1050, 1200	800
	2	750, 900, 1050, 1200	1100
	2	750, 900, 1050, 1200	1400
	2	750, 900, 1050, 1200	1700
	1	300, 400, 500, 600, 750, 900, 1050, 1200	800
	1	400, 500, 600, 750, 900, 1050, 1200	1100
	1	400, 500, 600, 750, 900, 1050, 1200	1400
	1	400, 500, 600, 750, 900, 1050, 1200	1700
	1.5	400, 500, 600, 700, 800	800
	2	750, 900, 1050, 1200	800
	2	750, 900, 1050, 1200	1100
	2	750, 900, 1050, 1200	1400
	2	750, 900, 1050, 1200	1700
 10/90	1	400, 500, 600, 750, 900, 1050, 1200	1100
	1	400, 500, 600, 750, 900, 1050, 1200	1400
	1	400, 500, 600, 750, 900, 1050, 1200	1700
	1.5	400, 500, 600, 700, 800	800
	2	750, 900, 1050, 1200	800
	2	750, 900, 1050, 1200	1100
	2	750, 900, 1050, 1200	1400
	2	750, 900, 1050, 1200	1700

Table 3

Thermophysical properties of AISI 316L according to (Mills, 2002).

Name	Value	
Ambient temperature T_0	293.15	K
Melting temperature T_m	1691	K
Density ρ	7269	kg/m ³
Specific heat capacity c_p at T_m	700	J/(kg•K)
Thermal diffusivity α at T_m	$6.11 \cdot 10^{-6}$	m ² /s
Latent heat of fusion L_f	260	kJ/kg

the mean square error (RSME). Only the data points resulting in the conduction mode are used for the comparison, as the model neglects the physical effects influencing the heat flow in transition and keyhole mode.

Using the Gaussian fit firmly separates the measured depths of the different beam profiles (see Fig. 5a). The beam profiles up to a core/ring

Table 4

Root mean square error (RMSE) between the heat conduction model with different spot sizing methods and the experimental data resulting in conduction mode (aspect ratio <0.5). The lowest values are marked in bold.

Power distribution core/ring	RMSE	Gaussian fit	86 % method	2 nd moment method
 95/5	Depth	0.611	0.350	0.253
	Width	0.165	0.172	0.177
 75/25	Depth	0.650	0.476	0.300
	Width	0.115	0.184	0.181
 65/35	Depth	1.049	0.238	0.112
	Width	0.183	0.212	0.185
 50/50	Depth	1.326	0.160	0.154
	Width	0.284	0.098	0.096
 35/65	Depth	0.190	0.426	0.221
	Width	0.157	0.112	0.104
 20/80	Depth	0.340	0.624	0.394
	Width	0.173	0.201	0.160
 10/90	Depth	0.377	0.801	0.492
	Width	0.198	0.281	0.211
Overall	Depth	0.623	0.593	0.369
	Width	0.189	0.215	0.176

power distribution of 50/50 are significantly less deep in the experiments than in the model. Above a dimensionless enthalpy of 2, the use of beam profiles 35/65, 20/80, and 10/90 results in slightly deeper melt tracks than calculated. Below the dimensionless enthalpy of 2, the calculated and experimental values of the beam profiles 35/65, 20/80, and 10/90 are very close. In line with this, the melt pools for the beam profiles up to 50/50 are significantly wider in the experiments than in the model. For the beam profiles above 50/50, the model and the experimental data match for a dimensionless enthalpy of less than 2. Above a dimensionless enthalpy of 2, the experimentally generated melt pools are narrower than calculated in the model. One reason for this is that the fit for the beam profiles up to 50/50 is strongly oriented towards the central peak and, therefore, determines the spot sizes to be comparatively small (see Table 1). From beam profile 35/65 onwards, the high intensity in the surrounding ring leads to the spot sizes being determined abruptly larger (see Table 1). As a result, these two groups of beam profiles deviate from the model in different directions. Physically, it can be deduced that the changed intensity distribution of the beam profiles up to 50/50 influences the melt track depth more than the model estimated. The main reason for this is that the intensity of the central Gaussian peak is assumed to be too high. Due to the too-small determined beam diameters of the beam profiles up to 50/50 and the too-large determined beam profiles from 35/65, the spot size and, thus, the determination method does not fit the model, which leads to the highest RMSE value of 0.623 for the depth and the second highest RMSE for the width of 0.189 in the entirety of the data points (see Table 4).

When using the 86 % method to determine the spot sizes, it is remarkable that the graphical distribution of the raw data and fits is reversed to a certain extent compared to the Gaussian fit (see Figs. 5 and 6). The melt track depths generated with the beam profiles with a power distribution of up to 50/50 are slightly deeper and narrower than the

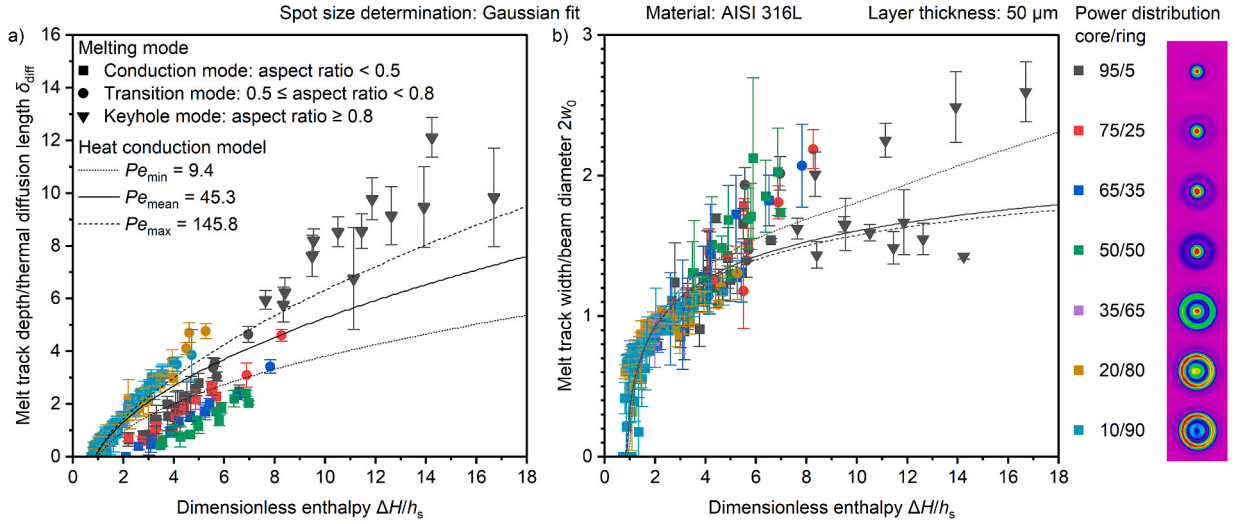


Fig. 5. a) Melt track depth related to the thermal diffusion length, and b) melt track width normalized to the spot diameter, both as a function of the dimensionless enthalpy, spot size determined using a Gaussian fit. The line graphs show the predicted depth and width values from an analytical heat conduction model. Pe_{min} (black dotted line), Pe_{mean} (black solid line), and Pe_{max} (black dashed line) represent the smallest, mean, and largest Peclet numbers used in the shown data set.

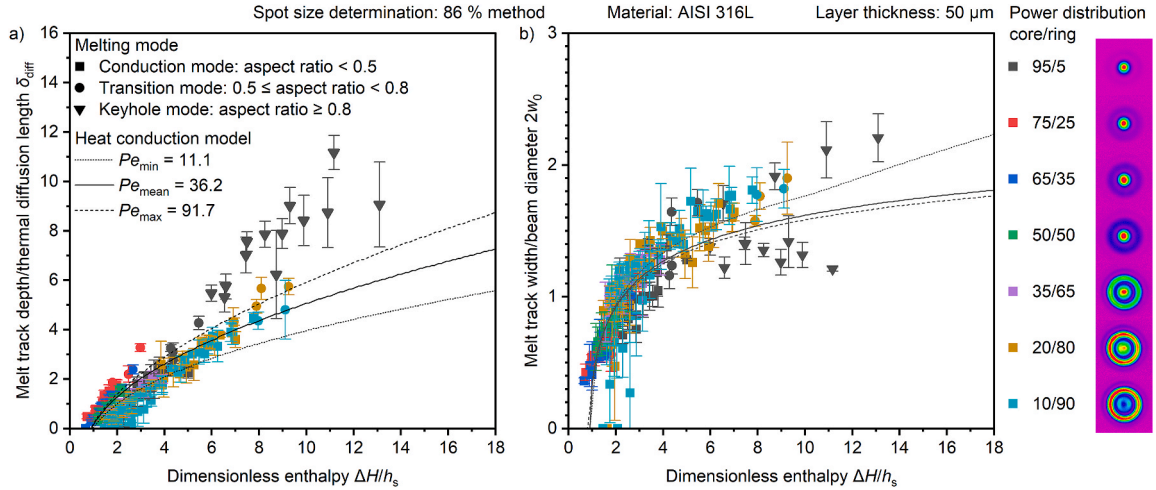


Fig. 6. a) Melt track depth related to the thermal diffusion length, and b) melt track width normalized to the spot diameter, both as a function of the dimensionless enthalpy, spot size determined using 86 % method. The line graphs show the predicted depth and width values from an analytical heat conduction model. Pe_{min} (black dotted line), Pe_{mean} (black solid line), and Pe_{max} (black dashed line) represent the smallest, mean, and largest Peclet numbers used in the shown data set.

data from the model. Using beam profiles from 35/65 results in shallower melt pools that tend to be wider compared to the model (see Fig. 6a and b). Physically, this suggests that the beam profiles up to 50/50 are determined as slightly too large and from 35/65 slightly too small to fit the model. Overall, the point clouds of the different beam profiles overlap more for the normalized melt track depth (see Fig. 6a), which results in a slightly reduced RMSE of the depth values of 0.593. Regarding the melt track width, the model with the 86 % spot size and the measured data exhibits a larger deviation, expressed by the increased RMSE of the width values of 0.215.

One noticeable point in the data is the change in the relative position of the measurement points for the 50/50 beam profile and the corresponding matching with the model of this data series in a direct comparison between the Gaussian fit and the 86 % method. When using a Gaussian fit to determine the spot size, the data points of this measurement series are located at the outer edge of the entirety of the data points and have the lowest melt track depth normalized to the thermal diffusion depth at constant dimensionless enthalpies. Therefore, the difference to the model is the highest ($\text{RMSE}_{\text{depth, Gaussian fit}} = 1.326$, $\text{RMSE}_{\text{width, Gaussian fit}} = 0.284$). When using the 86 % method, the data

points move towards the center of the total data points. This makes the error to the model quite small ($\text{RMSE}_{\text{depth, 86 \%}} = 0.160$, $\text{RMSE}_{\text{width, 86 \%}} = 0.098$). The main reason for this behavior with beam profile 50/50 is the intensity distribution between the central peak and the surrounding ring. The intensity in the ring is still low, so the Gaussian fit is strongly oriented towards the central peak (see Fig. 4c). This shows that the impact of the ring on the process result is underestimated, especially for beam profiles with low intensities in the ring. The determined spot size in the Gaussian fit increases only insignificantly between the beam profiles 95/5 and 50/50 compared to the change in the process results (see Table 1). Consequently, this method is unsuitable for determining appropriate beam diameters for the heat conduction model.

Fig. 7 shows the melt track depth related to the thermal diffusion length and the melting track width normalized to the spot diameter as a function of the dimensionless enthalpy with the spot size determined using the 2nd moment method. The spread of the data points of the 2nd moment method and their deviation from the model are significantly lower for both the normalized depth and width compared to the Gaussian fit and the 86 % method, which is confirmed by the lowest overall error of the entire data set ($\text{RMSE}_{\text{depth}} = 0.369$, $\text{RMSE}_{\text{width}} =$

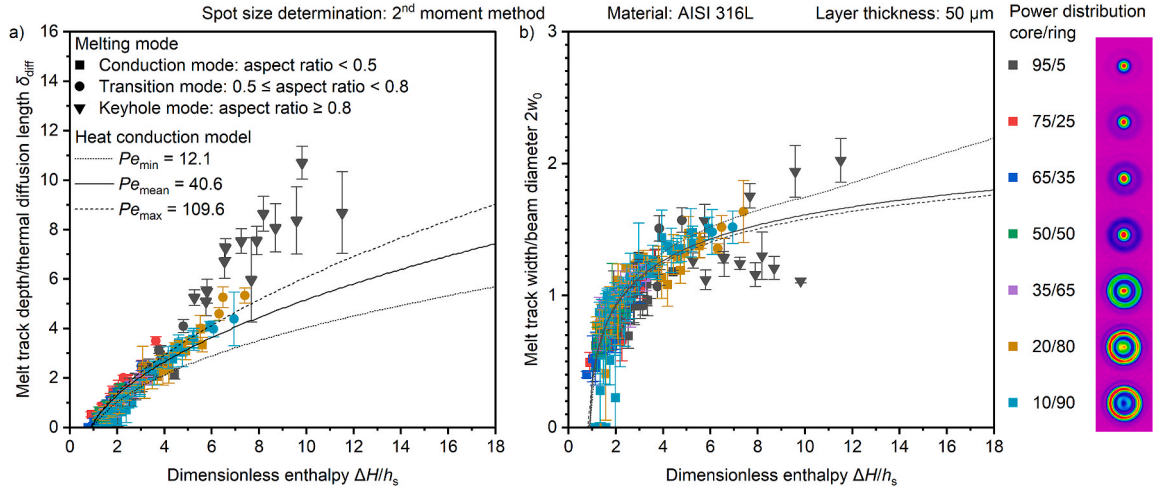


Fig. 7. a) Melt track depth related to the thermal diffusion length, and b) melt track width normalized to the spot diameter, both as a function of the dimensionless enthalpy, spot size determined using 2nd moment method. The line graphs show the predicted depth and width values from an analytical heat conduction model. Pe_{min} (black dotted line), Pe_{mean} (black solid line), and Pe_{max} (black dashed line) represent the smallest, mean, and largest Peclét numbers used in the shown data set.

0.176). Consequently, the 2nd moment method is the most suitable spot size determination method for estimating the melt track depth and width for a specific set of process parameters based on the dimensionless enthalpy and the Peclét number. The qualitative progression of the raw data is similar to the 86 % method. Using beam profiles up to 50/50 tends to result in deeper and narrower melt tracks than using beam profiles from 35/65. The comparison of the experimental data and the heat conduction model (with the 2nd moment method for spot size determination) in Fig. 7 clearly shows the limits of the heat conduction model. The experimental data assigned to the transition mode (aspect of the melt pool cross-section between 0.5 and 0.8) show higher normalized melting track depths than calculated by the model. In this case, additional effects such as vapor pressure and convection increase the melt pool depth. These effects have no noticeable influence on the melt track width since the lateral heat flow plays a subordinate role compared to the heat flow into the depth at the high Peclét numbers studied (Rubenchik et al., 2018). Parameter sets that result in keyholing (aspect of the melt pool cross-section higher than 0.8) result in significantly deeper melt track depths than predicted by the model due to the physical effects not considered in the model (changed absorption behavior, convection and vapor pressure).

Although the deviations between the heat conduction model and the experimental data become small when using the 2nd moment method, a systematic error is introduced by the assumptions of a Gaussian heat source in the model. The shape deviation of the intensity distribution results in the profiles with a large proportion of power in the ring (especially 20/80 and 10/90) having slightly flatter and wider melt pool cross-sections than predicted by the model. Beam profiles with a lower proportion of power in the ring (such as 75/25 and 65/35) result in slightly deeper and narrower melt pool cross-sections (see Fig. 7). In combination, the deviations lead to significant differences in the aspect ratio of the melt pool cross-sections and, thus, to different threshold values for the change in melting mode, defined by the aspect ratio of the melt pool cross-sections. While the transition mode occurs from a dimensionless enthalpy of approximately 3.8 for beam profile 95/5, this mode is already present from a dimensionless enthalpy of 2.3 for beam profile 75/25. For beam profile 20/80, the threshold for the transition mode is shifted upwards to approximately 5.5 (see Fig. 7). Fig. 8 shows examples of cross-sections created with beam profiles 95/5, 75/25, 65/36, and 20/80 with dimensionless enthalpies around 3.5. Despite similar dimensionless enthalpies, different melting modes are recognizable. Beam profile 95/5 is on the border to transition mode (see Fig. 8a). Beam profile 75/25 is already in transition mode (see Fig. 8b). Beam profiles

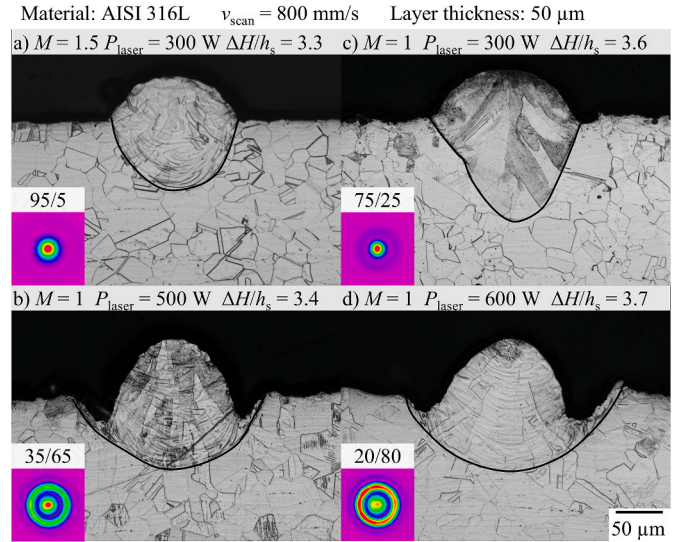


Fig. 8. Exemplary cross-sections created with similar dimensionless enthalpies.

65/35 and 20/80 are still in conduction mode (see Fig. 8c and d). Consequently, an additional indicator is required to improve the prediction of the melting mode.

4.3. Correlation between the maximum dimensionless enthalpy and the melting mode

The melting mode is assigned based on the aspect ratio of the cross-section of a single track. The aspect ratio is the melt track depth related to the width. This publication uses the definition of Patel and Vlasova (2020). Aspect ratios below 0.5 belong to the conduction mode, aspect ratios between 0.5 and 0.8 to the transition mode, and aspect ratios above 0.8 to the keyhole mode. As mentioned in the previous section, the beam profiles influence the aspect ratio and, thus, the melting mode. Beam profiles with more power in the ring tend to keyholing only at higher dimensionless enthalpies than beam profiles with a high proportion of power in the core (see Fig. 7). The main difference between the two beam profile types is the peak-to-mean intensity. The tendencies from Fig. 7 imply that deeper, narrower melt track cross-sections are created with higher peak-to-mean intensity ratios, meaning beam profiles with high peak-to-mean intensity ratios change the melting mode at

lower dimensionless enthalpies. To include this influence of the beam shape on the resulting melting mode, the dimensionless enthalpy is extended by half the peak-to-mean intensity ratio $\psi/2$. Thus, the maximum dimensionless enthalpy is calculated as

$$\frac{\Delta H_{\max}}{h_s} = \frac{\psi}{2} \frac{\Delta H}{h_s}. \quad (9)$$

By introducing the additional indicator, the beam shape is considered by adjusting the Gaussian heat source in the heat conduction model. Half the peak-to-mean intensity ratio is used, as the Gaussian source used in the model already exhibits a peak-to-mean intensity ψ_{Gauss} of 2. Fig. 9 shows how the intensity is modeled for the energy input of beam profiles 95/5, 50/50, and 10/90. The black solid line describes the measured intensity cross-section of the beam profile. The red dashed line corresponds to the Gaussian intensity distribution used in the heat conduction model for calculating the dimensionless enthalpy according to equation (3) (spot size determined using the 2nd moment method). The energy content (area below the rotational symmetric profile) corresponds to that of the measurement. As beam profile 95/5 does not correspond to a perfect Gaussian beam, but approx. 5 % of the power is emitted in the ring of the profile, the beam diameter is slightly larger than the diameter of the central Gaussian core of the profile. This results in a reduced modeled peak intensity (see Fig. 9a). The blue dotted line represents the intensity distribution for calculating the maximum dimensionless enthalpy. By considering the factor of $\psi/2$, the peak intensity used in the model is roughly adjusted to the value of the measured peak intensity. Although the energy in the model does not correspond to the measured energy due to the factor of $\psi/2$, the model extension works to separate the three melting modes.

Fig. 10 presents the aspect ratio as a function of the introduced maximum dimensionless enthalpy. Considering the peak-to-mean intensity ratio, the individual data series of the beam profiles are shifted on the abscissa until the transition areas of the melting modes move into similar areas. The crossover between the conduction and transition modes occurs in a maximum dimensionless enthalpy range from 5.4 to 7.1 (see the dashed area in Fig. 10). For keyholing, the crossover range is from 8.35 to 8.95 (see dotted area in Fig. 10). The accuracy of the melting mode estimation is in a similar range to King et al. (2014), reporting the melting mode change from conduction to keyhole for Gaussian beam profiles at a dimensionless enthalpy of 30 ± 4 when processing AISI 316L. By defining the spot size differently, King et al. (2014) had a higher threshold value of the dimensionless enthalpy for the melting mode change. Fig. 11 shows examples of four cross-sections created with similar maximum dimensionless enthalpies. Due to the different spot sizes and beam profiles, the melt track depths and widths differ significantly in absolute numbers. The shape of the top surface also differs. Fig. 11c and d shows that the melt pool begins to be depressed at the melt pool borders. These depressions are not visible in Fig. 11a and b. Nevertheless, it is clearly recognizable that all images of

Fig. 11 show cross-sections in the crossover area to the transition mode. The simplified approach of the maximum dimensionless enthalpy to estimate the melting mode does not reflect all the process characteristics of PBF-LB/M. Due to the simple analytical approach, some process-influencing variables, such as the hatch distances, layer thickness, and powder and powder bed properties, are neglected. As a result, no direct statements on dynamic process properties, such as spatter formation or melt flows, are possible from the presented dimensionless parameters. However, the demonstrated approach enables estimating the melting mode based on a one-dimensional parameter. This makes the maximum dimensionless enthalpy, at least for the type of beam profiles consisting of a central peak and a surrounding ring and the high Peclet numbers characteristic in PBF-LB/M, a suitable tool to reduce the experimental effort in parameter studies.

5. Conclusion

This paper addresses the current research trend of applying alternative beam profiles in laser-based powder bed fusion of metals. The influence of the beam profile and the process parameters on the melt track depth, width, and melting mode is investigated using a new experimental basis of single-track experiments with an applied 50 μm powder layer. Various ring-shaped beam profiles are used for the experiments during the laser-based powder bed fusion process of AISI 316L stainless steel. Several methods of spot size determination are investigated and discussed for the evaluation. The core findings can be summarized as follows.

- From the established spot size determination methods Gaussian fit, 86 % method, and 2nd moment method, the 2nd moment method is the most suitable to derive statements from the spot size on the melt pool cross-sections, independent of the beam profile used.
- An analytical heat conduction model with a moving Gaussian heat source provides results for the melt pool cross-section that are in good agreement with experimental data generated with ring-shaped beam profiles if the 2nd moment method is used for the spot size determination.
- The introduced maximum dimensionless enthalpy can estimate the aspect ratio and the melting mode for the investigated ring-shaped beam shapes. The upper limit of the conduction mode using a 50 μm powder layer is a maximum dimensionless enthalpy of 6.25 ± 0.85 . The lower limit for the keyhole mode is a maximum dimensionless enthalpy of 8.65 ± 0.30 . In between, the transition mode is present.

With the results shown, the melt track depth, width, and melting mode can be estimated using simple dimensionless parameters. Although the approach is based on an analytical heat conduction model and does not represent process dynamics, the experimental effort can be

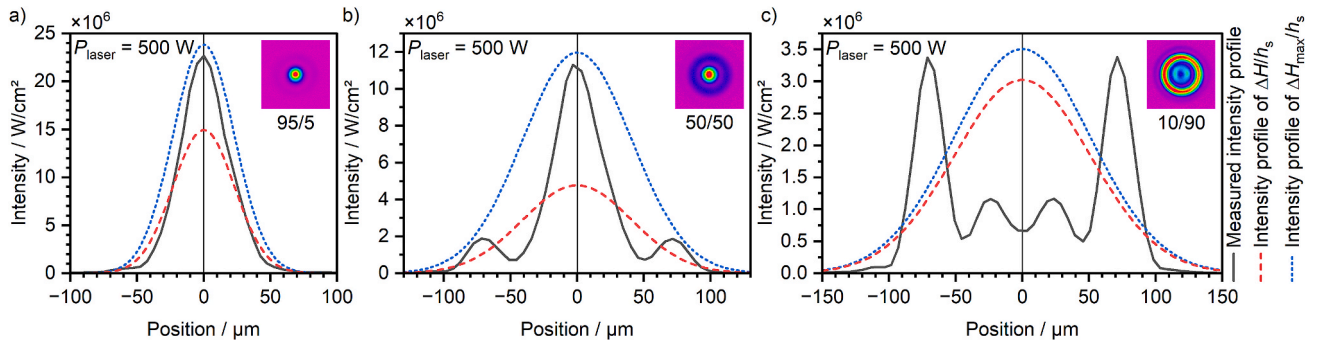


Fig. 9. Comparison of the measured beam profiles, those used for calculating the dimensionless enthalpy and those used for calculating the maximum dimensionless enthalpy a) 95/5, b) 50/50, and c) 10/90.

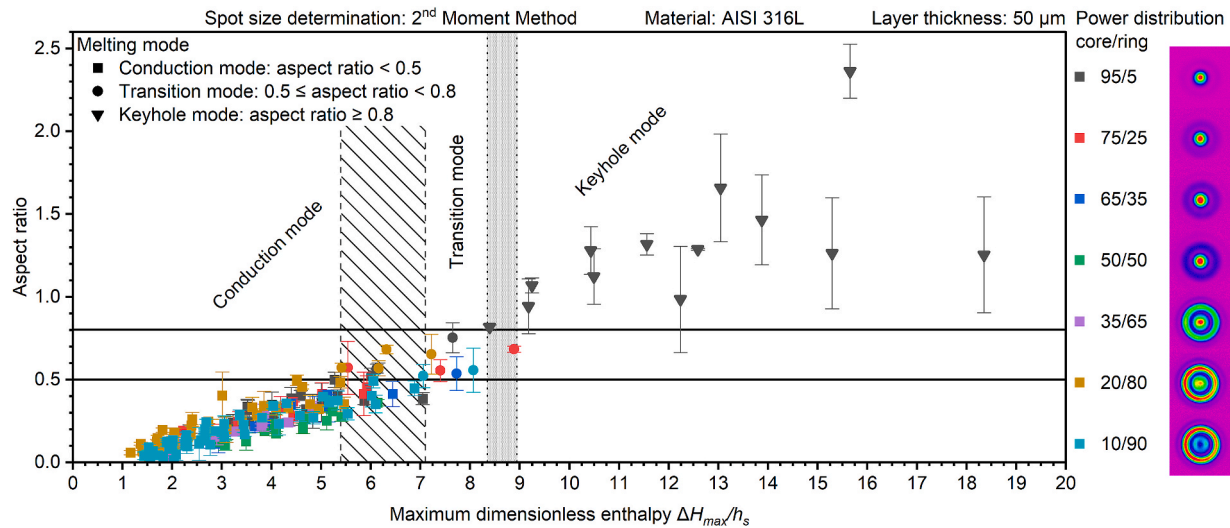


Fig. 10. Aspect ratio as a function of the product of the maximum dimensionless enthalpy, spot size determined using 2nd moment method.

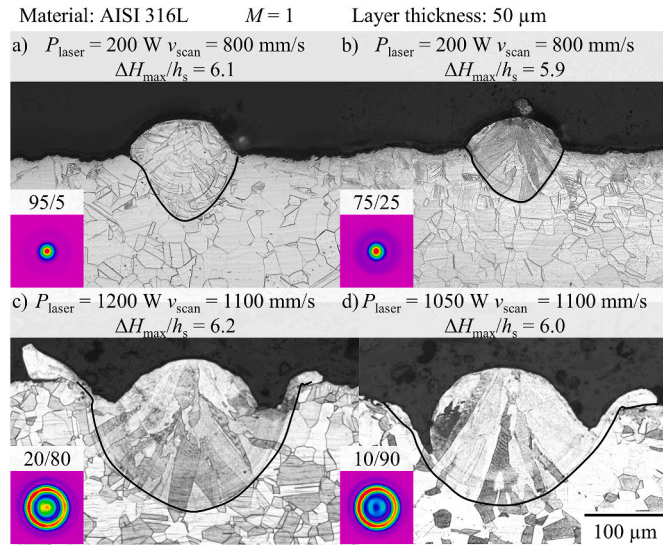


Fig. 11. Cross-sections created with similar maximum dimensionless enthalpies with beam shape a) 95/5, b) 75/25, c) 20/80, and d) 10/90.

significantly reduced. However, in the three-dimensional PBF-LB/M process, the thermal boundary conditions change compared to single-track investigations. Therefore, one next step is transferring the correlations to a three-dimensional process, considering the influence of hatch distance.

CRediT authorship contribution statement

Jonas Grünwald: Writing – review & editing, Writing – original draft, Visualization, Validation, Supervision, Software, Resources, Project administration, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **Moritz Wittmer:** Writing – review & editing, Formal analysis. **Katrin Wudy:** Writing – review & editing, Visualization, Supervision, Resources, Project administration.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Appendix A. Supplementary data

Supplementary data to this article can be found online at <https://doi.org/10.1016/j.aime.2025.100164>.

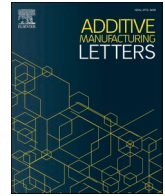
Data availability

The data used in the paper are available as supplementary material.

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Short Communication

Dimensionless process windows in laser-based powder bed fusion of AISI 316L using ring-shaped beam profiles

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ABSTRACT

The research trend to investigate the influence of alternative beam profiles on the process and component properties in laser-based powder bed fusion raises the question of how to compare the processes and process results generated with various beam profiles in different sizes. The current state of research mainly examines the process simplified on a single-track basis or addresses isolated aspects, such as the change in beam profile and size with constant absolute process parameters, which neglects the cross-effects of these parameters. Therefore, this paper presents a new approach to consider varied process parameters and their cross effects. The approach is based on a simple heat conduction model and allows the creation of beam shape and size-independent process maps. These dimensionless process maps are created by replacing the common dimensioned process parameters (laser power and scan speed) with combined dimensionless parameters (dimensionless enthalpy and Peclet number, each extended by a dimensionless hatch distance). This way, the parameters consider material and beam properties. Within the process maps, the process boundaries are predicted by simple geometric conditions of the calculated melt pools using the introduced heat conduction model. The model is experimentally validated by conducting a comprehensive parameter study using a multidimensional design of experiments with seven different beam profiles in various sizes and varying laser power, scanning speed, and hatch distance processing AISI 316L. The relative density and surface roughness are evaluated in the experiments. The predicted and experimentally determined process limits are in excellent agreement.

1. Introduction

Laser-based powder bed fusion of metals (PBF-LB/M) is an additive manufacturing process consisting of layer-wise melting metallic powder in a powder bed using a laser beam. The laser exposes the desired areas in the powder bed by scanning defined exposure patterns consisting of individual hatched scan tracks. State-of-the-art PBF-LB/M uses single-mode lasers with a Gaussian intensity distribution [1]. These typically provide spot sizes between 30 and 115 μm [1]. The resulting high intensities ensure the complete melting of the powder feedstock, resulting in dense components (relative densities greater than 99.5 %) [2].

With PBF-LB/M, the process parameters are typically optimized to result in highly dense samples with low surface roughness. Parameter combination ranges leading to the desired result span the process window. Outside the process window, various physical effects lead to undesirable process and component properties. When applying high laser energy densities, material flows are created, material vaporizes, and

vapor depressions form [3]. This can lead to unstable keyholing and keyhole porosity in the component [4]. When very low energy densities are used, the powder material is not melted sufficiently, leading to a lack of fusion [5]. If too high process speeds are selected, balling effects and bulging top surfaces can occur [6]. Above a critical scanning speed, the longitudinally elongated melt pool breaks up and forms melt accumulations [7]. Additionally, high lateral velocities can increase roughness and bulging surfaces, especially when using larger laser spots [6]. The bulging top surfaces of a layer can result in a process abortion due to collisions with the recoater [8]. The absolute boundaries of the process window regarding the process parameters depend on the system technology used, the processed material, the atmosphere, and a multitude of additional factors.

The Gaussian shape of the single-mode lasers used has attributed an influence on the process limits in some cases [9,10]. Therefore, current research trends are studies investigating the influence of alternative beam shapes on process and component properties in PBF-LB/M. Single-line experiments [11] and multi-layer experiments [12,13] have

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Nomenclature			
A	Absorption	T_0	Ambient temperature
c_p	Specific heat capacity	V	Build rate
d^*	Melt track depth	v_{scan}	Scanning speed
E_v	Volume energy density	w^*	Melt track width
g	Dimensionless temperature field	w_0	Spot radius
h	Hatch distance	x^-	Scan vector length
h^*	Dimensionless hatch distance	x^{-*}	Dimensionless scan vector length
h_s	Enthalpy at melting point	x, y, z	Spatial coordinates
L_f	Latent heat of fusion	x^*, y^*, z^*	Dimensionless spatial coordinates
Pe	Peclet number	α	Thermal diffusivity
P_{laser}	Laser power	δ_{diff}	Thermal diffusion depth
t	Time, integration variable	δ_{layer}	Layer thickness
T_m	Melting temperature	$\Delta H/h_s$	Dimensionless enthalpy
		ρ	Density

shown that the shape of the melt pool changes when applying non-gaussian beam profiles. Melt tracks produced with ring-shaped instead of Gaussian beam profiles are generally wider [9,14]. However, alternative beam profiles often reduce the melt track depth due to a typically lower intensity [9]. This shifts the process window for three-dimensional processes towards higher laser powers [13,15]. Nevertheless, Wischeropp et al. [10] demonstrated that using ring-shaped beam profiles helps to avoid keyhole porosity when processing AlSi10Mg. Moreover, it has been shown that the number and velocity of spatters could be reduced using non-Gaussian beam profiles, indicating reduced melt pool dynamics [16,17].

The cited studies show the potential of alternative beam profiles for enlarging the process window. However, most studies attempt to create similar process conditions using the same absolute process parameters and then compare the process results with common dimensioned process variables, such as the volume energy density, even if the beam sizes are changed. Galbusera et al. [13] stated that using the volume energy density exclusively for evaluating the process result may fail when using alternative beam profiles. Additionally, Bakhtari et al. [18] emphasize that using volume energy density is inappropriate when using alternative beam shapes because of its inefficiency in delivering reproducible results.

To overcome the challenge of comparing process behavior and results using laser spots in various sizes and shapes, this paper presents a new approach to process map representation as an alternative to the traditional $P_{\text{laser}} \cdot v_{\text{scan}}$ process maps and the deduction of dimensionless process windows within those maps. The approach is based on a simple heat conduction model and considers the spot size and thermophysical material properties in the characteristic values. This way, the dimensionless process maps enable mesoscopic process results to be compared using different beam profiles and spot sizes. The process window boundaries for specimens resulting in high relative density and low top surface roughness can be predicted using simple geometric criteria of the modeled melt pools.

2. Material and methods

This section first presents the established absolute characteristic parameters for PBF-LB/M followed by dimensionless parameters, which are combined to form the novel dimensionless process map representation and predict the process limits. This is followed by the experimental conditions and process parameters investigated. Finally, the evaluation methods for the roughness measurements and density determination are described.

2.1. Characteristic process parameters in laser-based powder bed fusion of metals

In order to combine the process parameters in one energy parameter, the volume energy density E_v is typically used for PBF-LB/M. This characteristic value relates the laser power P_{laser} used to the scanning speed v_{scan} , the hatch distance h , and the layer thickness δ_{layer} according to

$$E_v = \frac{P_{\text{laser}}}{v_{\text{scan}} \cdot h \cdot \delta_{\text{layer}}} \quad (1)$$

Process speed is evaluated by the build rate V . The build rate quantifies the volume added to the workpiece at a particular time during exposure according to

$$\dot{V} = v_{\text{scan}} \cdot h \cdot \delta_{\text{layer}} \quad (2)$$

The volume energy density E_v is often used because of its simplicity. However, it neglects many influencing variables that enormously impact the process dynamics and results, such as the spot size and material-specific properties. Therefore, its suitability as a characteristic single process parameter is limited [19]. The dimensionless enthalpy, according to Hann et al. [20], includes thermophysical material properties and the spot size of the laser beam in addition to the standard process parameters of the laser power and the scanning speed. According to [21], it is calculated as

$$\frac{\Delta H}{h_s} = \frac{A \cdot P_{\text{laser}}}{\rho \cdot (c_p \cdot (T_m - T_0) + L_f) \cdot \sqrt{\pi \cdot \alpha \cdot v_{\text{scan}} \cdot (2w_0)^3}} \quad (3)$$

with absorption A , density ρ , specific heat capacity c_p , the delta between melting and ambient temperature $T_m - T_0$, latent heat of fusion L_f , thermal diffusivity α , and laser spot radius w_0 . As such, the dimensionless enthalpy is a measure of the amount of energy introduced into a certain mass relative to the energy required to achieve the melting of this mass. In order to consider the speed aspects in laser material processes, Ion et al. [22] introduced the dimensionless scanning speed as the Peclet number, which is calculated according to

$$Pe = \frac{v_{\text{scan}} \cdot 2w_0}{\alpha} \quad (4)$$

Rubenchik et al. [23] showed that a dimensionless temperature field can be calculated with the Peclet number and the dimensionless coordinates x^* , y^* , and z^* based on the Eager-Tsai model [24]. Thereby, the x and y coordinates are normalized to the beam radius. The z coordinate is normalized to the thermal diffusion depth. The normalization can be summarized according to

$$x^* = \frac{x}{w_0}, y^* = \frac{y}{w_0}, z^* = \frac{z}{\delta_{\text{diff}}} \text{ with } \delta_{\text{diff}} = \sqrt{\frac{\alpha \cdot 2w_0}{v_{\text{scan}}}}. \quad (5)$$

The dimensionless temperature field is represented by the function g , which is calculated using

$$g = \int_0^\infty \frac{\exp\left(-\frac{z^{*2}}{4t} - \frac{y^{*2} + (x^* - t)^2}{\frac{4\sqrt{2t}}{Pe} + 1}\right)}{\left(\frac{4\sqrt{2t}}{Pe} + 1\right)\sqrt{t}} dt. \quad (6)$$

The melt pool boundaries occur at points where the dimensionless temperature field g reaches the dimensionless melting temperature. This can be expressed using the dimensionless enthalpy. According to Rubenchik et al. [23], the melt pool boundaries result at points with the fulfilled condition

$$g = \frac{\pi}{\sqrt[3]{2^3 \cdot \frac{\Delta H}{h_s}}}. \quad (7)$$

Using this model, the dimensionless enthalpy and Peclet number are suitable for predicting the dimensionless melt track depth d^* and melt track width w^* for a heat conduction-driven process regime in a known material [23], even if ring-shaped beam profiles are used [25]. However, the informative value of the dimensionless enthalpy and Peclet number are limited to single melt tracks. In a two- or three-dimensional additive manufacturing process, the hatch distance, layer thickness, and laser return time influence the process results. Since the layer thickness is kept constant in this study, the hatch distance and laser return time are considered exclusively. In addition to the number of tracks required per exposed area, the hatch distance determines the remelted area of neighboring scan tracks. The corresponding proportion significantly influences the accumulated heat. Accordingly, the hatch distance must be scaled with the expected melt pool width to obtain comparable processes. Since the coordinates x and y and thus the melt pool length and width in the model are normalized to the spot radius, the hatch distance is also normalized to the spot radius. This dimensionless hatch distance was already introduced by Thomas et al. [26] as

$$h^* = \frac{h}{w_0}, \quad (8)$$

whereby small values mean a larger overlap of the individual track cross-sections. An h^* of 1.4 corresponds to Meiners' [27] original recommendation for a hatch distance that results in dense components. Besides the hatch distance, the laser return time influences the accumulated heat in the process zone. In the presented approach, this part is considered via the mean dimensionless scan vector length $\bar{x}^*$, which results from the normalization of the mean scan vector length $\bar{x}$ to the spot radius w_0 plus the dimensionless hatch distance h^* according to

$$\bar{x}^* = \frac{\bar{x}}{w_0} + h^*. \quad (9)$$

Consequently, Eq. (9) assumes that hatch sections are scanned in a meandering pattern and only negligible delay times occur between two

neighboring scan tracks.

The build rate is independent of the energy input, as it only includes the scanning speed and the geometric parameters during the PBF-LB/M process. However, this does not provide direct transferability to different materials. In this publication, a dimensionless build rate is introduced. As the experiments are carried out with a constant layer thickness, it is neglected in calculating the characteristic value. As a result, the dimensionless build rate is the product of a dimensionless scanning speed, respectively Peclet number and the dimensionless hatch distance according to

$$Pe \cdot h^* = \frac{v_{\text{scan}} \cdot h}{\alpha}, \quad (10)$$

which means that the dimensionless build rate is proportional to the build rate for a constant thickness and material.

Dimensionless process maps can be drawn from the parameters defined in Eqs. (3), 8, and 10. The ordinate of these maps is formed by the ratio of dimensionless enthalpy to the dimensionless hatch distance and is, therefore, a measure of the energy input. The dimensionless build rate is plotted on the abscissa. This way, processes with different absolute parameters can be compared within one map.

This way, the approach is distinguished from previously published approaches by, e.g., Thomas et al. [26] or Patel and Vlasia [28]. Thomas et al. [26] used for the process map representation one energy dimension, which is calculated using the dimensionless energy and dimensionless velocity introduced by Ion et al. [22] and the self-introduced dimensionless layer thickness and hatch distance. The second dimension is the reciprocal dimensionless hatch distance. Thus, the map representation is particularly suitable for showing the influence of the dimensionless hatch distance on the process results. Patel and Vlasia [28] also extended the approach of Ion et al. [22] and included the dimensionless layer thickness of Thomas et al. [26] within their model. As the hatch distance was not included, the representation of Patel and Vlasia [28] is particularly suitable for observing the melting mode of individual scan tracks.

Besides the pure clustering of process result categories, an additional added value of processing maps is provided if process limits can be predicted. The basis for this prediction is outlined in the following section.

2.2. Prediction of process window boundaries

The heat conduction model from Eq. (6) can extract melt pool characteristics such as length, width, and depth using Eq. (7). In a previous study [25], it was shown that the model could also be used to calculate the melt pool cross-sections when using ring-shaped beam profiles under the condition that the spot size was determined using the 2nd moment method and that the process is still in a conduction-driven process regime. However, in the existing form, the model neglects the influence of neighboring scan tracks on the melt pool shape and size. In this publication, the dimensionless temperature field is therefore extended by two neighboring scan tracks, which are scanned bidirectionally. The temperature field during the exposure of the center of the third scan track results from

$$g = \int_0^\infty \frac{\exp\left(-\frac{z^{*2}}{4t} - \frac{y^{*2} + (x^* - t)^2}{\frac{4\sqrt{2t}}{Pe} + 1}\right) + \exp\left(-\frac{z^{*2}}{4t} - \frac{(y^* + h^*)^2 + (-x^* + \bar{x}^* - t)^2}{\frac{4\sqrt{2t}}{Pe} + 1}\right) + \exp\left(-\frac{z^{*2}}{4t} - \frac{(y^* + 2h^*)^2 + (x^* + 2\bar{x}^* - t)^2}{\frac{4\sqrt{2t}}{Pe} + 1}\right)}{\left(\frac{4\sqrt{2t}}{Pe} + 1\right)\sqrt{t}} dt. \quad (11)$$

The resulting melt pool of the heat source of the first term is analyzed to determine the melt pool dimensions. The second and third terms are implemented to consider the preheating of the analyzed process zone by heat accumulation of neighboring scan tracks. The spatial offset of the neighboring scan tracks in the y^* direction is included via the dimensionless hatch distance h^* . The laser return time is implemented via a local offset of the heat sources in the x^* direction with the dimensionless mean scan vector length $\bar{x}^*$.

Thus, the method described calculates the dimensionless temperature field resulting exclusively from heat conduction based on the process parameters and geometric boundary conditions. The process is assumed to be in a steady state while exposing the center of a third scan vector. This procedure simplifies the complex PBF-LB/M process to a high degree and does not fully represent the underlying physics. However, if the process is conduction-driven, the simplified assumptions are sufficient to predict the melt pool sizes and process limits associated with the melt pool dimensions.

Fig. 1 shows the x^*-y^* plane of an exemplary calculated temperature field, including the artifacts of the fictitious melt pools of the previous scan tracks. A dimensionless mean scan vector length of 71.7 was used in the figure, which would result from a 7 mm track length scanned with a 200 μm spot diameter and a dimensionless hatch distance of 1.7. As can be seen, the directly adjacent scan track (right in Fig. 1), in particular, influences the temperature field. The preheating effect of the next adjacent scan track (left in Fig. 1) is only slightly visible.

The boundaries of the process windows are predicted using geometric characteristics of the melt pool under consideration. For lack of fusion defects, it is assumed that there must be a minimum fusion depth. Since the melt track depths produced between two melt tracks differ significantly with different hatch distances (see Fig. 2), a calculated melt pool depth at the y^* -coordinate of $h^*/2$ (d_h^* in Fig. 2) of at least 5 μm is targeted. For verification of the criterion, the melt pool cross-section is assumed to be elliptical, resulting in the ellipse equation as

$$\frac{y^{*2}}{\left(\frac{w^*}{2}\right)^2} + \frac{z^{*2}}{d^{*2}} = 1. \quad (12)$$

The geometric relationships are sketched in Fig. 3. In order to obtain absolute values for the depth coordinate, the dimensionless melt track depth d^* must be extended by the factor $w_0/\sqrt{Pe}$, as the z -direction in the coordinate system used is normalized to the thermal diffusion depth. With this extension, the limit for lack of fusion results at the position $y^*=h^*/2$ from the condition

$$\frac{d^* \cdot w_0}{\sqrt{Pe}} \sqrt{1 - \frac{h^{*2}}{w^{*2}}} \geq 5 \mu\text{m}. \quad (13)$$

The criterion assumed by Kruth et al. [29] for balling is used for the process limit in the direction of bulging surfaces. Kruth et al. [29]

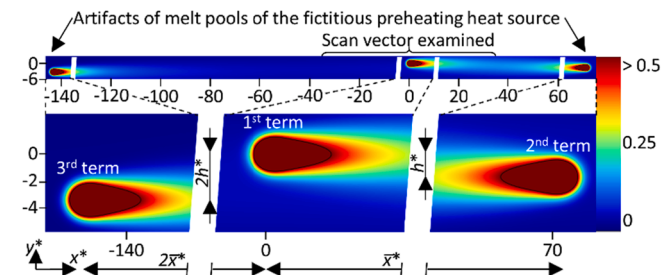


Fig. 1. Calculated dimensionless temperature field according to Eq. (11) using a Peclet number of 15, a dimensionless hatch distance of 1.7, and a dimensionless mean scan vector length of 71.7, melt pool boundaries are visualized black for a dimensionless enthalpy of 3.5.

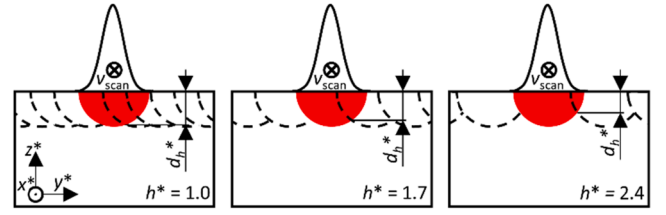


Fig. 2. Sketch of calculated melt pool cross-sections and the dependence of the resulting minimum melt track depth d_h^* on the dimensionless hatch distance h^* .

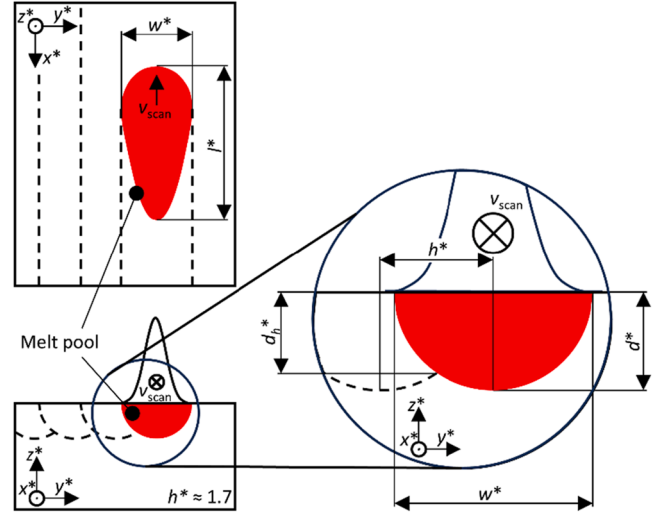


Fig. 3. Sketches of the sectional views of the melt pool in the process zone with the dimensionless melt pool dimensions depth d^* , depth at the half hatch distance d_h^* , length l^* , and width w^* , as well as the dimensionless hatch distance h^* .

describe the melt pool as a half-cylinder shape whose surface tension begins to break the melt pool into spherical accumulations when the length-to-width ratio exceeds 2.1. Therefore, the criterion for a process without bulging surfaces results from the dimensionless melt pool length l^* to width w^* ratio as

$$\frac{l^*}{w^*} \leq 2.1. \quad (14)$$

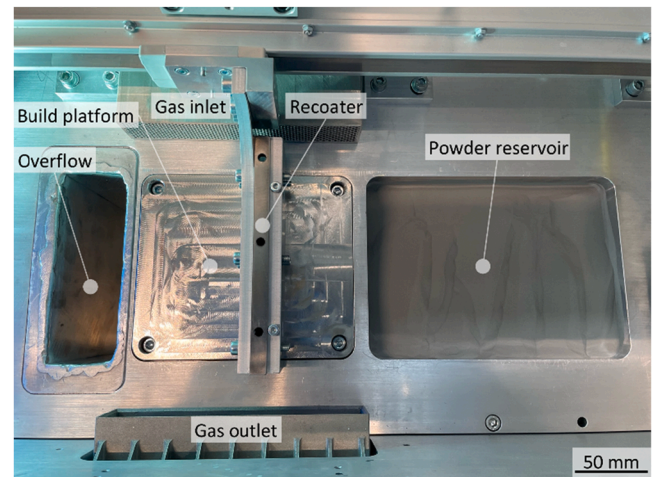


Fig. 4. Photo of the process chamber layout.

2.3. Experimental setup

The validation experiments are conducted on an in-house PBF-LB/M test bench. A fiber laser with switchable beam profiles is used as beam source (AFX-1000, nLIGHT, Inc., Vancouver, WA, USA). This beam source can distribute the laser power between a central Gaussian and a surrounding ring in seven steps. In this study, the beam profiles are named according to the relative power distribution scheme ratio "core/ring". A scanning system with four optical axes for the high-power laser beam is used for deflection, focusing, and expansion (AM-MODULE NEXT GEN, RAYLASE GmbH, Wessling, Germany). The PBF-LB/M process takes place in a self-built process chamber (see Fig. 4) under an argon atmosphere with a residual oxygen content of < 0.15 %. The shielding gas flow is directed over the process zone. In-house programmed software controls the hardware components, which was implemented based on the Machine Control Framework (Autodesk, Inc., San Rafael, CA, USA). This enables the experiments to be carried out automatically using a previously defined 3mf file.

2.4. Design of experiments

All seven beam profiles available with the system technology are investigated. The approximately Gaussian beam profile 95/5 is used as a reference for the state of the art. The alternative beam profiles can be used for different target sizes. Beam profiles with lower intensities in the ring (75/25, 65/35, and 50/50) can be used for local pre- and post-heating of the process zone, while beam profiles with higher intensities in the ring (35/65, 20/80, and 10/90) can possibly influence melt flows. Within the design of experiments, laser power, scanning speed, and hatch distance vary in three stages. The approximately Gaussian beam profile 95/5 is examined in two sizes to separate the influence of the spot size and the influence of the beam profile. Table 1 summarizes the process parameters used. Beam diameters were determined by caustic measurements with a camera-based focus beam profiler (FBP-2KF, CINOGY Technologies GmbH, Duderstadt, Germany) using the 2nd moment method. The parameters are chosen based on preliminary single-track experiments published in [25] to obtain process

Table 2

Thermophysical properties of AISI 316L according to [30].

Name	Value	
Ambient temperature T_0	293.15	K
Density ρ	7800	kg/m ³
Melting temperature T_m	1693	K
Specific heat capacity c_p at T_m	592.55	J/(kg•K)
Thermal diffusivity α at T_m	$6.44 \cdot 10^{-6}$	m ² /s

results showing the process window and boundaries. The dimensionless enthalpy, dimensionless hatch distance, and Peclet number were kept in a similar range in the parameter sets. This design leads to 216 parameter sets examined with three repetitions on different positions of the build platform, whose mean value is shown in the process maps.

AISI 316L (LPW-316-1, LPW Technology Ltd., Runcorn, Cheshire,

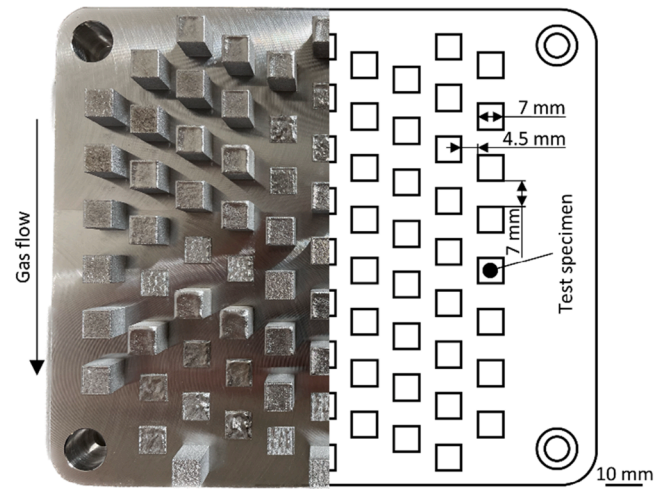


Fig. 5. Illustration of the build job design with a photo of a finished build job on the left and a schematic drawing on the right.

Table 1

Design of experiments.

Power distribution core/ring	Magnification	Spot diameter (2 nd moment) $2w_0$ in μm	Laser power P_{laser} in W	Scanning speed v_{scan} in mm/s	Hatch distance h in μm
 95/5	1	92	200, 250, 300	500, 800, 1100	60, 80, 100
	2	175	300, 400, 500 500, 600, 700	250, 500, 800	140, 175, 210
 75/25	1	125	250, 350, 450	500, 800, 1100	60, 80, 100
 65/35	1	142	300, 400, 500	500, 800, 1100	70, 90, 110
 50/50	1	163	400, 550, 650	500, 800, 1100	70, 90, 110
 35/65	1	189	400, 700, 1000	500, 800, 1100	100, 130, 160
 20/80	1	197	500, 800, 1100	500, 800, 1100	140, 170, 200
 10/90	1	205	500, 800, 1100	500, 800, 1100	140, 170, 200

United Kingdom) is used as the feedstock material. For the calculation of the dimensionless parameters, the thermophysical properties of AISI 316L are taken from [30] (see Table 2). Following King et al. [31], the absorption A is assumed to be 0.6.

2.5. Specimen design and measurement methods

Square cuboids with an edge length of 7 mm and a height of 10 mm are produced as test geometries. The specimens are exposed with a bidirectional pattern. Each layer rotates the scan vectors by a 67° angle increment. The build job design is shown in Fig. 5. Parameter sets that result in protruding parts colliding with the recoater are excluded from the build job within the first 20 layers. The roughness of all top surfaces (including the canceled specimens) is measured using fringe projection (VR-3100, Keyence Corporation, Ōsaka, Japan). Subsequently, the samples are separated from the build platform by wire electrical discharge machining and embedded in an epoxy resin parallel to the build direction. To create cross-sections of the test specimens, the samples are ground with silicon carbide paper in four stages (#180, #320, #800, #1200) and polished with lubricant and polishing suspension in 3 steps (3 μm , 1 μm , and OPS). The grinding plane is at least 2 mm inside the test specimen to exclude edge layer effects. The generated cross-sections are imaged using a digital microscope (VHX-7000, Keyence Corporation, Ōsaka, Japan). With the generated micrographs, the optical density is determined.

3. Results and discussion

The results and discussion section starts presenting the specimen densities and top surface roughness as a function of the dimensionless enthalpy per dimensionless hatch distance. This is followed by process maps of the part density and top surface roughness using dimensioned and dimensionless representations to benchmark the approach introduced in the methodology section with the current state of the art. Within the introduced dimensionless process maps, the process window boundaries for components with high relative density and low top surface roughness are predicted for different dimensional hatch distances using the conditions from the methodology section. Since a multidimensional experimental design with various beam shapes, spot sizes, and process parameters is investigated in this study, using common $P_{\text{laser}}-V_{\text{scan}}$ process maps is not feasible. Therefore, the volume energy density E_v as energy input and build rate V as process speed are used for the dimensioned process maps. The dimensionless process map uses the

dimensionless parameters dimensionless enthalpy per dimensionless hatch distance and the dimensionless build rate as axes.

Fig. 6 presents the measured data as a function of the dimensionless enthalpy per dimensionless hatch distance. Fig. 6a shows the relative component densities, while Fig. 6b shows the top surface roughness values. The density and the roughness values tend to increase with increasing dimensionless enthalpy per dimensionless hatch. However, the scatter of the values when considering only the energy input is so high that values with very low densities ($< 99\%$) and high roughnesses ($Sa > 90\ \mu\text{m}$) as well as high relative densities ($> 99.8\%$) and low roughnesses ($Sa < 40\ \mu\text{m}$) are present over an extensive range of the dimensionless enthalpy per dimensionless hatch distance. The standard deviation of the measured values increases with greater deviation from the desired values (high density and low roughness). Consequently, low mean values of density and high mean values of top surface roughness tend to have higher standard deviations, which is expected since the process is not reliable in these ranges. The form of representation in Fig. 6 shows that the energy-related dimension does not sufficiently reflect the process concerning the process results of density and top surface roughness. This way, the importance of suitable process maps is clearly emphasized.

Fig. 7 shows the relative density process map of the manufactured specimens with the two representation methods. Fig. 7a shows the dimensioned representation, while Fig. 7b shows the dimensionless process map.

At low build rates ($\leq 3\ \text{mm}^3/\text{s}$), beam profile 95/5 frequently results in keyhole porosity in the slowest scan velocity levels investigated. Build rates of approximately $3\ \text{mm}^3/\text{s}$ are common for state-of-the-art processes [32]. Beam profiles 75/25, 65/35, and 50/50 exhibit keyhole porosity at volume energy densities above $250\ \text{J}/\text{mm}^3$. No keyhole porosities are detected for the beam profiles 35/65, 20/80, and 10/90. Lack of fusion porosities occur above a build rate of approx. $6.5\ \text{mm}^3/\text{s}$ for the beam profiles 95/5, 35/65, 20/80, and 10/90 at volume energy densities below $85\ \text{J}/\text{mm}^3$. For the beam profiles 75/25, 65/35, and 50/50, no lack of fusion is detected in the examined parameter range. Additionally, below a build rate of $7\ \text{mm}^3/\text{s}$, the samples of beam profile 95/5 exhibit a lack of fusion at the lowest power levels and the largest hatch levels. With the combination of high build rates and high volume energy densities, the process often results in protruding specimens. The build job for the corresponding specimens is aborted. Consequently, no density values could be determined (gray symbols in Fig. 7a).

The process boundary to the lack of fusion is mainly due to insufficient energy for the complete fusion of the powder particles or a hatch

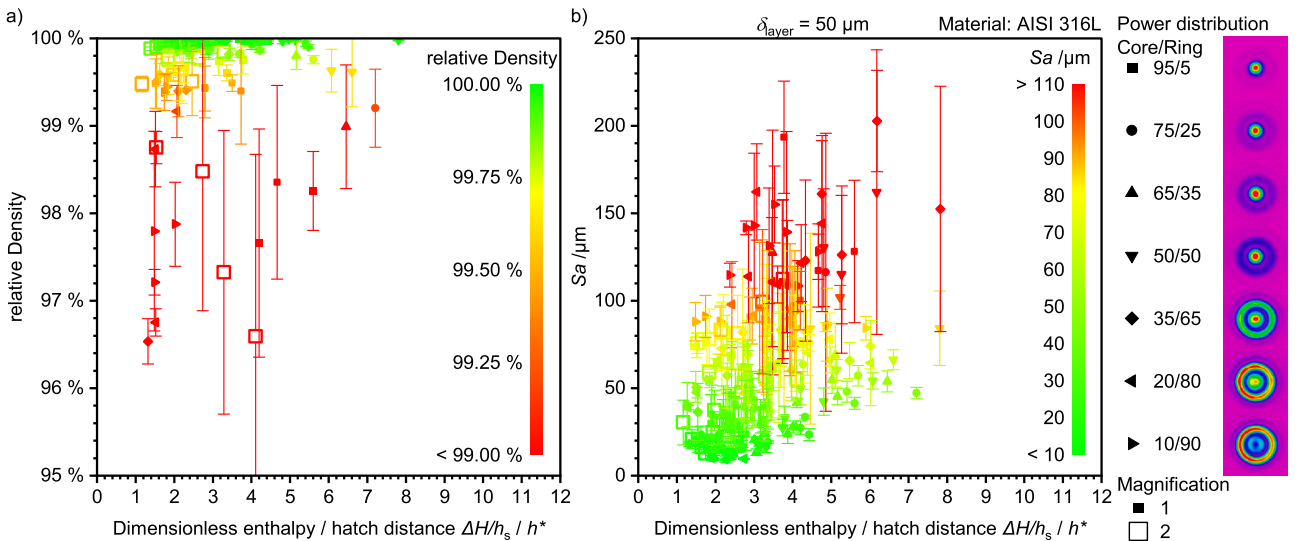


Fig. 6. a) Relative part density and b) top surface roughness as a function of the dimensionless enthalpy per hatch distance using different beam shapes.

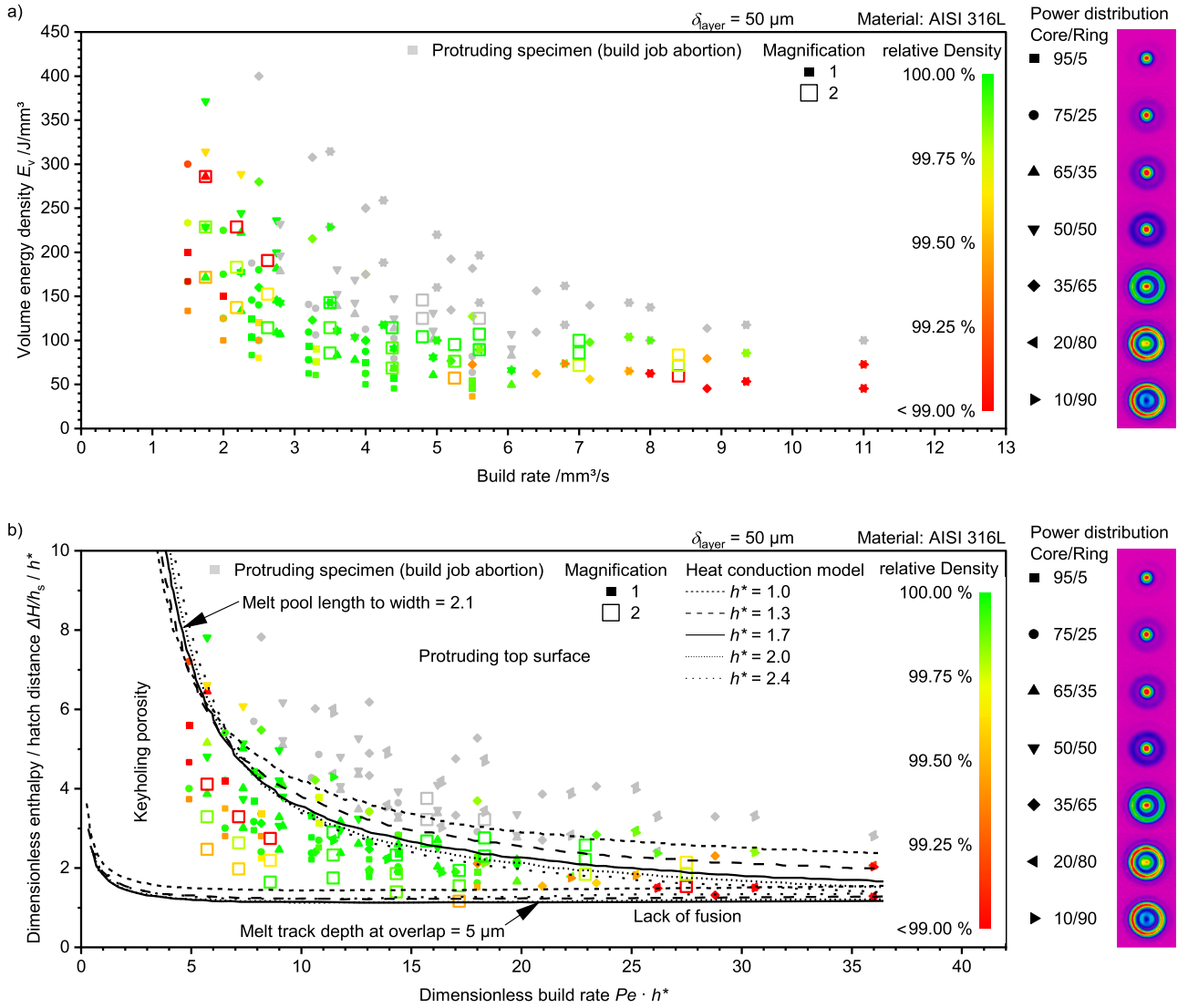


Fig. 7. Process maps of relative part density using different beam shapes a) based on volume energy density and build rate and b) based on dimensionless enthalpy per hatch distance and the dimensionless build rate, including predicted dimensionless process window boundaries for various dimensionless hatch distances h^* .

distance that is too large, resulting in the single tracks not being sufficiently connected. The boundaries of the density process window for keyholing porosity and protruding parts with protruding top surfaces are strongly blurred. For example, with the beam profiles 95/5, 20/80, and 10/90, it is still possible to generate highly dense components with a volume energy density of 120 J/mm³ and a build rate of 4.4 mm³/s. However, the processes with the beam profiles 75/25, 65/35, and 50/50 had to be aborted in this parameter range. This is mainly due to the lack of presenting the intensity distribution and the spot size in the characteristic values used. Galbusera et al. [13] have attempted to create comparability in a similar parameter study on an Al alloy using the volume energy density. However, the study concluded that using volume energy density to evaluate the process results fails when alternative beam profiles are used. Moreover, Bakhtari et al. [18] state that using volume energy density is inappropriate for alternative beam shapes because non-comparable values can be obtained for differing beam shapes. To address this issue, the new approach to presenting a dimensionless process map introduced in the methodology section is shown in Fig. 7b.

The dimensionless process map in Fig. 7b is qualitatively similar to the dimensioned process map in Fig. 7a. Keyhole porosity occurs mainly for the nearly Gaussian beam profile 95/5 below $Pe \cdot h^*$ of 9. For the

beam profiles 75/25, 65/35, and 50/50, keyhole porosity occurs above a dimensionless enthalpy per dimensionless hatch distance of 6. In the range of a dimensionless build rate below 9 and a dimensionless enthalpy per dimensionless hatch distance below 6, components with a very high density are produced for the beam profiles 75/25, 65/35, and 50/50. This means that in terms of keyhole porosity, the boundary of the process window is strongly dependent on the beam profile. The almost Gaussian beam profile 95/5 has the highest tendency to result in keyhole porosity. If keyholes are present during processing using non-Gaussian beam profiles in the same parameter range, distributing the laser power in the ring may stabilize the existing keyholes and thus prevent collapses that lead to porosities, as demonstrated by Yuan et al. [33] using x-ray imaging. The dependence of the keyhole porosity on the beam profile, even with similar dimensionless characteristic values, demonstrates the advantages of alternative beam profiles. At the same time, it shows that the heat conduction model is not suitable for reliably predicting keyhole porosity. This limit of the model was expected, as the model neglects convection effects caused by evaporation or melt flows, for example.

In the range of a dimensionless enthalpy per dimensionless hatch distance of 1.5 lies the transition from a process that results in dense components to a lack of fusion process. The transition is more evident

when using the dimensionless parameters independent of the beam profile than when using the dimensioned parameters. The condition criterion in Eq. (13) can predict this lack of fusion boundary. The improved overlap of the dimensionless process boundaries for lack of fusion and protruding specimens when using alternative beam profiles is mainly because the dimensionless approach moves the coordinate points for very small beam profiles to higher values on the ordinate. The reason is the heat conduction model behind the dimensionless enthalpy, which includes the thermal diffusion depth of the introduced heat over the interaction time between the beam profile and a considered point in the specimen. With smaller beam profiles, the considered volume (and thus the mass) is smaller, which leads to higher enthalpies introduced in the considered volume. This introduced enthalpy is not dependent on the beam shape and correlates very well with the phenomena of lack of fusion.

Analogous to the dimensioned process map, the data points of the bulging samples (gray dots in Fig. 7b) are located in an area with a high energy input (dimensionless enthalpy per dimensionless hatch distance) and high process speeds (dimensionless build rate). The boundary to this area of protruding parts is pronounced with the help of the dimensionless parameters. The regions of built and aborted specimens overlap

significantly less in Fig. 7b. With increasing dimensionless build rate, lower dimensionless enthalpies per dimensionless hatch distance result in protruding parts. Qualitatively, the condition from Eq. (14) describes the progression of the boundary between the highly dense samples and the samples having to be aborted during the process. However, some samples still exhibit high density values with parameters above the predicted process window limit.

Fig. 8 shows the process map concerning the top surface roughness of the samples. When using the dimensioned values, there is a large overlap of areas with high and low roughness (see Fig. 8a), similar to the density distribution in Fig. 7a. The different roughness areas are separated if the dimensionless process map approach is applied. The samples with increased keyhole porosity (see Fig. 4) exhibit high roughness values. Above the dimensionless build rate of 20, only a few specimens exhibit a roughness Sa of $< 40 \mu\text{m}$ (see Fig. 8b). The corresponding samples exhibit lack-of-fusion porosities (see Fig. 7b). The protruding specimens (gray symbols in Fig. 7), which were aborted during the build process, show very high roughness values ($Sa \geq 70 \mu\text{m}$). Remarkably, the roughness increases abruptly in the range predicted by the condition in Eq. (14), suggesting that this roughness is primarily caused by balling or another surface tension-driven physical phenomenon for long and

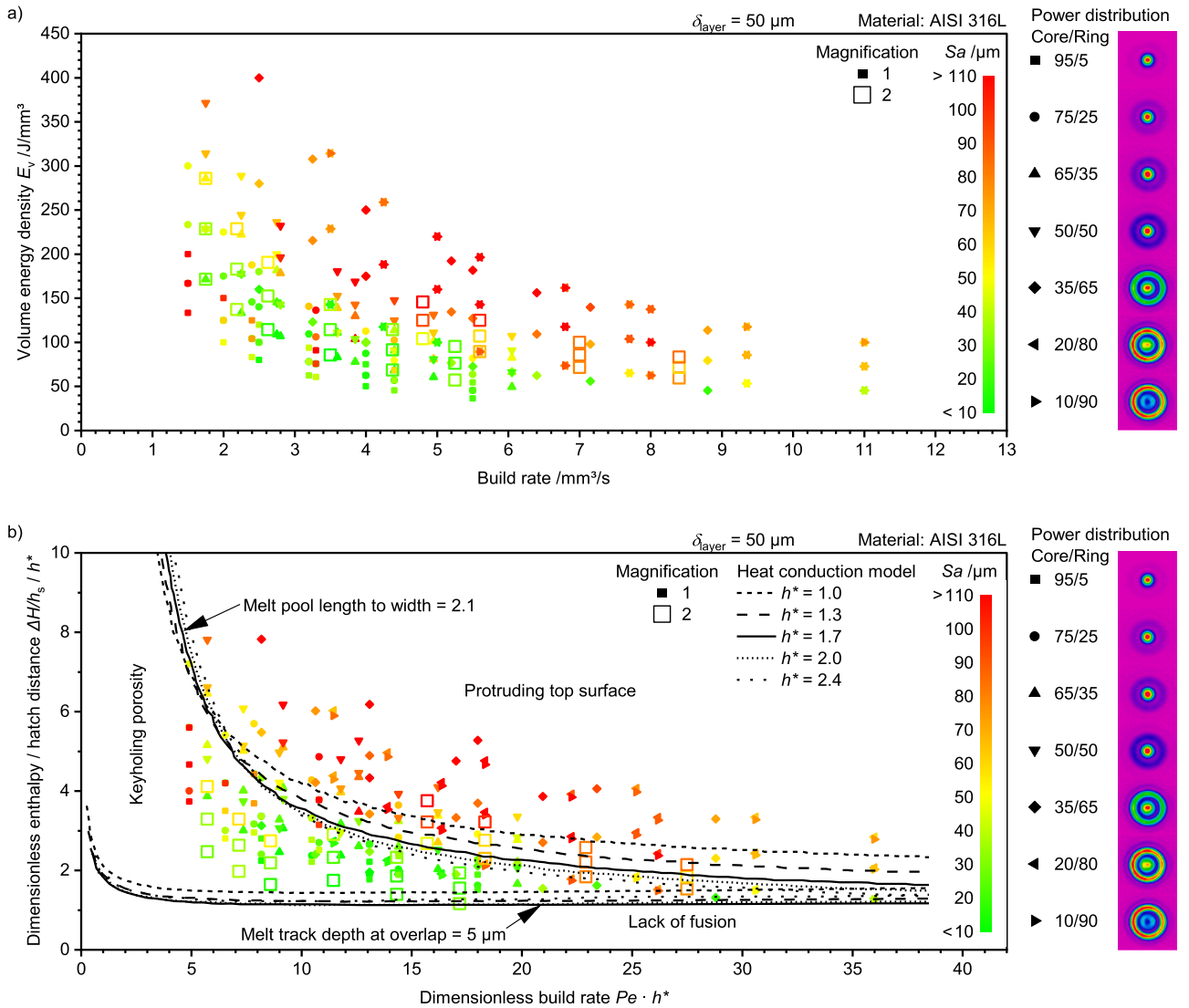


Fig. 8. Process map of top surface roughness Sa using different beam shapes a) based on volume energy density and build rate and b) based on dimensionless enthalpy per hatch distance and the dimensionless build rate, including predicted dimensionless process window boundaries for various dimensionless hatch distances h^* .

narrow melt pools produced at high Peclet numbers. At the same time, the high roughness does not directly lead to samples with a reduced roughness but initially results in dense, very rough, or wavy samples. This effect is illustrated in Fig. 7b and is also known from powder bed fusion using an electron beam [34].

As already introduced, the dimensionless process maps contain the predicted process limits as line graphs. The graphs show the parameter combinations in which the conditions from Eqs. (13) and 14 were fulfilled for various dimensionless hatch distances. The dimensionless hatch distances shown are examples of the levels of dimensionless hatch distances evaluated experimentally. The predicted process limits for the different dimensionless hatch distances shown are within a small corridor. However, the simulation shows an interesting trend: the limit for constant dimensionless build rates for small dimensionless hatch distances shifts towards larger dimensionless enthalpy per dimensionless hatch distance. This means that the process window for very small dimensionless hatch distances ($h^* < 1.0$) would be enlarged towards the limit of bulging respectively protruding surfaces. Thus, reducing the dimensional hatch could achieve higher build rates, which may sound counterintuitive. Interestingly, this happened without being explicitly discussed in many publications. In [6], it was summarized that in studies with larger beam profiles, the hatch distance often remains within the range of the Gaussian state-of-the-art process: about 70 μm [12,13], 90 μm [35], and 100 μm [36]. Reducing the hatch distance means the scanning speed would have to be significantly raised to increase the dimensionless build rate. The increased speed would require higher laser power (raised by the square root of the hatch reduction) to maintain dimensionless enthalpies in a suitable range.

To summarize the core findings regarding the process maps: When using dimensionless process maps, four general areas can be separated as in classic process maps (consisting of scanning speed and laser power for a Gaussian beam profile). The four emerging categories are shown in Fig. 6 with exemplary cross-sections and briefly described below:

- Keyholing (see Fig. 9a): The process shifts into keyholing in the areas with high energy inputs and low process speed. Keyhole porosities lead to a reduced relative density. In most cases, the top surface has an increased roughness. Beam profile 95/5 has the highest tendency to result in keyhole porosity.

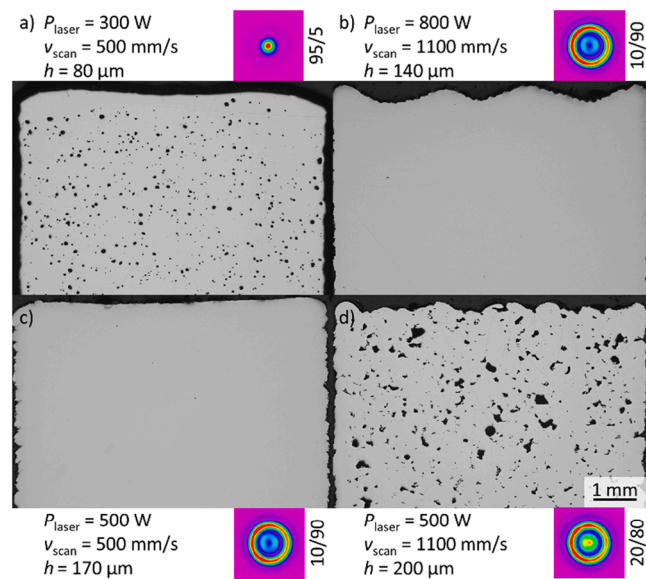


Fig. 9. Exemplary cross-section for specimen resulting in a) keyholing, b) bulging top surface, c) dense components with smooth top surface, and d) lack of fusion.

- Lack of fusion (see Fig. 9d): In the area with high process speed and low energy inputs, lack of fusion porosities occurs due to incomplete melting and fusion of powder particles. The top surface is typically slightly rough in these cases. The boundary is independent of a beam shape and can be predicted by a minimum calculated melt track depth at the center line between two adjacent melt tracks.
- Protruding specimens (see Fig. 9b): Specimens with very high top surface roughness values are produced when applying high process speeds and energy inputs (see Fig. 8). The specimens may have a high relative density. The manufacturing of the corresponding specimens was mostly aborted during the process due to recoater collisions. This boundary is hardly dependent on the beam profile when using dimensionless parameters. The boundary to this process regime can be well predicted with the condition according to Eq. (14), which states that the melt pool should be shorter than 2.1 times the melt pool width for a stable process.
- Stable process resulting in dense and smooth parts (see Fig. 9c): In the center of the dimensionless process map is an area where highly dense specimens with a low top surface roughness can be manufactured.

In conclusion, especially the process window boundary regarding keyhole porosity depends on the beam profile. Alternative beam profiles have a larger process window in the direction of the keyhole porosity than the almost Gaussian beam profile 95/5. The process limits regarding lack of fusion and protruding surfaces can be determined experimentally and predicted simulatively, independently of the beam profile.

4. Conclusion

The present study demonstrates the feasibility of a novel approach to process map representation using dimensionless parameters. A simple heat conduction model is used to predict the process boundaries within the dimensionless process maps. A comprehensive parameter study with seven different beam shapes of various sizes and varied laser power, scanning speed, and hatch distance is conducted and presented to prove the approach. To experimentally determine the process limits, the density and the top surface roughness are determined on cuboidal AISI 316L specimens ($7 \times 7 \times 10 \text{ mm}^3$).

The results can be summarized as follows: By using dimensionless process maps with the dimensionless enthalpy per dimensionless hatch distance as energy parameter and dimensionless build rate as velocity parameters, the process results using different beam profiles can be compared independent of the spot size:

- The experimental deduced process boundaries of all beam profiles investigated overlap in the dimensionless process maps in the direction of lack of fusion and protruding top surfaces. The boundaries can be predicted using the proposed simple heat conduction model. Consequently, the corresponding process boundaries are hardly dependent on the beam profile.
- The boundaries of the process windows regarding keyhole porosity differ between the use of almost Gaussian beam profiles and ring-shaped beam profiles. The samples produced using Gaussian beam profiles most frequently exhibit keyhole porosity. This proves on the one hand that alternative beam profiles stabilize the process. On the other hand, it shows that the simple analysis of heat conduction is not sufficient to draw conclusions in a process regime driven by convection and evaporation.

The results demonstrate that ring-shaped beam profiles increase the process reliability by reducing the tendency to unstable keyholing in PBF-LB/M manufactured parts. The consideration based on the newly introduced visualization with dimensionless parameters includes the size dependency of the process and, thus, introduces a new holistic

approach. Moreover, the simulation results show that it may be possible to increase the build rate, particularly with large beam profiles, by reducing the dimensionless hatch distance and simultaneously increasing the scanning speed overproportionally. This publication validates the model presented by processing the material AISI 316L with a constant layer thickness of 50 μm . However, by including the absorption and thermophysical properties of the processed material, the model and the method are transferable to further materials and beam shapes without extensive calibration effort. Furthermore, an extension of the model concerning thicker layers would be possible in future work, as a variation of the layer thickness has not yet been considered in the approach presented.

CRedit authorship contribution statement

Jonas Grünwald: Writing – review & editing, Writing – original draft, Visualization, Validation, Supervision, Software, Resources, Project administration, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **Katrin Wudy:** Writing – review & editing, Validation, Supervision, Resources, Project administration.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Data availability

Data will be made available on request.

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Influence of ring-shaped beam profiles on spatter characteristics in laser-based powder bed fusion of metals

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ABSTRACT

Despite the maturity of laser-based powder bed fusion of metals (PBF-LB/M), some barriers prevent the manufacturing process from fully being established in the industry. One drawback is spatter formation, which is disadvantageous to PBF-LB/M for three main reasons. First, adhering spatter can initiate coater collision, resulting in process failure. Second, large adhering spatter may cause lack-of-fusion defects as they require more energy to remelt sufficiently compared to unprocessed powder. Furthermore, big nonadhering spatter cannot be recycled as powder. The recycling of small spatter particles potentially results in degraded material properties. Ring-shaped beam profiles have been established for deep penetration welding to reduce spatter formation. Investigations on ring-shaped beam profiles in PBF-LB/M focus on improving productivity and process robustness. Qualitative spatter reduction in PBF-LB/M using ring-shaped beam profiles has also been reported. This publication quantitatively examines the influence of ring-shaped beam profiles on spatter formation in PBF-LB/M. Image processing algorithms of on-axis high-speed images are utilized for spatter detection and tracking. A self-developed spatter segmentation is used to determine the spatter size. A Laplacian of Gaussian filter is combined with a Kalman tracker to count and track the spatter. The results show that spatter formation is highly influenced by the beam profile and the chosen process parameters. Considering the melt track width, ring-shaped beam profiles could reduce the number of spatter per fused area. High numbers of spatter are generated when parameter sets result in balling. Moreover, spatter velocity is primarily dependent on the introduced dimensionless enthalpy.

Key words: additive manufacturing, PBF-LB/M, splatter, melt pool ejections, denudation, stainless steel AISI 316L, beam shaping, donut-shaped profile

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I. INTRODUCTION

Laser-based powder bed fusion of metals (PBF-LB/M) is the metal additive manufacturing process with the highest industrial maturity. During the process, subsequent powder layers are applied and selectively melted using a laser. Nowadays, single-mode laser beam sources with powers up to 1 kW are typically used.¹ The laser beam is scanned across the powder bed. Hereby, the desired areas of the deposited powder layers are fused by hatching individual melt tracks. Due to the high laser intensities and introduced energy, high temperatures, temperature gradients, and partial evaporation of the

material occurs during exposure.² The evaporation and the melt flows caused by temperature gradients lead to high dynamics in the melt pool² and can result in process errors such as unstable keyhole formation,³ balling,⁴ or spatter formation.⁵

In particular, spatter formation represents a major risk for process reliability. On the one hand, spatter adhering to already melted areas represent a risk for a robust coating process.⁶ On the other hand, incompletely remelted spatter affects solidification behavior,⁷ provoking lack of fusion,⁷ and porosity in the final part.⁶ Furthermore, due to its size, the largest amount of spatter cannot be

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reused for following processes, which reduces the material recyclability.⁵ Yet, the reuse of smaller spatter can adversely affect the material properties due to increased oxygen content.⁸ Spatter can be divided according to two different criteria. The temperature (cold or hot spatter⁹) and the origin (powder bed or melt pool⁹). Cold powder spatter are represented by the steam-induced entrainment of surrounding powder particles.⁵ This effect is also referred to as powder denudation¹⁰ and may redistribute powder around the melt track. Whenever entrained powder particles pass the laser beam, they get heated, eventually melted, and become hot powder spatter.⁹ Multiple powder particles can also fuse and, thus, form larger hot powder spatter.⁹ In addition, there are spatter originating from melt pool instabilities caused by recoil pressure and Marangoni flows.^{5,11} This hot melt pool ejecta is often called droplet spatter.⁵ Ly *et al.*⁹ investigated PBF-LB/M processes of AISI 316L and Ti6Al4V with laser powers between 200 and 300 W and scanning speeds between 1.5 and 2 m/s. They concluded that approximately 60 % of all spatter were hot powder spatter, around 25 % were cold powder spatter, and the smallest fraction of approximately 15 % was melt pool ejecta.⁹

Spatter characteristics, such as number, velocity, and size, are influenced by part geometry,¹² material,¹³ and process parameters.^{13–16} Larger amounts of spatter tend to occur in overhanging areas and areas with a high surface-to-volume ratio.¹² Gunenthiram *et al.*¹³ found that a higher amount of spatter is present for AISI 316L stainless steel than for A404 aluminum for similar process conditions. Furthermore, increased volumetric energy density results in larger amounts of spatter.^{13,14} Likewise, Andani *et al.*¹⁵ reported that either a decrease in scan speed or an increase in laser power results in a higher number of spatter, with scan speed having a greater effect compared to laser power. In contrast, Young *et al.*¹⁶ found no discernible effect of laser power and scan speed on spatter characteristics in their study. However, a larger spot size results in less spatter formation.^{14,16} Gunenthiram *et al.*¹³ and Sow *et al.*¹⁴ attribute the observed trends to the material evaporation and resulting vapor pressure at the melt pool surface, which promotes spatter formation. Also, in laser welding, the origin of spatter is attributed to local vaporization and the resulting acceleration of the molten metal.¹⁷ With PBF-LB/M, spatter formation is often considered an unavoidable process-by-product.⁵ To stabilize the process, an attempt is made to quickly remove any spatter from the process zone via an inert gas flow.¹⁸ However, there are also studies on how to prevent spatter formation by process adaptations. For example, adding helium to the shielding gas reduces the number of hot spatter.¹⁹ Moreover, transferring the strategy of powder bed presintering from electron beam-based powder bed fusion to PBF-LB/M is possible.^{20,21} This ensures the powder particles are already bonded together before the melt exposure.²¹ As a result, hot spatter can be reduced by up to 93 %.²⁰ Koike and Sugiura^{22,23} analyzed the spatter characteristics with increased gravitational force. Due to the increased gravity, the existing spatter was accelerated more toward the powder bed, resulting in a reduced trajectory height until the suppression of spatter at 10 G.^{22,23} All these approaches involve increased effort (presintering and high gravity) or cost increases (helium addition).

Another approach is to adjust the melt pool surface temperature and the resulting vapor pressure¹¹ by using alternative beam profiles. In laser deep penetration welding, intensity distributions with a ring surrounding the central intensity peak result in a reduced spatter formation.^{24,25} Gaussian beam profiles are used as the standard beam profile

in PBF-LB/M.¹ Several research groups study the influence of different beam profiles on spatter behavior. Slodczyk *et al.*²⁶ studied the amount of spatter using on-axis high-speed imaging. The PBF-LB/M process was parallelized by a multispot beam shape arranged perpendicular to the scan direction. A similar number of spatter per scan track was counted with a single-spot exposure, a three-spot exposure, and a five-spot exposure, offering enormous potential for reducing the number of spatter per fused area. Due to the distribution of the laser power to several spots, a homogeneous temperature distribution was assumed.²⁶ Tumkur *et al.*²⁷ compared the spatter characteristics of a Bessel beam shape with a Gaussian beam shape using off-axis high-speed imaging. When using the Bessel beam shape, the amount of spatter was reduced by 50 %.²⁷ Additionally, the velocity of hot spatter was significantly higher when using the Gaussian beam, indicating higher recoil pressures.²⁷ Okunkova *et al.*²⁸ found a qualitative reduction in the amount of spatter when changing the beam profile from a Gaussian beam profile to a ring-shaped beam profile. Rothfelder *et al.*²⁹ also found less spatter for a ring-shaped beam. However, Rothfelder *et al.*²⁹ used tungsten tracer particles and characterized the rearranged particles *ex situ* on the fabricated sample.

The cited studies^{26–29} suggest that spatter characteristics strongly depend on the intensity distribution. A change in intensity distribution is typically associated with beam enlargement and loss of resolution in the PBF-LB/M.³⁰ Some modern beam sources offer the ability to change the intensity distribution on the fly.^{30,31} Therefore, adjusting the beam profile according to the required resolution would be possible. For a deeper insight into spatter behavior with ring-shaped beam profiles, this paper presents an *in situ* investigation of the spatter characteristics using different ring-shaped beam profiles in different spot sizes and process parameters.

II. METHODOLOGY

A. Feedstock material

The feedstock material is AISI 316L stainless steel, a widely used material in additive manufacturing. MetcoAdd 316L-A (Oerlikon Metco Europe GmbH, Raunheim, Germany) is selected as powder material. The nominal powder diameters are specified with $d_{10,3} = 19 \mu\text{m}$, $d_{50,3} = 30 \mu\text{m}$, and $d_{90,3} = 46 \mu\text{m}$. The substrate plates are laser-cut, cold-rolled sheets with a corundum-blasted surface ($S_a \approx 3 \mu\text{m}$, $S_z \approx 42 \mu\text{m}$).

B. Experimental setup

The experiments are realized on a self-developed test bench. A fiber laser with seven switchable beam profiles and a maximum laser power of 1200 W (AFX-1000, nLIGHT Inc., Vancouver, WA, USA) is used as a beam source. The beam profiles are designed with a central Gaussian peak and a surrounding ring, between which the power can be distributed in discrete steps. In this publication, the shapes are named according to the relative distribution between the Gaussian core and the surrounding ring (see the first column in Table I). Beam deflection and expansion are performed using a four-axis scanning system (AM-MODUL, RAYLASE GmbH, Wessling, Germany). The scanning system is extended by a sensor module (RAYLASE GmbH, Wessling, Germany) that houses a high-speed camera (CP90-3-M-540, Optronis GmbH, Kehl, Germany) oriented

TABLE I. Process parameters for single-track exposure.

Power distribution	core/ring	Magnification	Spot diameter $2w_0$ (μm)	Laser power P_{laser} (W)	Scanning speed v_{scan} (mm/s)	Layer thickness t_{powder} (μm)
	95/5	1	93	100, 150, 200, 250, 300	800	50
		1.5	134	200, 250, 300, 350		
		2	176	200, 300		
	75/25	1	126	100, 150, 200, 250, 300	800	50
		1.5	185	200, 250, 300, 350		
		2	243	200, 300, 400, 500		
	65/35	1	143	100, 150, 200, 250, 300	800	50
		1.5	208	200, 250, 300, 350		
		2	273	200, 300, 400, 500		
	50/50	1	165	150, 200, 250, 300,	800	50
		1.5	239	250, 300, 350, 400, 450		
		2	314	400, 500, 600, 700		
	35/65	1	190	200, 300, 400, 500	800	50
		1.5	277	400, 500, 600, 700		
		2	365	600, 750, 900, 1050, 1200		
	20/80	1	198	300, 400, 500, 600	800	50
		1.5	288	400, 500, 600, 700, 800		
	10/90	1	205	300	800	50
		1.5	300	400, 500, 600, 700, 800		

coaxially to the processing laser. An additional movable focusing lens system is implemented to keep the camera in focus at different beam expansions and deflection angles. A frame grabber (Coaxlink Quad G3, Euresys SA, Seraing, Belgium) is used for high-speed image acquisition. To ensure an inert PBF-LB/M process environment, all studies are conducted inside an experimental chamber providing an argon atmosphere (maximum 0.2 % oxygen). During the exposure of the single tracks, the substrate is overflowed with a volume flow of 16 l/min to remove any process emissions. The optical part of the experimental setup is shown in Fig. 1.

C. Design of experiments

To investigate spatter characteristics, single tracks with a length of 12 mm are exposed on a 50 μm powder layer. The powder layer is applied manually using a 50 μm gauge strip. The process parameters used are shown in Table I. The listed spot sizes were determined camera-based at 320 W using the second moment method.

D. Determination of spatter characteristics

High-speed camera images are acquired and subsequently evaluated to characterize the spatter characteristics. The images are acquired without additional illumination of the process zone using two exposure times of 2 and 20 μs and a frame rate of 20 000 fps resulting in approximately 300 analyzed images per single track. Due to the high data rates, image processing is performed after acquisition by a separate script. Image processing is done in four steps. The steps are shown in Fig. 2 using an example image.

- (1) Initial data are all single images of an individual track taken by the high-speed camera.
- (2) In an image preprocessing step, the process zone is blacked to only reveal the spatter in the image. For this purpose, a process mask is calculated for each track, according to Haubold.³² To compute the threshold of the process mask, all images of the image series of a single track are averaged. From this averaged image, an Otsu threshold³³ is calculated. To generate a fitting process mask, empirical data have shown that an enlargement of this area is needed.³² Therefore, a morphological dilation filter with a seven times seven kernel size is applied. The pixel values of all pixels included in this area are set equal to 0 for each image of the series. The procedure is illustrated in Fig. 3.
- (3) For spatter size determination, image segmentation is performed by binarization with another Otsu threshold. To calculate this additional threshold, the Otsu method is applied to the averaged image with an excluded process zone. This calculation procedure is shown in Fig. 4. For spatter detection, the average image (background model) is subtracted from each image of the respective track with excluded process zone (see Fig. 5). The resulting image foreground is binarized with the calculated global threshold. As only the spatters are shown in these images, the remaining connected components can be analyzed concerning spatial expansion. It should be noted that the detected spatter size represents the projected size of a hot, thermal self-luminescent spatter. Using the Otsu threshold on this projected representation, the spatter are detected as slightly larger than they are. Nevertheless, qualitative trends in spatter size can be determined.

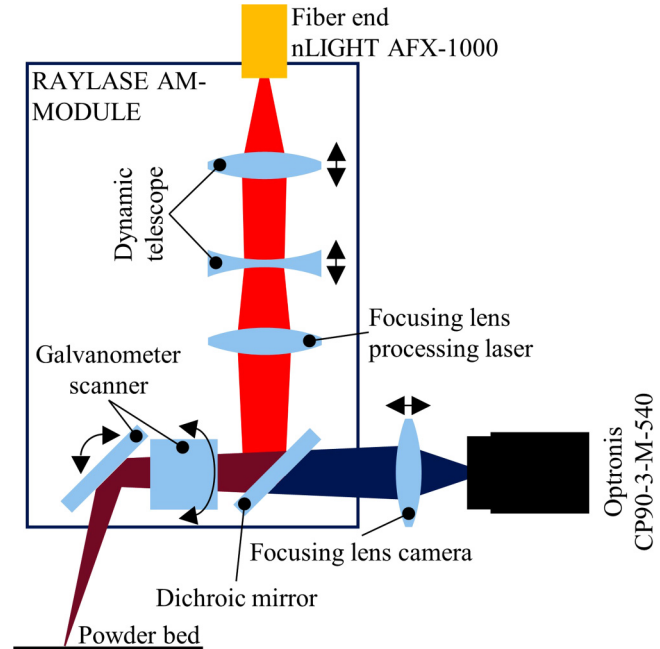


FIG. 1. Schematic of the optical test setup.

- (4) The number and the velocity of the spatter are determined by applying the combination of a Laplacian of Gaussian (LoG) detector followed by a Kalman tracker. This approach has already been implemented by Jäger *et al.*³⁴ in laser welding and by Slodczyk *et al.*²⁶ in PBF-LB/M and showed good results. The implementation is done using TrackMate.³⁵

III. RESULTS AND DISCUSSION

The results are structured according to the spatter number, size, and velocity. In the graphs, the parameter sets are classified by the single-track surface. “Stable” designates parameter sets exhibiting smooth surfaces. “Balling” denotes single tracks with melt accumulations along the single track. “Keyholing” refers to parameter sets with a weld depth greater than 80 % of the single-track width. To make the results of the different beam profiles comparable, the dimensionless enthalpy is used as a reference value. Hann *et al.*³⁶ introduced this parameter and showed the potential for predicting the weld depth in laser welding. In this paper, the dimensionless enthalpy, according to King *et al.*,³⁷ is used, who adopted the dimensionless enthalpy to determine the keyhole mode in PBF-LB/M. The characteristic value is calculated according to

$$\frac{\Delta H}{h_s} = \frac{A \cdot P_{\text{laser}}}{\pi \cdot h_s \cdot \sqrt{\alpha \cdot v_{\text{scan}} \cdot w_0^3}}, \quad (1)$$

with absorption A , laser power P_{laser} , enthalpy at melting h_s , thermal diffusivity α , scan speed v_{scan} , and spot radius w_0 . The

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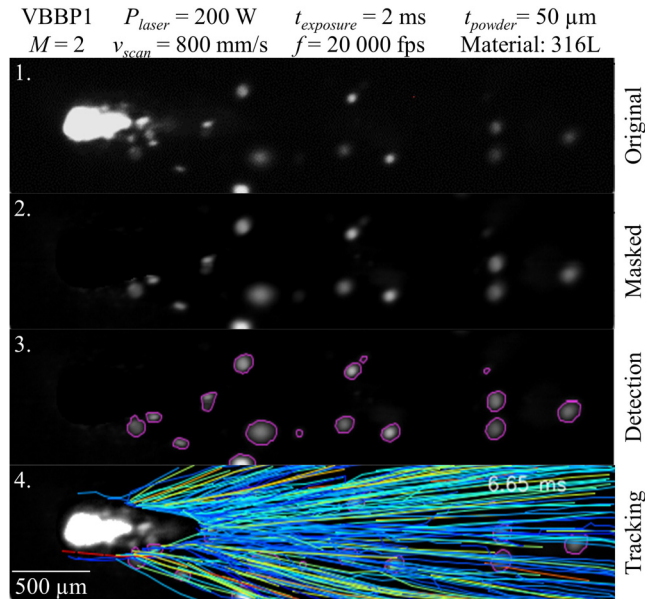


FIG. 2. Four steps of image processing.

absorption A is assumed to be 0.5. The thermal diffusivity is given by $\alpha = \kappa / (\rho \cdot c_p)$ with the thermal conductivity κ , the density ρ , and the specific heat capacity c_p . The enthalpy at melting is given by $h_s = \rho \cdot c_p \cdot T_m$.³⁷ The thermal properties are determined according to Mills³⁸ as melting temperature T_m of 1693 K, density ρ of 7800 kg/m³, specific heat capacity c_p at T_m of 592.55 J/(kg K), and thermal conductivity κ at T_m of 29.77 W/(m K).

A. Influence on the number of spatter

Figure 6 shows the number of tracked spatter per millimeter single-track length. Overall, the results are rather heterogeneous, so

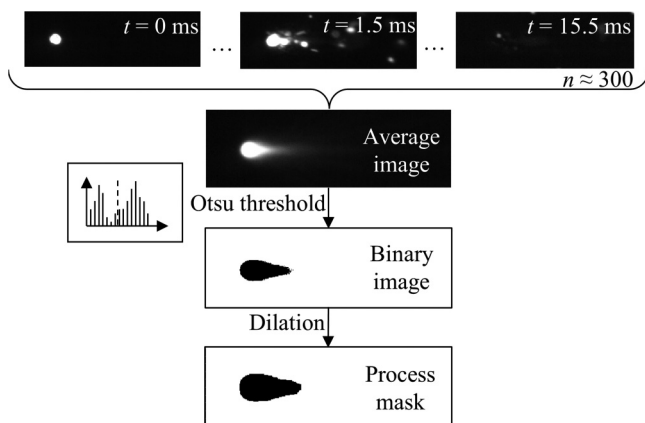


FIG. 3. Approach to the calculation of a process mask.

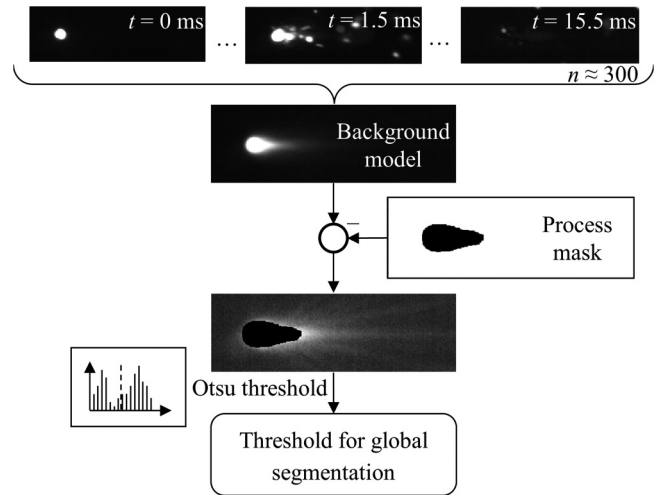


FIG. 4. Approach to the calculation of a threshold for global segmentation.

no direct correlation between the introduced dimensionless enthalpy and the amount of spatter per millimeter single-track length can be derived. However, some of the selected process parameter sets result in track surfaces that exhibit balling. These process parameters result in significantly higher spatter formation than parameter sets that result in regular smooth surfaces of the single tracks. One reason may be the presence, respectively, the intensity of Marangoni flows in the melt pool. These flows are caused by temperature differences, the corresponding surface tensions in the melt pool, and typically flow from the hot melt pool area within the laser spot backward to the cold tail of the melt pool.¹¹ Depending on the stability and velocity of the melt, the material either solidifies at the melt pool tail, flows forward again toward the laser spot, or is ejected from the melt pool as spatter.¹¹ When single tracks show balling characteristics, the weld is broken

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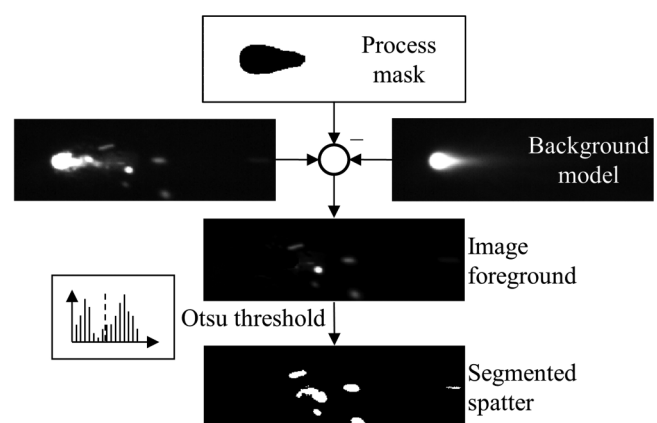


FIG. 5. Approach to the spatter segmentation.

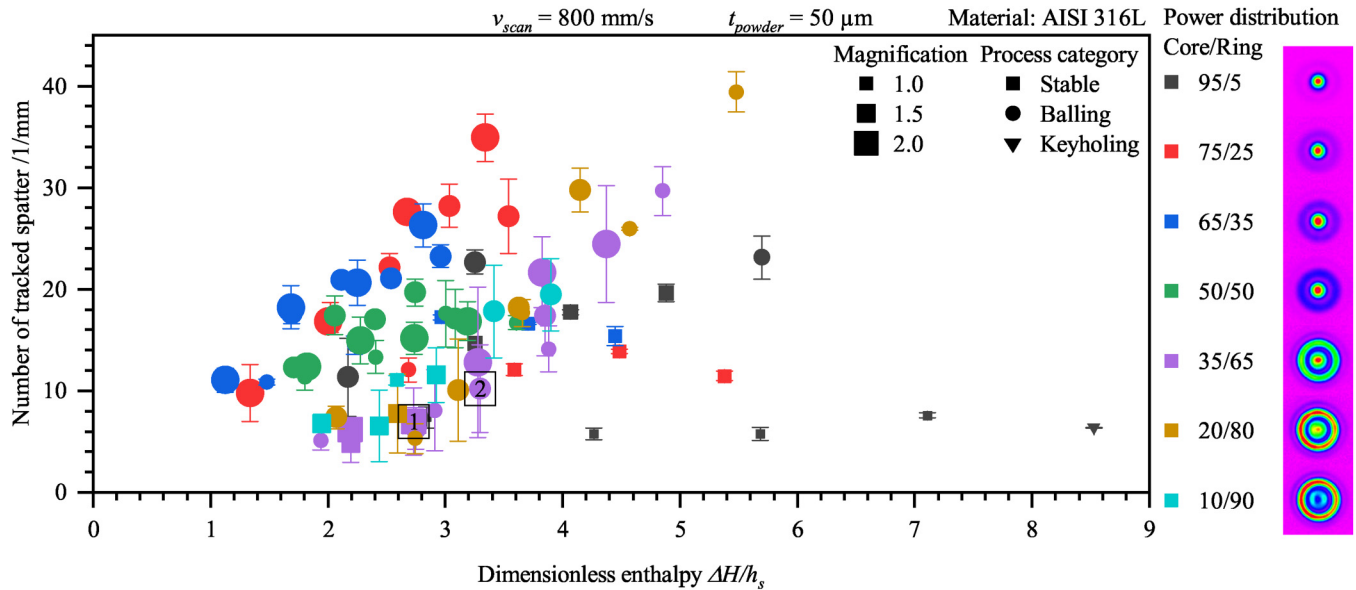


FIG. 6. Number of tracked spatter per mm single-track length as a function of the dimensionless enthalpy.

due to plateau-Rayleigh instability, and individual melt accumulations (balls) have formed.³⁹ This effect is produced by an unstable melt flow into the melt pool's rear area without a corresponding backflow.¹¹ The balling effect can be controlled by adjusting the energy input into the melt pool—usually by adjusting the laser

power to the scan speed.¹¹ With less energy introduced, the melt pool has a lower temperature and, therefore, solidifies earlier, preventing interruption of the melt track induced by surface tension.⁴⁰ Along with the balling tendency, material vaporization increases with increased energy introduced into the material. As the metal

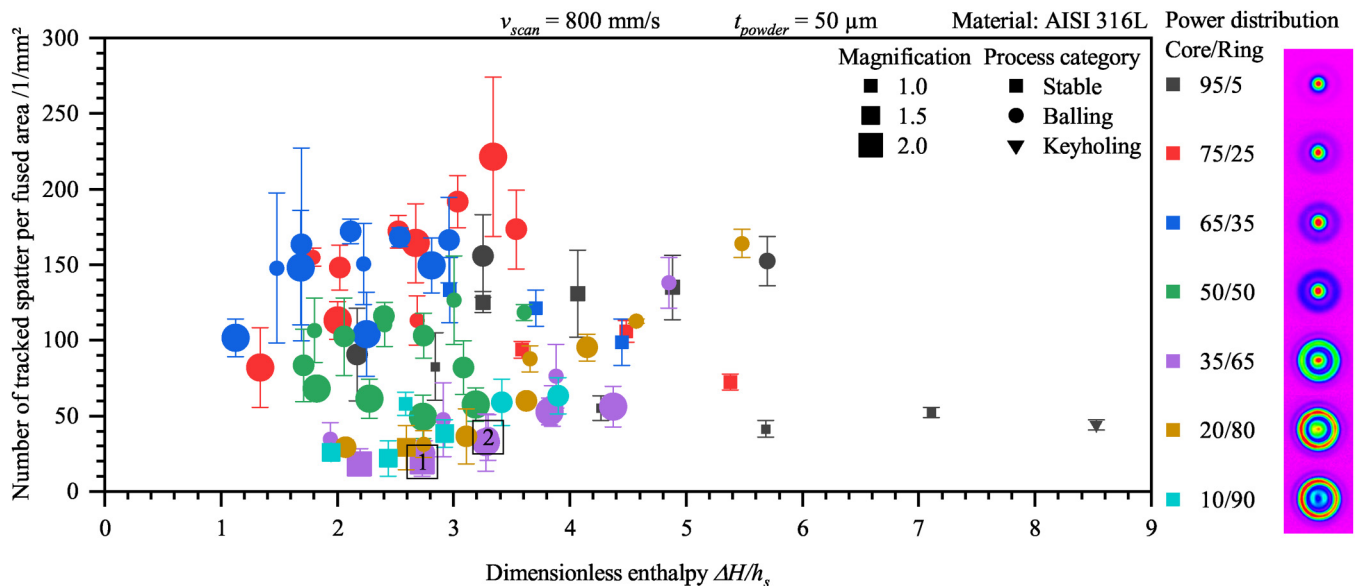


FIG. 7. Top view of a single track with corresponding fused area.

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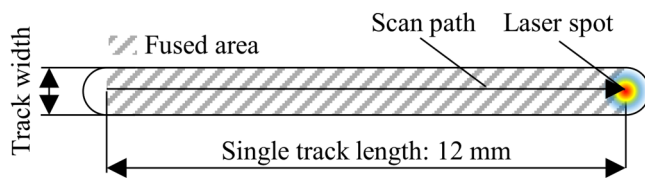


FIG. 8. Number of tracked spatter per fused area as a function of the dimensionless enthalpy.

vapor flow increases, more powder particles are drawn in the laser beam and heated. These particles are also detected as hot spatter, resulting in more tracked spatter. Remarkably, the high spatter numbers and balling tracks in the experiments occur at comparatively low dimensionless enthalpies. The manufacturing of parts with high densities ($>99.6\%$) using Gaussian intensity distributions is reported in a range of dimensionless enthalpy $\Delta H/h_s$ between 3.4 (Ref. 41) and 11.⁴² In the case of having the effect at low dimensionless enthalpy, it can also be caused by insufficient bonding to the substrate and, consequently, insufficient heat dissipation.

The theories described provide an approach to explain why the balling parameter sets also result in an increased number of spatter. Overall, higher power fractions in the ring result in a lower number of spatter per millimeter track length, except the almost Gaussian “95/5” beam profile. The “75/25” and “65/35” beam profiles have the highest number of spatter. The beam profiles “35/65”, “20/80”, and “10/90” have the lowest number of spatter. The “50/50” beam profile lies in between. However, depending on the selected process parameters (laser power, scan speed, beam size,

and beam profile), the single tracks have different widths. Depending on the single-track width, larger or smaller hatch distances can be selected in the three-dimensional process. Large hatch spacing can be selected for large single-track widths, reducing the number of single tracks required for an exposed area. Therefore, the number of tracked spatter is related to the fused area in Fig. 7. The fused area refers to the top view and is calculated by multiplying the single-track length of 12 mm and the track width (see Fig. 8). The widths data are taken from cross sections. With a larger beam size and more power in the ring, the track widths get larger. This reduces the number of hatched tracks per area for beam profiles with a high power fraction in the ring. Beam profiles “35/65”, “20/80”, and “10/90” have on average significantly less spatter per fused area than the beam profiles “95/5”, “75/25”, and “65/35”. The beam profile “50/50” with an average tracked spatter of approximately 90 1/mm^2 is in a transition range. The dependence of the amount of spatter per fused area on the beam profiles could be attributed to the peak intensities inherent in each profile. The beam profiles “35/65”, “20/80”, and “10/90” show a significantly reduced peak intensity compared to beam profiles “95/5”, “75/25”, “65/35”, and “50/50”. The lower peak intensity in the center, coupled with the higher power in the ring, is expected to result in a decrease in the peak temperature of the melt pool, thereby reducing the temperature gradient.⁴³ Consequently, there are two implications: First, the weakened outflow of metal vapor leads to reduced entrainment of powder particles in the gas jet. Second, the diminished temperature gradient potentially mitigates the intensity of Marangoni flows, resulting in a decrease in the number of melt pool ejecta. However, for a comprehensive assessment of the gas jet, it is crucial to evaluate not only the quantity of the spatter but also its velocity and size.

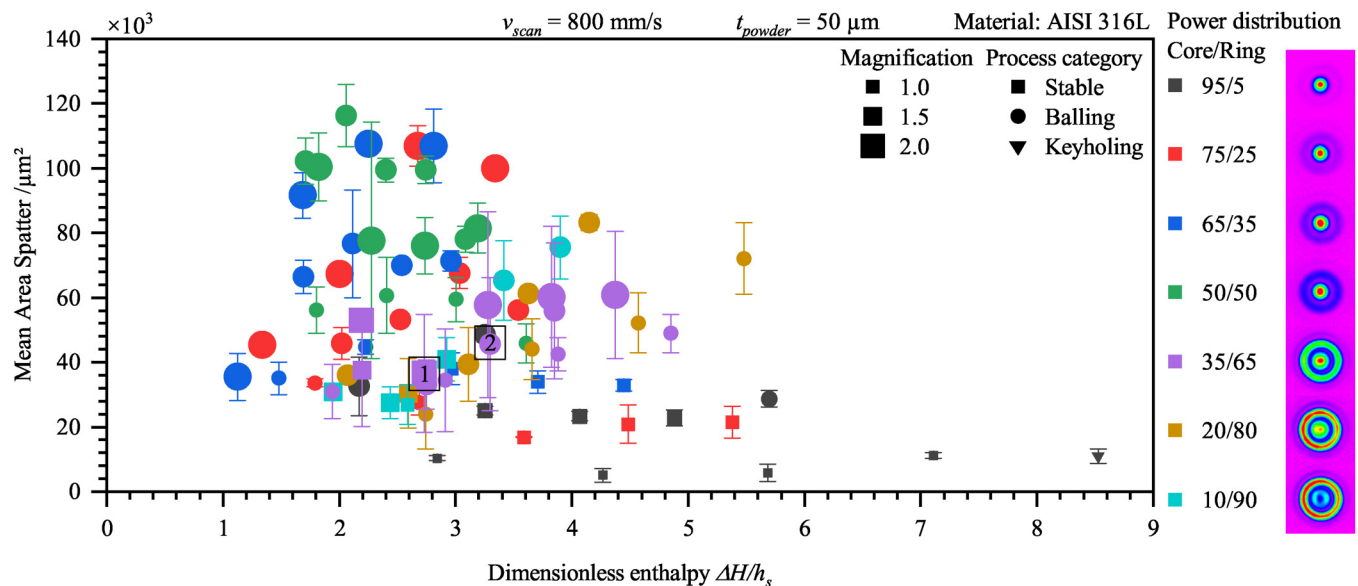


FIG. 9. Mean spatter area as a function of the dimensionless enthalpy.

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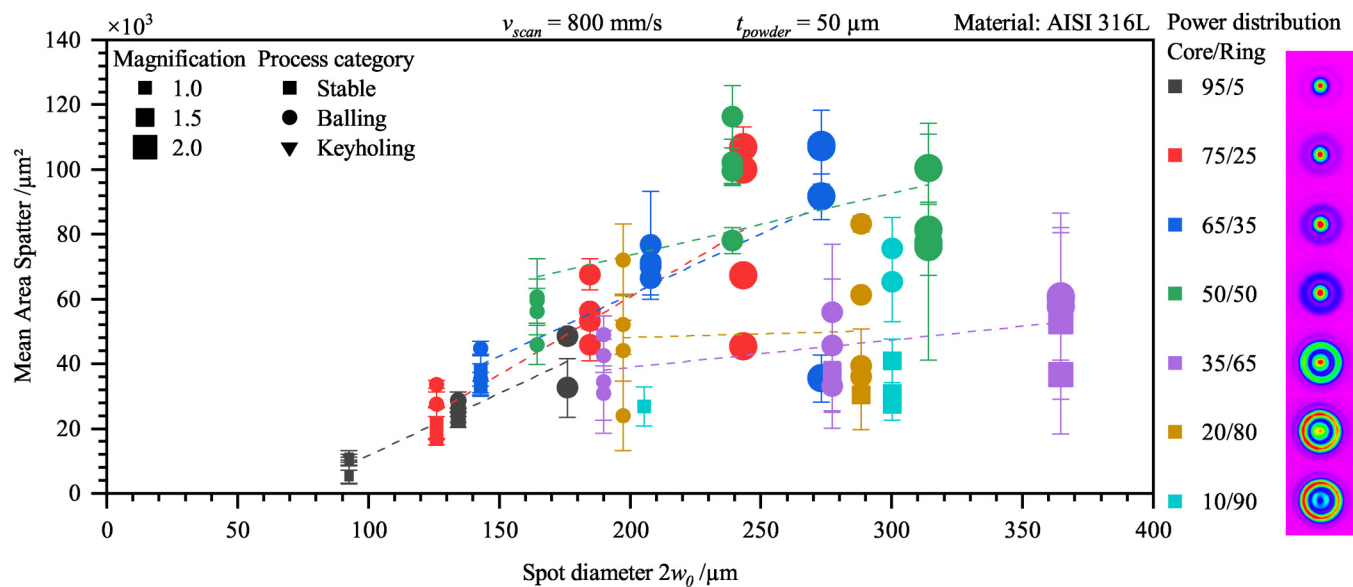


FIG. 10. Mean spatter area as a function of the spot radius.

B. Influence on the spatter size

The mean spatter area as a function of the dimensionless enthalpy is illustrated in Fig. 9. When using the nearly Gaussian beam profile “95/5”, the smallest spatter particles are observed. As the power distribution to the surrounding ring increases, the average spatter sizes increase until a power distribution of “50/50”. With a further increase of the ring power up to “10/90”, the spatter sizes tend to decrease again. Notably, parameter sets resulting in smooth single-track surfaces exhibit significantly smaller spatter compared to those resulting in balling. Furthermore, the magnification, and, thus, the spot size tend to influence the spatter size. Larger magnification values result in larger spatter. Figure 10 presents the spatter sizes as a function of spot sizes. For identical beam profiles, the mean area of the spatter increases with the spot size. In Fig. 10, linear fits for constant beam profiles are drawn. For the beam profiles “95/5”, “75/25”, and “65/35”, the fit functions and their slopes are in a similar range. From beam profile “50/50” on, the slope for beam profiles “35/65” and “20/80” decreases with increasing power fraction in the ring. Due to the small number of parameter sets for beam profile “10/90”, no fit function is available for this profile.

One possible explanation for the variation in spatter sizes between stable and balling processes may lie in the occurring type of spatter. In a stable process, the detected spatter primarily consists of hot powder spatter.⁹ These particles retain their original powder particle diameter when they do not agglomerate. Melt pool ejecta typically exhibit larger dimensions compared to the used powder. Wang *et al.*⁶ reported that spatter diameters were 3.5 times larger than the original powder particle size. Therefore, the dependence of the spatter size on beam size can be attributed to the scaling of melting phenomena. With larger spot sizes, the

corresponding increase in melt pool size scales the flows within the volume, resulting in larger spatter. Beam profiles “20/80” and “10/90” have a significantly reduced peak intensity to mean intensity ratio compared to all other beam profiles. This characteristic can mitigate the temperature gradients, thereby reducing the Marangoni flows and resulting in smaller average spatter sizes due to decreased presence of melt pool ejecta.

C. Influence on the spatter velocity

In Fig. 11, the average velocity of the tracked spatter is presented as a function of dimensionless enthalpy. Remarkably, the spatter velocities are similar across all beam profiles at constant dimensionless enthalpy. Tumkur *et al.*²⁷ suggest that the high velocity of hot spatter indicates high recoil pressures. In the present context, this implies that the mean vapor pressure depends solely on the dimensionless enthalpy, reflecting the mean energy input during spot crossing. Consequently, the beam profile demonstrates no discernible influence.

Several of the described influences on the spatter characteristics can be exemplified by two parameter sets (black squares in Figs. 6, 7, 9, and 11), which differ only in the laser power and spot size: parameter set 1 (750 W, 365 μm) and parameter set 2 (600 W, 277 μm). Parameter set 1 results in a lower number of spatter due to less energy introduced, resulting in less metal vapor and less melt pool ejecta (see Figs. 6 and 7). For the same reason, the average spatter velocity of parameter set 1 is slightly lower (see Fig. 11). Because parameter set 2 produces balling effects, the average spatter size is larger than for parameter set 1 (see Fig. 9) due to a higher portion of melt pool ejecta compared to powder spatter.

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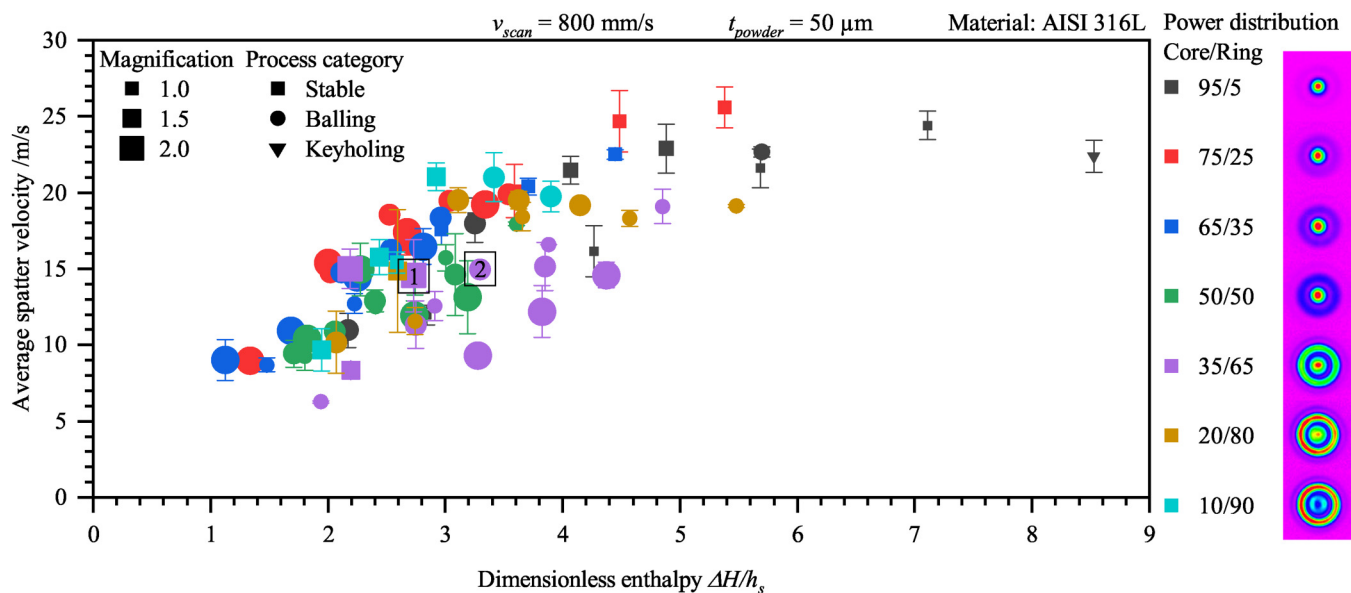


FIG. 11. Average spatter velocity as a function of the dimensionless enthalpy.

IV. CONCLUSION

In this study, an investigation of various hot spatter characteristics during PBF-LB/M of AISI 316L was conducted. The experimental setup involved the utilization of seven beam profiles with different spot sizes and a variation in process parameters. High-speed on-axis camera images were recorded and evaluated using various image processing algorithms for spatter evaluation. The key findings can be summarized as follows.

- (1) The number of hot spatter relative to the fused area primarily depends on the beam profile. Beam profiles with high peak intensities result in a significantly higher amount of spatter per fused area than beam profiles with lower peak intensities. In this investigation, a factor of 3–4 is observed for constant dimensionless enthalpies.
- (2) The spatter size is influenced by two main factors: the spot size and the beam profile. Larger spot sizes generally result in larger spatter sizes, while beam profiles with lower peak intensities tend to generate smaller spatter sizes. The hypothesis is that the origin of spatter changes with the beam profile. Due to the lower peak intensity, beam profiles characterized by less power in the center and more power in the ring exhibit a significant reduction in ejections from the melt pool.
- (3) The spatter velocity is primarily influenced by the dimensionless enthalpy introduced. The beam profile and its corresponding peak intensity have a minor influence. High dimensionless enthalpies result in faster spatter, which indicates high recoil pressure and high process dynamics, limiting the process robustness.
- (4) Across all spatter characteristics examined (number, size, and velocity), single tracks with balling surfaces consistently yield

higher characteristic values than those resulting in smooth surfaces.

In future work, validating the hypotheses made, e.g., using a Smooth Particle Hydrodynamics (SPH) simulation, will be necessary.

AUTHOR DECLARATIONS

Conflict of Interest

The authors have no conflicts to disclose.

Author Contributions

Jonas Grünwald: Conceptualization (lead); Data curation (lead); Formal analysis (lead); Funding acquisition (supporting); Investigation (lead); Methodology (lead); Software (lead); Validation (lead); Visualization (lead); Writing – original draft (lead); Writing – review & editing (lead). **Jan Reimann:** Conceptualization (supporting); Data curation (supporting); Investigation (supporting); Software (supporting); Writing – review & editing (supporting). **Katrin Wudy:** Conceptualization (supporting); Resources (lead); Supervision (lead); Writing – review & editing (supporting).

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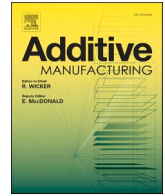
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Tailored microstructures in laser-based powder bed fusion of 316L using beam shaping: From microstructure formation in individual specimens to their transition in segmented parts

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ABSTRACT

Ring-shaped beam profiles are increasingly being used in laser-based powder bed fusion of metals (PBF-LB/M). Due to a changed temperature distribution and progression during processing, the ring-shaped beam profiles have the potential to adjust the solidification behavior and can therefore affect the microstructure formation. This publication shows first how the use of different ring-shaped beam profiles influences the as-built microstructure of AISI 316L. Secondly, it demonstrates how this can be used to tailor the microstructure locally. In order to investigate the emerging microstructure in the transition zone of locally different microstructures, two parameter sets are combined to manufacture different segments within a single component. The results show that using state-of-the-art strategies with Gaussian beam profiles can result in strong $\langle 100 \rangle$ || BD textures, like reported in the state of the art. In comparison, it is possible to rotate the texture to $\langle 111 \rangle$ || BD when using ring-shaped beam profiles. Moreover, a correlation between the grain size and the spot size of the beam profile used is employed. When combining two parameter sets within one part, the height of the transition zone and the emerging microstructure in subsequently manufactured layers strongly depend on the microstructure of previously manufactured layers. If a fine-grained microstructure is initially formed, only a small transition zone (< 5 layers) is created. Resulting grains in subsequent layers tend to be smaller. The resulting grains in the subsequent manufactured layers are significantly larger, and a long transition zone (≈ 30 layers) is created if the microstructure of the previously manufactured layers is coarse. The results demonstrate that microstructure tailoring is possible using alternative beam profiles. However, for precise control, the microstructure of the underlying layers must be considered in the design. The systematic investigation of the transition zone between two distinct microstructures is a key novelty of this study. The newly revealed findings on the interaction between initially formed and subsequently developing microstructures can also be applied to further additive manufacturing and joining processes.

1. Introduction

Laser-based powder bed fusion of metals (PBF-LB/M) refers to an additive manufacturing process in which powder is deposited layer by layer within a powder bed and fused with the substrate or underlying layers. A laser beam is applied to provide the energy required to melt the powder feedstock. The current state of the art uses single-mode laser beam sources with Gaussian intensity distributions. In standard PBF-LB/

M systems, these have spot sizes between 30 and 115 μm [1].

In research and, gradually, in industry, ring-shaped intensity distributions of the laser are being used more frequently. Various processing advantages are attributed to these ring-shaped intensities. These include a larger process window [2], spatter reduction [3,4], as well as partial process acceleration [5], and the ability to process materials that are challenging to process using Gaussian laser beams (e.g., due to the susceptibility to cracking, such as CM247, IN738, ...) [6,7].

During the state-of-the-art PBF-LB/M process, small melt pools travel

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Nomenclature

A	Absorption
c_p	Specific heat capacity
E_V	Volume energy density
h	Hatch distance
h^*	Dimensionless hatch distance
h_s	Enthalpy at melting point
L_f	Latent heat of fusion
Pe	Peclet number
P_{laser}	Laser power
T_0	Ambient temperature
T_m	Melting temperature
v_{scan}	Scanning speed
w_0	Spot radius
α	Thermal diffusivity
δ_{layer}	Layer thickness
ρ	Density
φ_{rot}	Scan rotation angle between layers
$\Delta H/h_s$	Dimensionless enthalpy

through the component at high speeds. This results in high local temperatures, temperature gradients, and cooling rates. The fast time scales on small spatial length scales result in typical process characteristics PBF-LB/M is well known for, such as the rapid solidification and characteristic melt pool shapes [8]. These process characteristics determine process and component properties, including the as-built microstructure of manufactured samples.

Typically, metals solidify as polycrystals during PBF-LB/M. The microstructure of these polycrystalline metals mainly consists of two scales (see Fig. 1a). The larger scale is represented by grains, while the crystal lattice forms the smaller scale. Grains are divided from each other by grain boundaries, which are developed in places where the misorientation angle of adjacent sub-microstructures, i.e., the crystal lattice, exceeds a specific value [9]. Within one grain, the crystal lattice is built up by material-dependent unit cells, which can have different orientations (see Fig. 1b). The resulting crystallographic texture strongly influences the macroscopic properties of the material, such as the yield strength and Young's modulus [10].

In as-built condition, the microstructure of additively manufactured components is predominantly governed by the remaining solidification microstructure of single melt tracks. This remaining microstructure depends on the initial solidification, the partial remelting, and the further alteration through the reheating cycles throughout the build process [11]. The boundary conditions comprise the geometric shape of the solid/liquid phase boundary, as well as the constellation of the solid

phase present at the phase boundary and the solidifying phase of the melt. For the PBF-LB/M manufacturing of AISI 316L used in this study, the idealized assumption of austenite being the present solid phase and also the solidifying phase is warranted [8,11]. As such, nucleation on the phase boundary is not required, and the existing grains with renewed access to the melt may grow further, leading to epitaxial grain growth. With the face-centered cubic (fcc) crystal structure of AISI 316L, the preferred growth direction is in the $\langle 100 \rangle$ directions of the crystal lattice [9]. This means, in single melt tracks, $\langle 100 \rangle$ columnar grains grow along the thermal gradient perpendicular to the melt pool boundary. Due to the partial remelting of previously fabricated melt tracks and layers during PBF-LB/M, the columnar grains beneath incrementally grow, leading to the distinctive microstructure comprised of long columnar grains growing through multiple layers [12,13]. This can result in a $\langle 100 \rangle$ texture in the build direction (BD) [44]. The underlying reason is that the alternating exposure direction of successive melt tracks induces differing temperature gradients, which, for a sufficient number of layers, leads to the survival of grains oriented in the direction of the average temperature gradient parallel to BD [15].

By adjusting the scanning strategy, the results can be varied [13,16] due to a changed temperature distribution and the temperature progression, which influence the resulting microstructure. For example, if the process parameters are chosen in a way that grains with their $\langle 100 \rangle$ direction oriented 45° towards BD, which is usually the case for grains on the sides of one melt track, are not remelted, it has been reported that the $\langle 110 \rangle$ direction is the preferred texture in BD [14,17]. Whereby the orientation to the scan direction can be broken by rotating scan strategies (e.g., 67° rotation between the layers), resulting in a fiber texture with $\langle 110 \rangle$ orientation to BD [18].

Besides the scanning strategy, alternative beam profiles are another method for adjusting the temperature distribution and its progression. These no longer provide the process energy in a Gaussian shape, but can deviate from it and thus influence the heating and cooling as well as the solidification process. This results in a change of the melt pool cross-sectional shape when applying non-Gaussian beam profiles while producing single melt tracks [5,19] and three-dimensional specimens [20, 21]. For example, melt tracks tend to be wider when produced with ring-shaped instead of Gaussian beam profiles [5,19]. As a result, processes with alternative beam profiles can produce samples with distinct microstructure characteristics. Galbusera et al. [20] found the beam profile to affect the melt pool depth during processing AlSi7Mg0.6. Beam profiles with a higher peak intensity resulted in deeper melt pools, correlating with a higher ultimate tensile strength and elongation at break. Additionally, the same research group found that the grain aspect ratio of a low-Si alloyed steel can be adjusted using different beam profiles [21]. Roehling et al. [22] showed that using elliptical beam profiles changes the grain nucleation and, correspondingly, the grain morphology towards an equiaxed shape when processing AISI 316L. The propensity of the traversal elliptical laser beam to form smaller, more equiaxed grains was attributed to lower thermal gradients. However, it was noted that the scan speeds of < 200 mm/s in this study do not reflect the typically higher scanning speeds of around 1000 mm/s used in the industry. A similar trend was observed by Bremer et al. [23] in powder-based directed energy deposition of an austenitic steel with a ring-shaped beam profile. Compared to using a top hat profile, using a ring-shaped beam profile with the same process parameters resulted in a finer-grained and more equiaxed microstructure. These observations correlate with a lower cooling rate and the ratio of the thermal gradient and solidification rate measured [23] and simulated [24] when using the ring-shaped beam profile.

In addition to changes in the grain size and shape, the texture of additively manufactured components can be adjusted using alternative beam profiles. Pérez-Ruiz et al. [25] demonstrated that ring-shaped beam profiles can imprint a $\langle 111 \rangle \parallel$ BD texture on the microstructure when processing IN718. The key to this texture shift was traced back to the change in melt pool curvature when using ring profiles

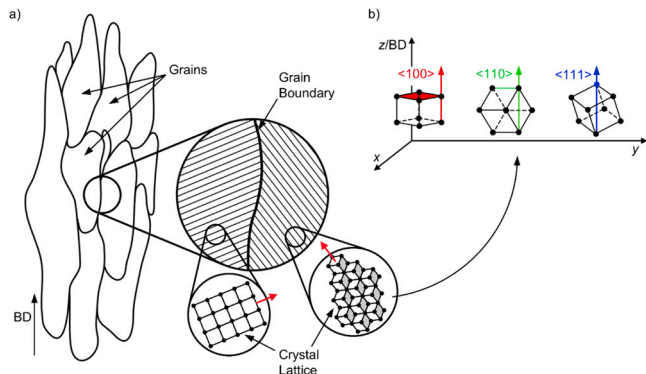


Fig. 1. Schematic representation of the microstructure in a metal, a) on a large scale and b) on a sub-microstructural level.

compared to a Gaussian laser beam. In contrast, Mirzabeigi et al. [26] found that ring-shaped beam profiles produce a fiber-textured microstructure with a $\langle 100 \rangle$ orientation along BD in IN718. Both studies used hatch distances in the range of the beam radius and identical scanning speeds. The main differences between the two studies are the spot size and the rotation angle of the scan pattern between layers. While Pérez-Ruiz et al. [25] used spot sizes in the order of 140 μm and no rotation of the scan pattern, Mirzabeigi et al. [26] applied spot sizes in the range of 400 μm and rotated the scan patterns between layers by 67° . Due to the small spots and the lack of rotation in Pérez-Ruiz et al. [25], strong textures can form, which are mainly influenced by the spatial position of the solidification front. The large spots used by Mirzabeigi et al. [26] presumably lead to very long melt pools with a preferred $\langle 100 \rangle$ crystal growth direction close to the build direction. Combined with a 67° rotation of the scan pattern, the observed fiber texture forms. However, it is not entirely clear how the results of Pérez-Ruiz et al. [25] would change if a scan rotation angle were used.

The cited studies demonstrate that alternative beam profiles can modify the as-built microstructure. However, most studies use the same absolute process parameters with different beam profiles and spot sizes and compare the process results based on process variables such as the dimensioned volume energy density. In [27], it was shown that dimensionless parameters are suitable for ensuring comparability between different process parameter sets with beam profiles of various sizes. Therefore, this study examines comparable parameter sets based on dimensionless parameters. Moreover, an interesting application based on the known influence of alternative beam profiles on the microstructure is to enable the manufacturing of load-case-based numerically optimized microstructures [28]. Local tailoring of the microstructure is crucial for this application. Therefore, it is essential to understand the influence of alternative beam profiles on the resulting microstructures in both large volumes and the transition zone between regions manufactured with different beam profiles. For this fundamental understanding, this paper uses a hierarchical approach, investigating first the microstructure of components manufactured with various beam shapes but comparable parameter sets. In the second step, two beam shapes, resulting in distinct microstructures, are selected to produce different segments within one component, the transition zone of which is analyzed. The current state of the art lacks a targeted investigation of the transition zone between two segments with locally tailored microstructures. Thus, this paper contributes to filling the corresponding research gap.

2. Materials and methods

This section first presents the experimental setup for the investigations, followed by the material used and the parameter sets investigated. Subsequently, the evaluation methods for the microstructure evaluation are described.

2.1. Experimental setup

An in-house developed experimental test bench is used to produce the specimens. Various beam profiles are generated using a fiber laser with switchable beam profiles (AFX-1000, nLIGHT, Inc., Vancouver, WA, USA). The investigated beam profiles are ring-shaped distributions that differ in the power distribution between a central Gaussian and a surrounding ring. According to this power distribution, the beam profiles are named “power fraction core/power fraction ring”. A four-axis scanning system (AM-MODULE NEXT GEN, RAYLASE GmbH, Wessling, Germany) is used for laser beam deflection, focusing, and expansion. The build process of the specimens takes place within a self-built process chamber including all modules of an industrial PBF-LB/M system, such as a supply and a build platform, an overflow and a shielding gas flow. Argon is used as a shielding gas. The residual oxygen content is kept below 0.15 %.

2.2. Design of experiments

All experiments are carried out on the material AISI 316L (LPW-316-1, LPW Technology Ltd., Runcorn, Cheshire, United Kingdom). This study aims to gain a fundamental understanding of the influence of alternative beam profiles on the microstructure formation of fcc materials. Fundamental findings regarding the microstructure at the grain and texture level should also be transferable to other fcc materials (such as other austenitic steels or nickel-based alloys).

The process parameter selection is based on a preliminary study [27] which includes the introduction of dimensionless process windows. The dimensionless enthalpy $\Delta H/h_s$, dimensionless hatch h^* , and Peclet number Pe are used for these process maps. These parameters are calculated according to

$$\frac{\Delta H}{h_s} = \frac{A \cdot P_{\text{laser}}}{\rho \cdot (c_p \cdot (T_m - T_0) + L_f) \cdot \sqrt{\pi \cdot \alpha \cdot v_{\text{scan}} \cdot (2w_0)^3}} \quad (1)$$

$$h^* = \frac{h}{w_0} \quad (2)$$

$$Pe = \frac{v_{\text{scan}} \cdot 2w_0}{\alpha} \quad (3)$$

with absorption A , laser power P_{laser} , density ρ , specific heat capacity c_p , the delta between melting and ambient temperature $T_m - T_0$, latent heat of fusion L_f , thermal diffusivity α , scanning speed v_{scan} , laser spot radius w_0 , and hatch distance h . Although the PBF-LB/M process is highly dynamic, with changing material properties and absorption behavior [29, 30], a simplified steady state process with constant material properties common in several scaling laws [31] is assumed for the design of experiments. This dimensionless approach provides a means of calculating a dimensionless temperature field for the process zone, normalized to the beam diameter used and the thermophysical properties of the material being processed [32]. In this way, the approach assumes that similar dimensionless parameters result in a similar dimensionless temperature field [27], resulting in solidification conditions (growth rate and thermal gradient values) scaled to the beam size. This allows the investigations to consider the influences of beam size and shape separately. For calculating the dimensionless parameters, the recommended thermophysical properties from [33] are used (see Table 2). Analogous to [34], a constant absorption of 0.6 is assumed, as experimentally verified by King et al. [35].

In order to examine comparable parameter sets for the different beam profiles, the dimensionless enthalpy, dimensionless hatch, and Peclet number were kept in a similar range. In the first step, 13 cuboid samples are manufactured with a constant parameter set, respectively. At least one parameter set is examined for each beam profile (see Table 1). As shown in Table 1, the specimens can be divided into two ranges of dimensionless hatch ($h^* \approx 1.3$ – 1.4 and $h^* \approx 1.6$ – 1.7). This way, the influence of the remelted cross-sectional area is investigated. In the second step, two parameter sets are combined to manufacture different segments within one specimen. The parameter sets used for these segmented specimens are marked in bold in Table 1. Regardless of the type of specimen, the samples are exposed using a bidirectional hatching with a rotation angle between layers φ_{rot} of 67° , industrially established to avoid anisotropies in the x-y plane [36].

2.3. Specimen design

Two types of specimens are produced. In order to investigate the influence of the beam profile on the microstructure that develops in a larger volume, specimens with one constant parameter set are manufactured. These are cuboid specimens with the dimensions $7 \times 7 \times 10 \text{ mm}^3$ (see Fig. 2a), where the 10 mm height consists of 7 mm sample height and 3 mm cutting allowance. To investigate the transition of

Table 1
Parameter sets investigated.

Power distribution core/ring	Magnification	Spot diameter (2 nd moment) $2w_0$ in μm	Laser power P_{laser} in W	Scanning speed v_{scan} in mm/s	Scan rotation angle ϕ_{rot}	Dimensionless enthalpy $\Delta H/h_s$	Hatch distance h in μm	
							$h^* = 1.3\text{--}1.4$	$h^* = 1.6\text{--}1.7$
	1	92	200	800	67°	3.8	60/65	80
	2	175	500	500	67°	4.7		120/140
	1	125	250	800	67°	3.0		100
	1	142	300	800	67°	3.0	90	110
	1	163	400	800	67°	3.3	110	
	1	189	400	500	67°	3.3	130	160
	1	197	500	500	67°	3.9	140	170
	1	205	500	500	67°	3.7	140	170

Table 2
Thermophysical properties of AISI 316L according to [33].

Name	Value	
Ambient temperature T_0	293.15	K
Density ρ	7800	kg/m ³
Melting temperature T_m	1693	K
Specific heat capacity c_p at T_m	592.55	J/(kg•K)
Thermal diffusivity α at T_m	$6.44 \cdot 10^{-6}$	m ² /s

stages (#180, #320, #800, #1200) and polished in two steps (3 μm and 1 μm). Subsequently, chemical etching with V2A etchant at 50 °C reveals the melt track geometries. The etched sample cross-sections are examined using a light optical microscope (DM6 M, Leica Microsystems GmbH, Wetzlar, Germany). Following, the samples are polished in three stages (3 μm , 1 μm , and 0.25 μm) to remove the etched structure. Finally, electron backscatter diffraction (EBSD) data are measured (JSM7600, JEOL Ltd., Akishima, Japan) with an Oxford Symmetry EBSD detector (Oxford Instruments plc, Abingdon, United Kingdom). For the specimens produced using a single parameter set, an area of approximately $0.5 \times 2 \text{ mm}^2$ (width $\times$ height) in the center of the sample is scanned with a step size of 2 μm . The image height corresponds to approximately 40 layer heights. These measurements reconstruct an average of approximately 700 grains, with the measurement with the lowest grain number showing slightly more than 200 grains, which limits the statistical significance but is sufficient for the comparative nature of the analysis. The comparability of the sections is verified by several images of different areas on multiple samples. Within the segmented specimen, an area of approximately $3 \times 5 \text{ mm}^2$ is scanned using a step size of 4 μm , resulting in a minimum number of 2000 grains. All EBSD analyses exclusively consider measurement points below a mean angular deviation of 1° and grains above a size of 5 pixels. Grain boundaries with a mismatch angle $> 5^\circ$ are marked in black in the inverse pole figure (IPF) maps.

From the EBSD data, the crystallographic texture, grain size as the equivalent grain radius (radius of a circle with the area of the grain), and the grain morphology as the mean grain shape aspect ratio (maximum expansion of the grain in relation to the minimum expansion of the grain) are extracted. To quantify the size of the transition zone of the segmented samples (see Fig. 2b), the area fractions of the grains intersected by the beam switching plane are determined. As schematically shown in Fig. 3, this is done by selecting the grains intersected by the beam profile switching plane (see grains with yellow hatching at 45° in Fig. 3). The area of these grains $A_{\text{intersected}}$ is related to the area of all existing grains $A_{\text{intersected}} + A_{\text{new}}$. Accordingly, the area fraction of the cut grains $A_{\text{intersected},\%}$ is calculated as

$$A_{\text{intersected},\%} = \frac{A_{\text{intersected}}}{A_{\text{intersected}} + A_{\text{new}}} \quad (4)$$

This area fraction is evaluated in 16 μm steps at various distances from the beam profile switching plane of the beam switching. An area

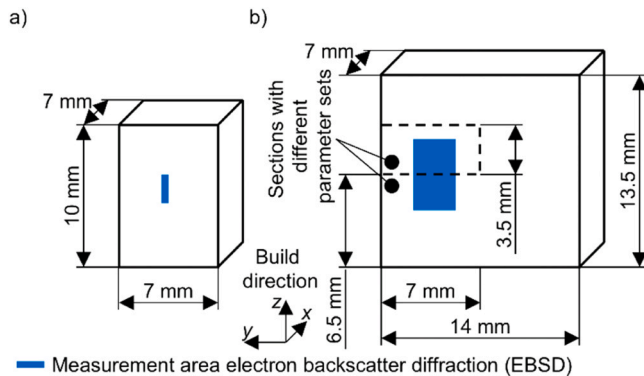


Fig. 2. Drawing of the test specimens used, including the outlined measuring range of the electron backscatter diffraction (EBSD) measurements of a) single-parameter set specimen and b) segmented specimen manufactured with two combined parameter sets.

locally tailored microstructures, cuboid specimens are built consisting of two different segments with outer dimensions of $14 \times 7 \times 13.5 \text{ mm}^3$ (see Fig. 2b). Each segment has a scanned area of $7 \times 7 \text{ mm}^2$ to transfer the results of the samples from Fig. 2a. The height is derived from three times the height of a segment of 3.5 mm and a 3 mm cutting allowance.

2.4. Specimen preparation and evaluation methods

To enable EBSD measurements, the specimens are embedded in a conductive carbon-based polymer. The samples are ground in four

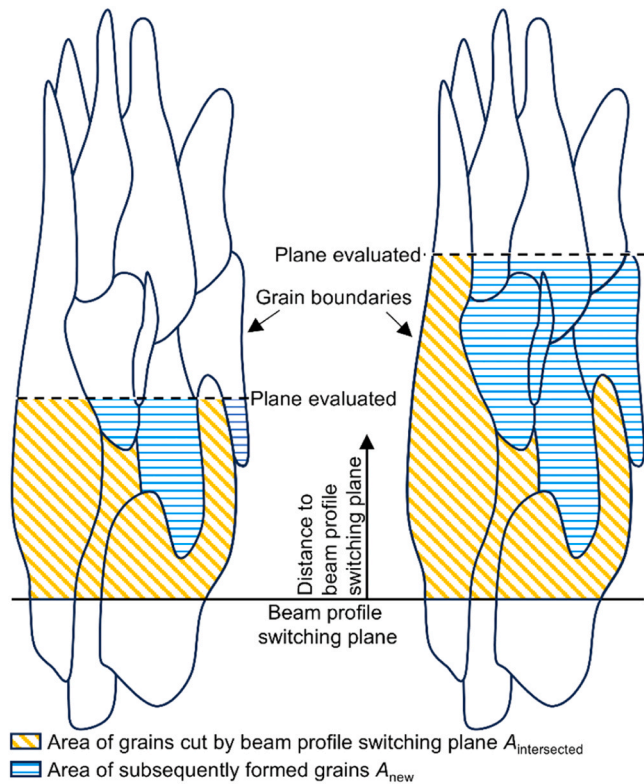


Fig. 3. Schematic illustration of the evaluation method for determining the size of the transition zone between two segments.

fraction of 50 % is defined as the numerical value for the height of the transition zone. All evaluations use the open-source MATLAB toolbox MTEX (Version 5.11.2).

3. Results and discussion

The following section presents the influence of the beam profile and the dimensionless hatch distance on the as-built microstructure. The second step demonstrates how the microstructure varies depending on the microstructure of initially manufactured layers. While not explicitly addressed in this study, it should be noted that all parameter sets examined resulted in densities of at least $99.94\% \pm 0.02\%$.

3.1. Influence of the beam profile and the dimensionless hatch on the as-built microstructure

This section presents and discusses the microstructure analyses for single-parameter specimens. Since it is well-known that the remelted fraction of the individual melt tracks influences the resulting microstructure [17], the analyses are divided into two stages of dimensionless hatching ($h^* = 1.3$ – 1.4 and $h^* = 1.6$ – 1.7), whereby a smaller dimensionless hatch results in a larger overlap area of the individual melt tracks.

Fig. 4 shows sections of the IPF maps of the samples produced using different beam profiles and a similar dimensionless hatch h^* between 1.3 and 1.4, as well as the corresponding IPF parallel to the build direction (BD/z) and the y axis. The orientation of the crystal structure parallel to BD was used for the color code (z-IPF). The columnar grain growth typical for PBF-LB/M is visible in all IPF maps. Of particular interest are the IPFs, which show that the preferred orientation is rotated when changing the power distribution between the central Gaussian core and the surrounding ring. Additionally, Fig. 5 shows the fiber volume fraction of the high-symmetry directions $\langle 100 \rangle$, $\langle 110 \rangle$, and $\langle 111 \rangle$ with a tolerance angle of 10° to quantify the texture proportions of the

respective crystal directions in the respective samples. As expected from the literature [14], when using beam profile 95/5, the specimen shows a firm $\langle 100 \rangle \parallel$ BD texture with a volume fraction of approximately 20 % (see Fig. 5). The geometric necessity of the face-centered cubic unit cell of AISI 316L results in an orientation preference parallel to the y direction between $\langle 100 \rangle$ and $\langle 110 \rangle$. If the power is shifted towards the surrounding ring (beam profiles 65/35 and 50/50), this preferred direction is broken up, and no strong textures can be seen parallel to the build direction. The corresponding volume fractions are each less than 10 %. The texture parallel to the y direction is similar for the beam profiles 95/5 and 65/35. With beam profile 50/50, no preferred orientation is detectable. When using beam profile 35/65, the preferred orientation parallel to BD is split. The larger fraction (24 %) of the grains is oriented in the $\langle 100 \rangle$ direction, while a minor fraction (14 %) is oriented in the $\langle 111 \rangle$ direction. These orientations parallel to BD also shift the preferred crystallographic orientation in y more towards a $\langle 110 \rangle$ preference. There is a strong correlation between the power fraction of the ring and the emphasis of the $\langle 111 \rangle$ texture parallel to the build direction. The texture becomes more pronounced with an increased power fraction in the ring. When using beam profiles 20/80 and 10/90, the samples have a preferred orientation in the $\langle 111 \rangle$ direction with volume fractions of approximately 25 %.

One possible theory for the observed microstructure is based on the change in the melt pool shape and the corresponding tilting of the solidification front using different beam profiles. Fig. 6a shows that using beam profile 95/5 results in comparatively deep and narrow melt pools. Consequently, the side walls of the melt pool have a steep inclination close to parallel to the BD, enabling the grains to grow from the melt pool bottom and the side walls with the preferred $\langle 100 \rangle$ equivalent orientation. Biswas et al. [37] state that if the misorientation between two adjacent growth directions matches the crystal symmetry, grain growth may continue epitaxially, increasing texture. Since the occurring orientations in a specimen of the beam profile 95/5 are aligned, the highly $\langle 100 \rangle$ textured microstructure results (see Fig. 4, beam profile 95/5).

If some power is introduced into the ring (beam profiles 65/35 and 50/50), the melt pools become slightly wider and the temperature gradient to the sides is reduced. This reduces cooling rates, leading to less preferential orientation in BD while retaining columnar grain structure.

The melt pools have less steep side walls when using high power fractions in the ring (beam profiles 35/65, 20/80, and 10/90). Overall, the radius of curvature of the melt pools at the bottom is significantly larger. Using beam profile 10/90 leads to a maximum melt pool side wall angle of approximately 45° to the x-y plane (see etched microsection in Fig. 6b). It is assumed that the solidification fronts are additionally significantly tilted to the y-z plane (see schematic melt pool shape in Fig. 6b). This favors the growth of $\langle 111 \rangle \parallel$ BD oriented grains, regardless of the scan rotation angles.

The three-dimensional melt pool shape is adopted from a previous study [38], which investigates various simulation approaches for PBF-LB/M. Using a validated high-fidelity multi-physics simulation, the melt pool shapes outlined result for beam shapes and process parameters similar to those used in Fig. 6b, supporting the assumptions made. Additionally, the hypotheses are supported by the observation that for the sample of beam profile 50/50, the z-IPF shows no clear preference, which indicates higher melt pool inclination angles compared to beam profile 95/5, leading to a slightly inclined $\langle 100 \rangle$ direction, which seems to be broken up in z-direction by the 67° hatch rotation. Regardless of the beam profile, the solidification at the bottom of the melt pool directly behind the beam center can continue to grow slightly tilted to the x-y plane in $\langle 100 \rangle$ direction (see Fig. 6b), which explains the small fraction of grains with $\langle 100 \rangle$ orientation for most samples. The large, remelted area of the melt tracks leads to a continuation of mainly $\langle 111 \rangle$ and partially $\langle 100 \rangle$ orientation within a layer. When this is continued over several layers with the scan rotation of 67° , grains

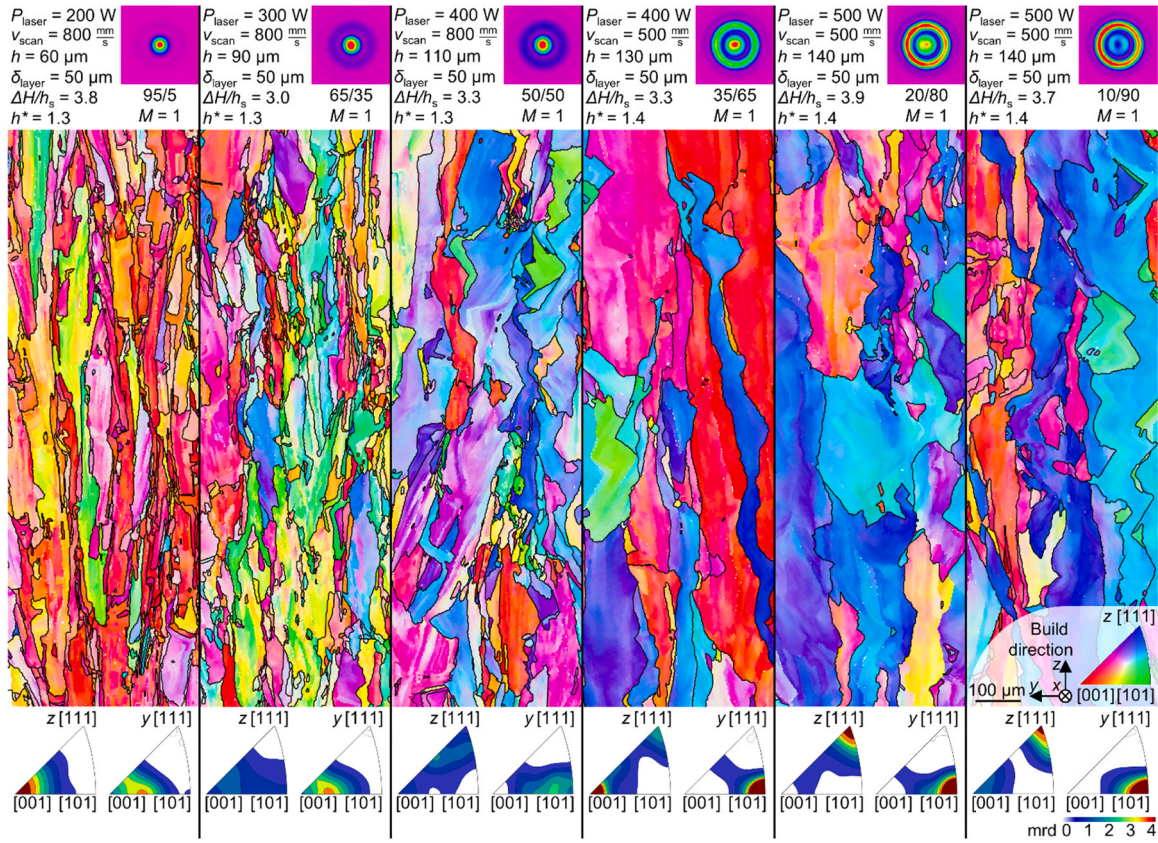


Fig. 4. Sections of the IPF maps and corresponding IPFs of the specimens resulted in a dimensionless hatch h^* between 1.3 and 1.4. The sample produced with the almost Gaussian beam profile 95/5 has a firm $\langle 100 \rangle \parallel \text{BD}$ texture. When using beam profiles 65/35 and 50/50, the texture in BD is significantly reduced. Beam profile 35/65 results in a texture in BD divided between $\langle 100 \rangle$ and $\langle 111 \rangle$, with a larger fraction of grains oriented in $\langle 100 \rangle \parallel \text{BD}$. A strong $\langle 111 \rangle \parallel \text{BD}$ texture is present in beam profiles 20/80 and 10/90.

are favored, which have a symmetry operation around BD of around 67° as well, since this would allow them to have their epitaxial growth closely aligned with the temperature gradient. This tilted grain growth can be referred to Pham et al. [17], who concluded that grain growth in the melted area would be favored in the direction already present at the melt pool boundary due to the effect of side branching. However, this growing direction might not perfectly align with the temperature gradient. This gives grains with an orientation of $\langle 111 \rangle \parallel \text{BD}$, which exhibit a symmetry operation of 60° around BD, the highest probability of continued growth, leading to the observed texture heavily leaning towards that.

The observations align with those of Pérez-Ruiz et al. [25], who used a uniform scanning direction and therefore attribute the resulting $\langle 111 \rangle$ texture of IN718 solely to the curvature at the bottom of the cross-sections of the solidified melt tracks. The use of non-Gaussian beam profiles (comparable to 65/35, 20/80 in this study) resulted in flatter, wider melt pools with a larger radius of curvature compared to the use of nearly Gaussian beam profiles (95/5 in this study) [25]. From this, Pérez-Ruiz et al. [25] deduced that the unit cells also tilt accordingly, resulting in a $\langle 111 \rangle$ texture. Thus, although a different scanning strategy was used in the present study than in [25] (67° rotation compared to uniform scan direction), the resulting microstructure is comparable. This shows that the tailoring of texture towards a $\langle 111 \rangle \parallel \text{BD}$ is also possible using an industrially viable scan rotation strategy by using the adjusted melt pool curvature and the symmetry operation of the $\langle 111 \rangle$ orientation around BD.

Perpendicular to the building direction (y-direction), the intensely textured sample of beam profile 95/5 has a preferred orientation that aligns with $\langle 100 \rangle$ for the most part. This shifts to a strong $\langle 110 \rangle$ preference for the specimens with higher ring intensity (above a

35/65 distribution). This observation coincides with the change in orientation preference in BD. Due to the discussed theory, this is explainable via geometric considerations. During the process, a hatch rotation angle φ_{rot} of 67° is used between the layers. According to [39], this reduces the directional preference in the x-y plane to a minimum and leads to a more random orientation. For a $\langle 100 \rangle \parallel \text{BD}$ microstructure, the reduction in texture works quite well since the rotation symmetry angle of the $\langle 100 \rangle$ crystals is 90° around z/BD. In this case, the symmetry and layer rotation angles are different enough to prevent a preferred orientation in the x-y plane. When the preferred direction in BD is tilted towards $\langle 111 \rangle$, the rotation symmetry angle changes to 60° , which seems close enough to the interlayer hatch rotation angle of 67° that $\langle 110 \rangle$ is preferential in this plane. Therefore, all samples with a $\langle 111 \rangle \parallel \text{BD}$ preference also have a $\langle 110 \rangle$ preference in the y-direction.

Fig. 7 shows the IPF maps of the investigated parameter sets, which result in a larger dimensionless hatch h^* between 1.6 and 1.7 and correspondingly a smaller remelted area. The columnar grain growth typical for PBF-LB/M is visible in these IPF maps. For the parameter sets with h^* between 1.6 and 1.7, significantly less texture is detectable than for those produced with a dimensionless hatch of 1.3–1.4 (Fig. 4). In particular, the samples produced using beam profiles with a small power fraction in the ring (beam profiles 95/5, 75/25, and 65/35) show hardly any preferred crystal orientation in BD and the y-direction. The samples manufactured with more power in the ring (beam profiles 35/65, 20/80, and 10/90) show a slight texture parallel to BD and a very clear texture in the y-direction. Parallel to BD, these samples show several orientations, ranging from $\langle 100 \rangle$ to $\langle 111 \rangle \parallel \text{BD}$, while having developed some grains with orientations in between. Beam profile 35/65 shows a slight rotation away from $\langle 111 \rangle \parallel \text{BD}$, which can similarly be observed

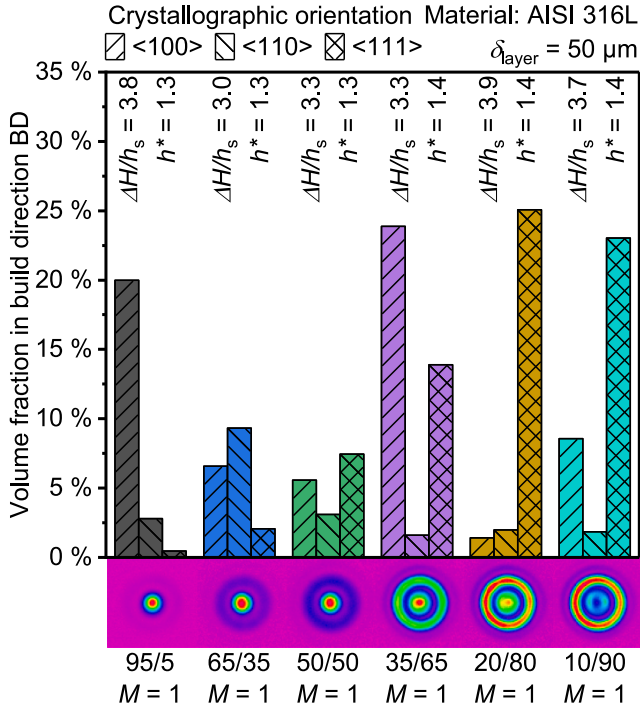


Fig. 5. Volume fractions close to the high-symmetry directions $\langle 100 \rangle$, $\langle 110 \rangle$, and $\langle 111 \rangle$ orientations within a tolerance angle of 10° in build direction of specimens with a dimensionless hatch h^* between 1.3 and 1.4. Using beam profile 95/5 results in a high volume fraction of approximately 20 % in $\langle 100 \rangle$. With beam profiles 65/35 and 50/50, no large volume fraction is present in a specific direction. When using the beam profile 35/65, the $\langle 100 \rangle$ direction and the $\langle 111 \rangle$ direction are highly pronounced, with the $\langle 100 \rangle$ direction having the highest volume fraction with 24 %. Beam profiles 20/80 and 10/90 result in large volume fractions of approximately 25 % in the $\langle 111 \rangle$ direction.

via a shift in preferred orientation in the y-direction.

The enlarged dimensionless hatch causes smaller remelted areas of the individual melt tracks. For samples manufactured with the beam profile 95/5, this smaller remelted area prevents the material from developing a preferred orientation, resulting in less texture. The results align with various studies showing and discussing that the textures in PBF-LB/M of 316L using Gaussian beam profiles decrease with increasing hatch distances [40,41]. Therefore, this phenomenon is not discussed in detail. In the case of the beam profiles with a large power fraction in the ring (beam profiles 20/80 and 10/90), the increase in the hatch distance means that the central part of the melt pool cross-section is no longer remelted, allowing the $\langle 100 \rangle$ direction to become more pronounced. On the one hand, this leads to an increase in the fraction in the build direction, exceeding the fraction of the $\langle 111 \rangle$ orientation. On the other hand, grains with $\langle 100 \rangle$ orientation are also partially formed parallel to y due to the geometry of the face-centered cubic unit cells.

In addition to the crystallographic texture, the grain size and morphology are evaluated from the EBSD measurements. Fig. 8 shows the equivalent grain radius $r_{50,2}$ and the corresponding volume energy density E_V for the parameter sets investigated. The parameter sets are sorted according to the spot size used (increasing spot size from left to right). Larger spot sizes tend to result in larger grains. Heiden et al. [42] reported that higher applied volume energy densities result in coarser microstructures with larger grains. One explanation for the coarser microstructure is a change in the average cooling rate when using larger beam profiles. Coarser microstructures typically result from lower cooling rates [43]. This study keeps the dimensionless enthalpy per dimensionless hatch in a similar range. Due to the dependence of this parameter on the spot size, constant dimensionless enthalpies result on the one hand in higher volume energy densities for larger spot sizes (see transparent bars in Fig. 8). Due to a higher average energy content in the material (larger beam profiles combined with higher volume energy density), the average thermal gradient is supposed to be lower, which reduces the average cooling rate, resulting in a coarser grain microstructure. On the other hand, lower cooling rates can be estimated by the dimensionless approach used for the design of experiments (see section 2.2). According to this approach, a similar dimensionless temperature

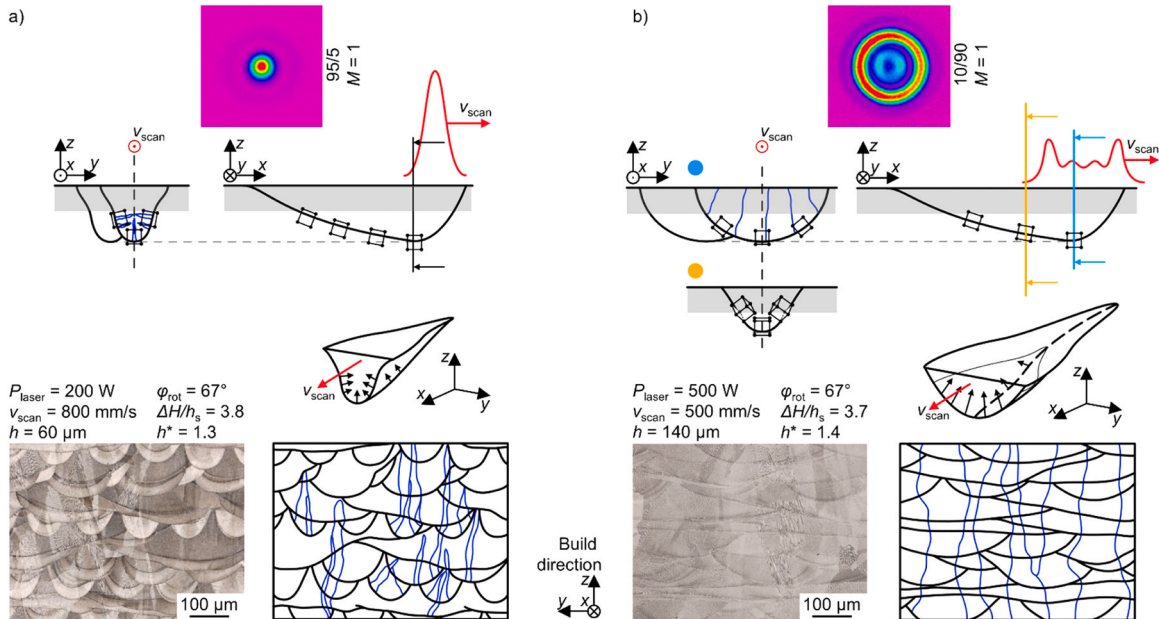


Fig. 6. Schematic illustration of the theory for forming the preferred crystal orientation of the microstructure during solidification and corresponding etched cross sections using a dimensionless hatch of 1.4; a) beam profile 95/5 generates deep and narrow melt pools resulting in steep side walls and an inclination close to parallel to the BD, enabling the grains to grow from the melt pool bottom and the side walls with the preferred $\langle 100 \rangle$ equivalent orientation; b) beam profile 10/90 creates melt pools with less steep side walls with a maximum angle of approximately 45° to the x-y plane resulting in a rotated preferred growth direction. The three-dimensional schematic melt pool representations are based on high-fidelity multi-physics simulations that were carried out as part of [38].

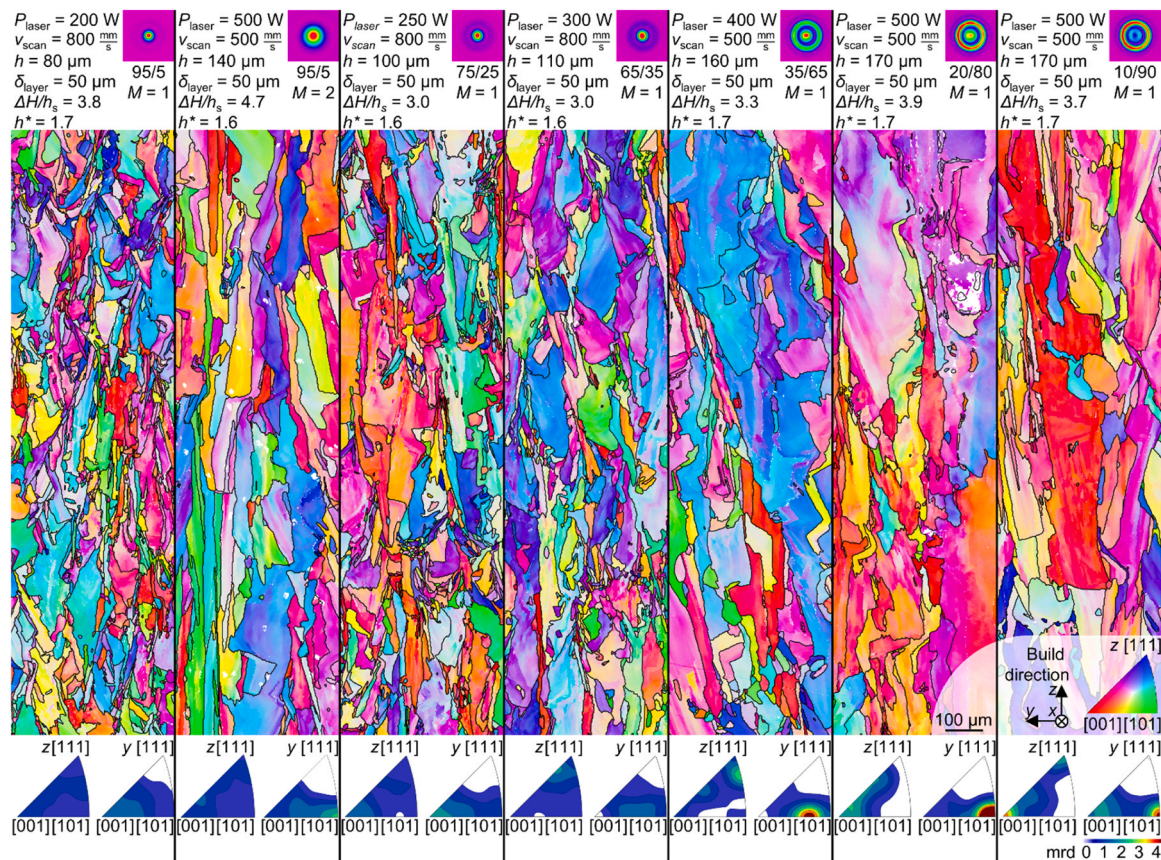


Fig. 7. Sections of the IPF maps and corresponding IPFs of the specimens resulted in a dimensionless hatch between 1.6 and 1.7. The textures in BD are significantly less pronounced than when using a dimensionless hatch between 1.3 and 1.4.

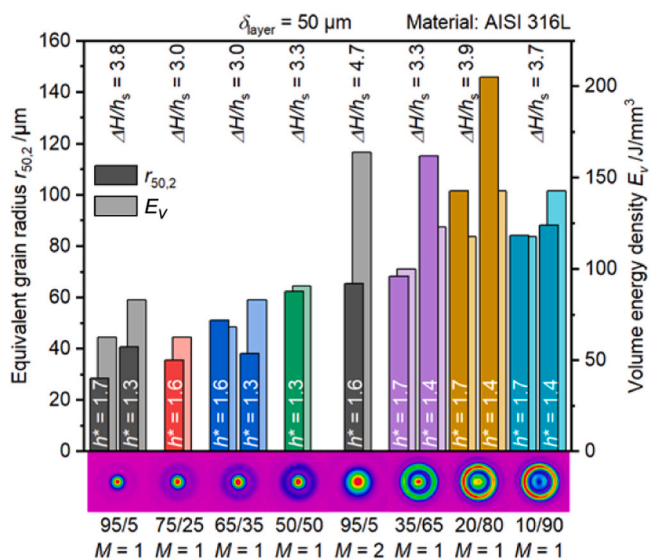


Fig. 8. Equivalent grain radius $r_{50,2}$ of the specimens with the respective dimensionless enthalpy $\Delta H/h_s$ and dimensionless hatch h^* . The beam profiles are sorted by size. The use of larger beam profiles tends to result in larger grains.

field results from similar dimensionless enthalpy values and Peclet numbers [32]. Since the coordinates of this field are scaled to the spot size, the dimensioned temperature field for larger spot sizes is larger in dimensional coordinates, equally resulting in reduced thermal gradients.

Fig. 9 shows the mean aspect ratio of the grain shapes of the specimens. Aspect ratios between 3 and 5 can be observed for all specimens. These high aspect ratios are typical values for columnar dendritic grain growth, as expected for PBF-LB/M specimens. This means that the ratios of thermal gradient and solidification rate for the parameters

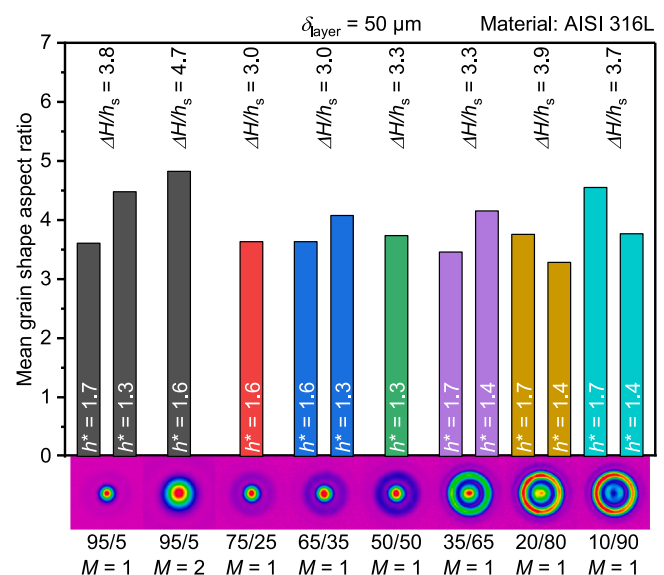


Fig. 9. Mean grain shape aspect ratios of the specimens with the respective dimensionless enthalpy $\Delta H/h_s$ and dimensionless hatch h^* . All samples show mean grain shape aspect ratios between 3 and 5, representing typical values for the columnar dendritic grain growth expected for PBF-LB/M.

investigated are assumed to be in a similar range for the parameter sets investigated. Roehling et al. [22] report that beam shaping offers the potential to tailor the thermal field during processing in a way that equiaxed dendritic grain growth can occur. However, this is not the case within the parameter space investigated in this study. To achieve remaining equiaxed grains, the quotient of thermal gradient and growth rate would have to be reduced. However, compared to crystallographic texture, grain morphology has little influence over macroscopic material properties [13,44].

The ability to locally influence the microstructure by adjusting the beam profile provides the basis for creating functionally graded components. Therefore, the combination of two parameter sets remains to be investigated in the second part of this study.

3.2. Impact of the combination of beam profiles on the as-built microstructure and the transition zone

In order to enable load-case-optimized microstructures using PBF-LB/M, it is essential to fundamentally understand the microstructure behavior in the transition zone between two exposure strategies. Fig. 10 shows optical microscopy images of two specimens of two segments manufactured with different beam profiles. Beam profiles and parameter sets that differ significantly in the results of the microstructure analysis of the single parameter set samples were selected for the mutual influence studies: First, a state-of-the-art beam profile in a standard size (beam profile 95/5 with $M = 1$) and second an enlarged state-of-the-art beam profile (beam profile 95/5 in $M = 2$), both were combined with an ring-shaped beam profile with approx. 200 μm spot size (beam profile 10/90 in $M = 1$).

When combining the small standard beam profile (95/5 with $M = 1$) and the larger ring-shaped beam profile (10/90 with $M = 1$), the transition between the different segments is clearly recognizable (see Fig. 10a). The transition is less clear when combining two beam profiles with similar sizes (beam profile 95/5 with $M = 2$ and beam profile 10/90 with $M = 1$, see Fig. 10b). Although the areas are distinguishable due to the differently shaped melt pool cross-sections, it is recognizable that grains grow across the boundary of the segments (see Fig. 10b). However, to draw valid conclusions about the crystal and grain structure, EBSD analyses are required.

Fig. 11 shows the IPF maps of the segmented samples measured in the area of the first beam switching plane (see marked area in Fig. 2b). Fig. 11a shows results from the EBSD measurements of the specimen combining the beam profiles 95/5 with $M = 1$ and 10/90 with $M = 1$,

while Fig. 11b shows results of the specimen combining the larger beam profile 95/5 with $M = 2$ and 10/90 with $M = 1$. For the combination of the smaller beam profile 95/5 and the beam profile 10/90, a grain size transition between the two segments can clearly be distinguished in the IPF map. The white solid line marks the border between the segments and thus the beam profile switching plane. The crystallographic orientation of both segments is almost identical, exhibiting a mostly evenly distributed orientation with a slight preference to $\langle 100 \rangle$ parallel to BD. In y, the texture is in between a $\langle 100 \rangle$ and a $\langle 110 \rangle$ orientation (see Fig. 11a).

In contrast, the specimen combining the larger beam profile 95/5 with $M = 2$ and 10/90 with $M = 1$ does not exhibit an evident grain size transition. Yet, the preferred texture orientation shifts from a slight $\langle 100 \rangle \parallel$ BD preference with a $\langle 111 \rangle$ fraction (lower segment in Fig. 11b) to a slight $\langle 111 \rangle$ preference (higher segment in Fig. 11b). Along the y direction, the texture shows a preferred $\langle 110 \rangle$ orientation for both segments.

To the right of the upper part of the IPF maps in Fig. 11, the area fractions of the grains cut by the beam profile switching plane are shown in relation to the area of all grains in the measurement areas. At an area fraction of 50 % the height of the transition zone is marked. When using the small, almost Gaussian beam profile 95/5 to manufacture the initial layers, the transition zone is comparatively small at a height of approximately 175 μm (3.5 layers). When using the larger, nearly Gaussian beam profile 95/5 in initially manufactured layers, the transition zone increases to 1475 μm (29.5 layers). For comparison, Sofinowski et al. [16] conducted experiments on the dependence of the preferred grain orientation in the y direction on the scan vector direction using state-of-the-art Gaussian beam profiles. They reported transition heights between the textures between 760 μm and 1120 μm (corresponding to 19–28 layers with 40 μm layer thickness), which aligns with the specimen manufactured with similar beam diameters in this study (see Fig. 11b). While in the experiments by Sofinowski et al. [16], the texture orientation in the scanning plane was heavily affected by the change of the scan rotation, the texture parallel to BD remained almost constant at $\langle 110 \rangle$ over the entire sample cross-section. Furthermore, the height of the transition zone was influenced by the present interface grain morphology, meaning that larger grains tend to grow epitaxially and gradually rotate their grain orientation around BD until the growth direction is again most closely aligned with the new temperature gradient [16]. This behavior can also be observed on the specimen manufactured using similar beam sizes (see marked grain in Fig. 11b). It can thus explain the longer transition zones when larger grain sizes are

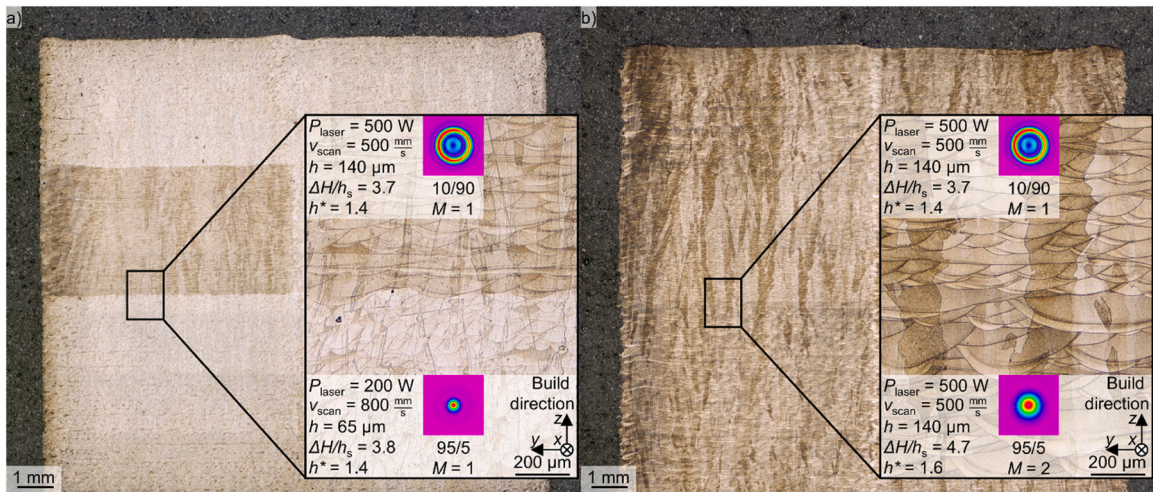


Fig. 10. Etched cross-sections of specimens consisting of segments with different parameter combinations a) beam profile 95/5 with $M = 1$ (spot size of 92 μm) and beam profile 10/90 with $M = 1$ (spot size of 205 μm) and b) beam profile 95/5 with $M = 2$ (spot size of 175 μm) and beam profile 10/90 with $M = 1$ (spot size of 205 μm).

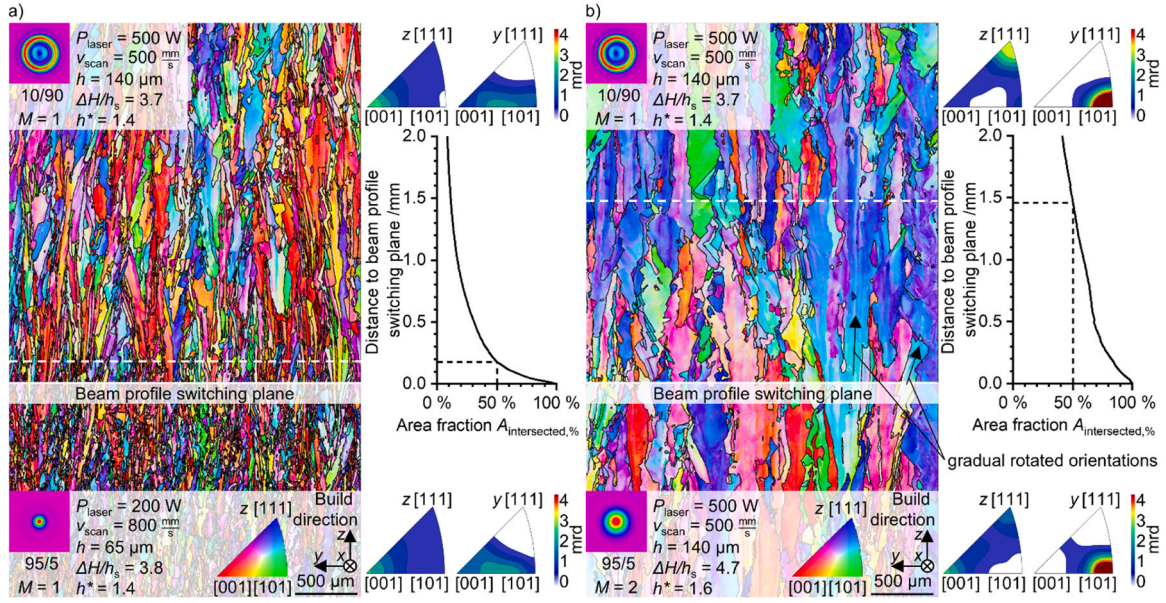


Fig. 11. IPF maps, IPFs, and evaluated transition zones of the specimen combining two segments with different parameter sets and beam profiles, a) beam profile 95/5 with $M=1$ and beam profile 10/90 with $M=1$ and b) beam profile 95/5 with $M=2$ and beam profile 10/90 with $M=1$.

present at the interface. This also influences the microstructure in subsequent layers.

Comparing the microstructural outcomes of the specimen manufactured using beam profile 10/90 reveals a strong dependence of the emerging microstructure on the microstructure of the substrate material

or underlying layers. Fig. 12 compares the microstructures produced using beam profile 10/90 with $M=1$ and the parameters $P_{\text{laser}} = 500 \text{ W}$, $v_{\text{scan}} = 500 \text{ mm/s}$, and $h = 140 \mu\text{m}$. The IPFs show that the single-parameter set cuboid sample produces the strongest $\langle 111 \rangle$ preference. Closest to this texture comes the segmented specimen, where the corresponding segment was built on a coarse microstructure produced using the enlarged beam profile 95/5 ($M=2$). This crystallographic orientation could not be reproduced when using the fine-grained microstructure derived from using the small beam profile 95/5 ($M=1$) in the initially manufactured layers. The grain sizes also differ depending on the base microstructure present (see Fig. 13). The grains of the test specimen that were built on the segment with the small beam profile 95/5 are the smallest, with an equivalent grain radius of $66.1 \mu\text{m}$, followed by the grains of the test specimen that was manufactured directly on the build platform with an equivalent grain radius of

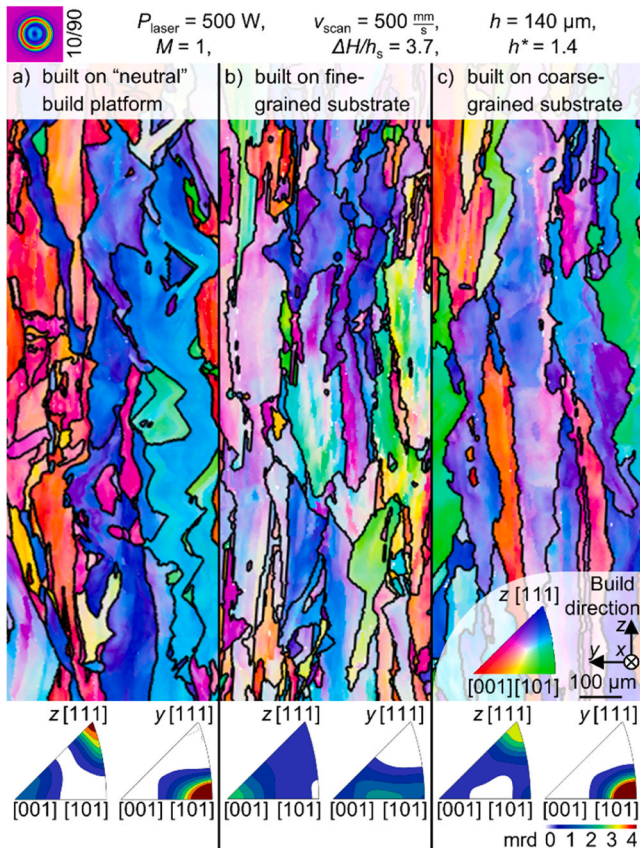


Fig. 12. Comparison of IPF maps using beam profile 10/90 with $M=1$ and the parameter set $P_{\text{laser}} = 500 \text{ W}$, $v_{\text{scan}} = 500 \text{ mm/s}$, and $h = 140 \mu\text{m}$ on different underlying microstructures.

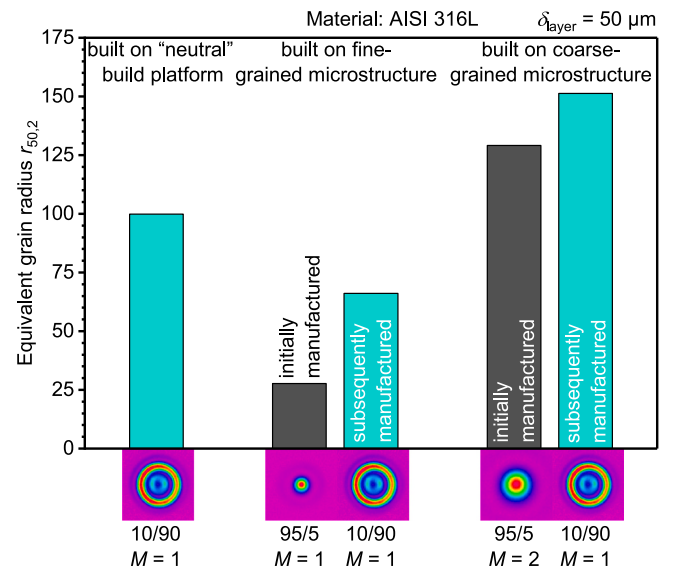


Fig. 13. Comparison of the equivalent grain radius $r_{50,2}$ using beam profile 10/90 with $M=1$ and the parameter set $P_{\text{laser}} = 500 \text{ W}$, $v_{\text{scan}} = 500 \text{ mm/s}$, and $h = 140 \mu\text{m}$ on different underlying microstructures.

99.9 μm . The largest grains with an equivalent grain radius of 151 μm are found in the sample using the parameter set on a sample built with the large beam profile 95/5 ($M = 2$).

This means that with one identical parameter set (beam profile 10/90 with $M = 1$, $P_{\text{laser}} = 500 \text{ W}$, $v_{\text{scan}} = 500 \text{ mm/s}$, and $h = 140 \mu\text{m}$), three different solidification microstructures can be induced depending on the underlying microstructural conditions. Using an identical beam profile and the same process parameters in this region, the solidification conditions (such as thermal gradient, growth rate, and latent heat release) and the thermal history are similar. Therefore, the different microstructures received are likely to be driven primarily by epitaxial grain growth, as discussed in [45]. The different grain sizes can be attributed to the non-required nucleation formation in AISI 316L [8]. As a result, the grains present in the initial layers continue to grow. Under constant solidification conditions, this leads to a grain size in subsequently manufactured layers corresponding to the initial layers' grain sizes. The different textures in the initially produced layers additionally influence the microstructure in the subsequently manufactured layers. Mainly since the 10/90 ring parameter set produced a split texture in the single parameter set cuboid sample, which also exhibited an amplitude in the $\langle 100 \rangle$ direction while consisting of a larger fraction of $\langle 111 \rangle$ grains, the existing $\langle 100 \rangle$ orientation of the microstructure of the sample produced using the small Gaussian beam profile might be energetically more favorable (see Fig. 11a). The ability of beam profile 10/90 to produce $\langle 100 \rangle$ -oriented grains in addition to $\langle 111 \rangle$ may prolong the transition of crystallographic texture from a predominant $\langle 100 \rangle$ to a $\langle 111 \rangle$ texture. This suggests that the initial grain size and crystallographic orientation of the substrate or initially manufactured layers play an essential role in determining the final texture and grain size of the inner segment, or at least the build height required for a corresponding shift. This is particularly important when combining several parameter sets in one component, as typically several millimeters of the connection to the substrate are cut off when using fixed parameters in the complete component.

The findings regarding the segmented specimen can be summarized into two points. First, there are two types of transitions. On the one hand, there is a grain size transition, meaning that the grain size changes significantly between the two segments. The other is a texture transition in which the preferred crystallographic orientation shifts between the two segments (analogous to the observations of Sofinowski et al. [16]). Interestingly, in the specimens analyzed in this study, when one of the two transition types occurs, the other does not. In the case of the specimen combining the small beam profile 95/5 with $M = 1$ and 10/90 with $M = 1$, there is a grain size transition but no noticeable texture transition. Combining the larger nearly Gaussian beam profile 95/5 with $M = 2$ and 10/90 with $M = 1$, there is a texture transition, but no visible grain size transition. Second, the grain size of the emerging microstructure depends on the grain size of the initially manufactured layers. If a fine-grained substrate is used, smaller grains result than if a coarse-grained microstructure is used as a base.

4. Conclusion

The present study investigates the influence of alternative beam profiles on the as-built microstructure during laser-based powder bed fusion of AISI 316L. Single-parameter cuboidal specimens and cuboidal specimens consisting of two segments are built using different beam profiles to demonstrate the influence of the beam profile and the hatch distance on the resulting microstructure.

The results can be summarized as follows:

- The crystallographic texture is strongly dependent on the beam profile and dimensionless hatch distance:
 - o For the nearly Gaussian beam profile 95/5, a $\langle 100 \rangle \parallel \text{BD}$ texture can be created using a small hatch distance ($h^* = 1.3$).

- o The ring-shaped beam profiles 20/80 and 10/90 have the potential to generate a $\langle 111 \rangle \parallel \text{BD}$ texture.
- The grain sizes correlate with the size of the laser beam profiles in the parameter sets investigated. Larger beam profiles result in larger grains. This is attributed to the scaled temperature field aimed for in the experimental design.
- When combining various parameter sets within one specimen, the microstructure of initially manufactured layers strongly influences the emerging microstructure and the size of the corresponding transition zone:
 - o If smaller grains are present in initially manufactured layers, a fine-grained microstructure forms in subsequent layers. A coarse-grain microstructure is formed if larger grains are present in initially manufactured layers.
 - o The resulting texture in subsequently manufactured layers is strongly oriented towards the initial texture.
 - o The height of the transition zone depends on the grain size in the initially manufactured layers. The transition zone is smaller on a finer-grained basis.

The results demonstrate the potential of alternative beam profiles to tailor the microstructure of additively manufactured components. This potentially allows for locally modifying mechanical properties, as these mainly depend on crystal orientation.

Combining different beam profiles during manufacturing one part requires the consideration of transition zones and the microstructure of initially manufactured layers, as the resulting as-built microstructure varies significantly depending on the microstructure of preceding layers. This observation is based on the investigation of the transition zone between two distinct microstructures and their interaction, being one of the key novelties of this publication. The demonstrated understanding and its application in design and processing contribute to enabling the additive manufacturing of metallic components with locally tailored microstructures and thus functionally graded component properties. However, to gain more detailed insights, further studies could investigate the influences on other material properties such as surface roughness, residual stresses, and mechanical characteristics.

CRedit authorship contribution statement

Jonas Grünwald: Writing – review & editing, Writing – original draft, Visualization, Validation, Supervision, Software, Project administration, Methodology, Investigation, Formal analysis, Data curation, Conceptualization. **Raphaela Pfund:** Writing – review & editing, Software, Investigation, Data curation. **Daniel Röhrer:** Writing – review & editing, Software, Investigation, Data curation. **Leonhard Hitzler:** Writing – review & editing, Writing – original draft, Formal analysis. **Jan Torgersen:** Writing – review & editing, Supervision, Resources. **Katrin Wudy:** Writing – review & editing, Supervision, Resources, Conceptualization.

Declaration of generative AI and AI-assisted technologies in the manuscript preparation process

During the preparation of this work, no generative AI technologies were used. The authors used “Grammarly” to check the manuscript for spelling, grammar, and style. After using this tool/service, the authors reviewed and edited the content as needed and take full responsibility for the content of the published article.

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Declaration of Competing Interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

Data availability

Data will be made available on request.

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