1. Introduction

Forming tools often exhibit complex geometries and internal cooling channels, to grant fast cooling rates of the forming parts and reduce thermal gradients and thermal stresses inside the tool.\(^1\) The construction of complex-shaped cooling systems located close to the tool surface is desired for optimized tool efficiency and part quality.\(^4\) Commonly, cooling channels are implemented in forming tools by machining or as pipes during casting of the tool.\(^4,5\) However, especially subtractive manufacturing processes, such as drilling, are strongly restricted in terms of undercuts or internal structures, limiting the geometrical complexity of the manufactured tools. During the last years, additive manufacturing (AM) processes have gained importance in the production of complex-shaped parts and represent an alternative manufacturing route for complex-shaped tools.\(^6\) A highly developed AM process for the production of steel parts is laser powder bed fusion (L-PBF). During L-PBF, a powder bed is selectively melted by a focused laser beam, creating a part layer by layer from 3D computer-aided design data.\(^10\) Within the process, the focused laser radiation creates small melt pools, which solidify rapidly due to high cooling rates caused by a strong heat flux from the melt pool into the already solidified part. The movement of the laser across the powder bed and the repeated melting of new layers lead to a temporally and locally unstable heat flow,\(^11\) resulting in a heterogeneous microstructure of L-PBF-built parts.\(^13,14\) Moreover, the complex heating and cooling conditions in a part during the L-PBF process induce high residual stresses, especially if thermal stresses are superposed by transformation stresses. Therefore, crack formation and propagation can occur during L-PBF, limiting the variety of processable steels to those with a good ductility and a nonalloyed behavior. Thus, most research concentrated on austenitic stainless steels, such as AISI 316L or maraging steels.\(^14\) While the weldability and thus the processability of these steels with L-PBF is good, the mechanical and tribological properties of these steels generally do not meet the requirements for forming tool applications in terms of hardness, strength, and wear resistance. In forming applications, commonly martensitic secondary hardenable tool steels are used as tool materials because of their good wear resistance, high strength, and thermal stability.\(^15\) However, the processing of martensitic secondary hardenable tool steels with L-PBF is a challenging task because of the formation of high residual stresses due to martensitic transformation, thus leading to distortion and cracking. While the production of crack-free and dense samples from various hot work tool steels by preheating the parts to a temperature higher than the respective martensite start temperature is widely reported, the complex heating and cooling conditions combined with the complex transformation and precipitation behavior of secondary hardenable tool steels are still discussed controversially.\(^16,17\) The buildup of parts during L-PBF is performed by the melting and the solidification of a large number of small material

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Laser powder bed fusion (L-PBF) of forming tools has become of major interest in the tooling industry because of the high geometrical flexibility of this process. During L-PBF, a metallic powder bed is melted selectively by a laser beam, enabling the layer-wise manufacturing of parts from 3D computer-aided design data. The process is characterized by a locally and temporally unsteady heat flow in the solidified part and in the melt pool, causing nonequilibrium solidification and phase transformations. In addition, rapid heating and cooling occur, promoting the formation of microstructural defects, cold cracks, and distortion. Because of the high tendency to form cold cracks, processing of martensitic tool steels is still a challenging task. Tool steel X65MoCrWV3-2 is processed by L-PBF and the resulting microstructure and the associated local properties are investigated by microhardness measurements, nanoindentation, and scanning electron microscopy. It is gathered from the investigations that regions of different microstructures and mechanical properties on both micro- and macroscale are present in the L-PBF-densified steel. The different microstructures and properties are the result of the alternating heat insert at different temperature regimes, forming heat-affected zones in which the tempering processes are triggered and strongly varying properties are generated.

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volumes, and the repeated heating and cooling influences the microstructure in terms of heat-affected zones (HAZ) significantly. Thus, microstructural features on different dimensional scales define the material properties in L-PBF-built condition. Therefore, this work characterizes the microstructure and the associated mechanical properties of L-PBF-built martensitic X65MoCrWV3-2 tool steel on macro-, micro-, and submicron scale. Electron scanning microscopy, microhardness measurements, and nanoindentation are used to link the material’s properties with the respective microstructure.

2. Experimental Section

2.1. Materials

In this work, martensitic secondary hardenable X65MoCrWV3-2 tool steel is investigated in L-PBF-built and laser-melted condition. For the characterization of the steel in L-PBF-built condition, gas atomized powder, provided by Deutsche Edelstahlwerke Specialty Steel GmbH & Co. KG, was processed with L-PBF. As determined by laser diffraction measurements using MasterSizer 2000 from Malvern Instruments Ltd. and scanning electron microscopy (SEM), the used powder showed mainly spherical particles with a low amount of irregularly shaped particles or satellites and a unimodal particle size distribution (d50 = 44.51 μm). The size distribution and the morphology of the particles result in sufficient flow properties, characterized by a flow time of 15.61 s, a bulk density of 4.07 g cm⁻³, and a packing density of 52%, which were measured in accordance to standard DIN ISO 697 and DIN EN ISO 4490.

In addition to the L-PBF buildup, the influence of the melting of a single layer on the already solidified material was investigated. For this purpose, the surface of X65MoCrWV3-2 bulk material was melted with the laser of the L-PBF 100 device representing the densification of a single layer in the L-PBF-process. Bulk material was provided by Dörrenberg Edelstahl GmbH. The samples were taken from cast, forged, and soft-annealed material by cutting and were subsequently austenitized at a temperature of 1050 °C for 20 min and quenched in water. Finally, the samples were ground with SiC abrasive paper with a mesh size of 1000. The chemical composition of the powder and the bulk material was measured by energy-dispersive X-ray spectroscopy (EDS) and optical emission spectroscopy (OES), respectively, and is listed in Table 1. The presence of certain contents of elements W and V is indicated by a + in Table 1. For reasons of secrecy, the exact values may not be mentioned here.

| Element | C | Cr | Mo | Si | W | V | O | N | Fe |
|---------|---|----|----|----|---|---|---|---|---|
| X65MoCrWV3-2 powder (EDS) | 0.60 | 2.45 | 3.58 | 0.32 | + | + | 0.017 | 0.017 | bal. |
| X65MoCrWV3-2 bulk (OES) | 0.65 | 1.92 | 2.9 | 0.19 | + | + | - | - | bal. |

2.2. Laser Powder Bed Fusion

The L-PBF-densification of the investigated steel was performed using a Realizer SLM 100 device, equipped with a ytterbium fiber laser with an effective output energy \( P_{\text{eff}} \) of 73.5 W at a wavelength of \( \approx 1064 \) nm and a focal diameter of \( \approx 90 \) μm. During the buildup process, the build platform was heated to 300 °C, using a specially designed heating device. As process atmosphere, Ar gas was used and the O₂ content inside the build chamber during the process was kept <0.3 vol%. Based on preliminary results, samples were produced with a point distance between two adjacent laser spots \( d_\text{p} \) of 30 μm, an exposure time per laser spot \( t_\text{p} \) of 100 μs, and a slice thickness of 30 μm. An alternating x-y-scanning strategy was used.\(^{[18]}\) With respect to Equation (1) and with the effective laser power \( P_{\text{eff}} \), the point distance \( P_{\text{p}} \), and the exposure time \( t_\text{p} \), the energy input per length unit \( E_\text{l} \) for the used laser parameters was calculated to be 245 J m⁻¹.

\[
E_\text{l} = \frac{P_{\text{eff}}}{E_\text{p}}
\]

To investigate the influence of the melting of a single layer on the solidified part, laser melting of the surface of cubic bulk samples (8 × 15 × 50 mm) was performed, using the built-in laser in the SLM 100 machine. With this method, a microstructure was produced, which corresponds to the microstructure of the last solidified layer, and microstructural changes inside the already solidified material can be observed. During melting, the laser exposure time was set to 1000 μs at a point distance of 30 μm. Laser melting was performed at room temperature and 200 °C, to reproduce the influence of preheating of the samples on microstructural evolution. The melted surface area was 5 × 5 mm in size and divided into the inner hatch area and an outline hatch that was created subsequent to the hatch area by a single laser track. The schematic movement of the laser during the melting process is shown in Figure 1b.

2.3. Microscopy

Microstructural investigations were conducted on samples taken from L-PBF-built and laser-melted bulk material. For cross-sections, the samples were cut as shown in Figure 1. Finally, the samples were ground with SiC abrasive paper (mesh sizes 320–1000) and polished with a diamond particle suspension from 9, 3, to 1 μm. To make the microstructure detectable by means of optical microscopy (OM), the polished samples were etched with NITAL solution.

As microscopic investigation techniques, both SEM and OM were used. SEM investigations were conducted using a MIRA 3 SEM from Tescan in the secondary electron contrast mode at an acceleration voltage of 15 keV and a working distance of 15 mm. In addition, the local chemical composition of the L-PBF-built steel was measured with these parameters by means of EDS. For this purpose, an EDS detector type XmaxN (Oxford Instruments) was used. OM investigations were conducted with an OM type BX60M from Olympus.
2.4. Nanoindentation and Hardness Testing

For the characterization of the local mechanical properties of the investigated steel in L-PBF-built and laser-melted condition, nanoindentation and microhardness measurements were carried out. In the cross-sections of the laser-melted sample surfaces, arrays of $10 \times 30$ and $10 \times 20$ equidistant (50 $\mu$m) nanoindents were created, covering the last solidified outline hatch, as well as a part of the hatch area and a part of the sample, which was not melted by the laser. The nanoindentation tests were conducted with a CSM NHT indenter equipped with a Berkovich diamond indenter tip. The target load was set to 50 mN.

In addition to the nanoindentation tests, Vickers microhardness tests were conducted on the cross-sections of the L-PBF-built and laser-melted samples, using a normal force of $F_N = 0.9807$ N. By using microhardness measurements, it was possible to characterize a comparably large cross-section of an L-PBF-built sample.

3. Results and Discussion

3.1. Microstructure in L-PBF-Built Condition

The L-PBF-built X65MoCrWV3-2 tool steel shows a heterogeneous microstructure, which is characterized by a hierarchical composition. Similar to other Fe-based materials in the L-PBF-built condition, the investigated steel shows a fine subgrained microstructure consisting of cellular to columnar dendrites with relatively strong microsegregations in the interdendritic regions (Figure 2 and 3a). As determined by EDS measurements, mainly heavier elements, such as Mo, W, and V, segregate into the interdendritic spaces (Figure 2).

This microstructure is attributed to the process-specific high cooling rates which act during L-PBF, favoring a fast solidification with strong constitutional undercooling effects and a limited diffusion of especially heavy elements like Mo or W. Moreover, carbon is enriched in the interdendritic spaces, as indicated by the presence of carbides. On the one hand, the observed carbon segregations could be a result from solidification. On the other hand, the segregation of carbon might be a result of solid-state diffusion, because diffusion of light elements like carbon can occur at lower temperatures compared with heavy substitutional elements. The temperature level necessary for carbon diffusion could be provided by the iterating heat insertion into the solidified material during the L-PBF-process, as indicated by numerical simulations of the temperature distribution in parts during L-PBF. As reported in the literature, the interdendritic spaces possess a high dislocation density. Thus, the carbon enrichment might be connected with pronounced interactions between carbon atoms and dislocations and/or substitutional atoms in the interdendritic spaces. On a larger scale, martensitic plates are present in the microstructure of the L-PBF-built X65MoCrWV3-2 steel (Figure 3b). The martensite plates show a significant larger size than the observed subgrain dendrites. Furthermore, the plates extend over several dendrites, as shown in Figure 3a. Thus, the martensitic
microstructure appears to be superimposed on the subgrain structure. These findings indicate that martensite plates grow between large-angle grain boundaries and/or other existing martensite plates and extend over the segregation network.

3.2. Local Hardness in L-PBF-Built Samples

In L-PBF-built condition, the investigated steel samples possess a highly heterogeneous hardness distribution. As shown in Figure 4, the microhardness in the cross-section of the investigated steel varies in a range of 600–940 HV 0.1. Schematically, two different hardness regimes can be found: in the last solidified layer (referred to as region 1), a very high hardness of \( \approx 940 \) HV0.1 is reached, followed by a pronounced drop of the hardness close to the last solidified layer. In the previously densified layers (referred to as region 2), the hardness remains relatively stable in a range of 620–740 HV0.1, showing only a slight decrease with increasing distance to the last solidified layer. Correlating to the aforementioned hardness regimes, a locally different etching behavior was found in the respective regions. While martensite plates are easily detectable in region 2, only a small amount of martensite plates is visible in region 1. This indicates significant microstructural differences of the respective regions evoked by their different thermal history. The high hardness in the last solidified layer confirms the existence of martensite in region 1. The presence of the phase martensite was verified in our previous work by means of electron backscatter diffraction (EBSD) measurements.\(^{[18]}\) The remarkably lower hardness in region 2 suggests
3.3. Local Microstructure and Properties of Laser-Melted Surface Areas

To understand the microstructural processes triggered by the repeated heat flow during L-PBF, an investigation of the influence of a single laser-melted layer on the material was conducted at cross-sections of laser-melted bulk samples. The bulk samples were conventionally hardened before laser melting, creating a homogeneous fully martensitic microstructure, which allows to follow the microstructural changes during laser melting. Figure 5 shows SEM images of the cross-section of a laser-melted sample surface. As shown by the dotted lines, three regions of different microstructures can be recognized. At the surface of the sample, the melted and solidified region is located. Inside this region, the typical cellular to columnar dendrites are present, and no carbide formation can be observed. Thus, the microstructure shows great similarity to the microstructure of the L-PBF-built steel. The melted area is segmented into the former melt pools that tempering effects take place in the previously densified layers during the buildup of the sample. These tempering processes are induced by the alternating heat input during the melting of subsequent layers. Based on these results, the microstructure of the investigated martensitic steel can be understood as a product of a complex thermal history originating from the L-PBF-process, which represents a kind of multistep heat treatment. As a further indicator of a pronounced heterogeneity and high complexity of the present microstructure, the microhardness values scatter remarkably. Corresponding to the process-specific melt pool sizes in the range of hundreds of micrometers, microhardness varies by up to ≈70 HV0.1 in a distance of 100 μm.

Table 2. Results of the size measurements of the respective areas of the laser-melted samples.

| Preheating temperature | Depth of melted area in μm | Width of vHAZ in μm |
|------------------------|----------------------------|---------------------|
| RT                     | 277.4 ± 51.50              | 56.1 ± 11.8         |
| 200 °C                 | 296.6 ± 53.75              | 75.0 ± 23.4         |

For a quantification of the volume of the melted and heat-affected material, the depth of the former melt pools and the width of the darkly contrasted area were measured, as schematically shown in Figure 5a. Both values show a relatively strong scattering (Table 2). This is attributed to the unsteady heat input during laser melting. The local energy input depends on several conditions, inter alia the absorptivity of the material and the degree of contamination of the shield gas atmosphere above the melt pool. Regardless of the scattering, the use of a base plate preheating results in an increase in the melted and heat-affected area. The base plate preheating shifts the overall temperature level to higher temperatures, resulting in larger heat-affected volumes. Thus, a greater influence on the mechanical properties during laser melting can be anticipated when using a base plate preheating. Corresponding to the conducted laser-melting experiments, the local hardness and microstructure of L-PBF-built tool steels are influenced by the preheating temperature. The higher temperature level and increased heat input at higher preheating temperatures were found to result in a more homogeneous microstructure, having a lowered hardness.

To investigate the resulting local mechanical properties of the different heat affected regions, nanoindentation tests were performed on the cross-section of the last solidified outline hatch. The results of the nanoindentation testing are shown as a heat map in Figure 6. Independent from the used preheating temperature, the nanoindentation maps display a high hardness (bright colors) in the former melt pool of the investigated hatch. Additionally, a high hardness was found in the visually 

![Figure 5. SEM images of a cross-section of laser-melted surface of X65MoCrV3-2 steel: a) overview image; b) close-up.](image-url)
recognizable HAZ. With increasing distance from the aforementioned region, hardness shows a strong decrease (dark colors) in an area having a width of \( \approx 100 \) \( \mu \text{m} \) from the fusion line. By a further increase in distance, hardness increases slightly and fits the hardness of the uninfluenced material (RT: \( \approx 10.37 \) GPa and 200 \( ^\circ \text{C} \) \( \approx 10.19 \) GPa) in a distance of \( \approx 600–700 \) \( \mu \text{m} \) to the fusion line. From the obtained results, a plausible theory on the microstructural evolution of the heat-affected material can be derived. The high hardness in the solidified steel indicates the presence of an untempered martensite that has formed due to strong thermal gradients from the melt pool center toward the solid material and very high cooling rates in this area. However, the complete melting of the material and the resulting high solution state as well as fast cooling are generally assumed to stabilize retained austenite, lowering hardness. Contradictorily, the reason of the high hardness of the steel in this area could be found in the fine subgrain structure, which is known to possess high dislocation density and strength. With increasing distance to the melt pool center, the heat input by the laser beam is not high enough to cause melting of the steel. Instead, a complete reaustenitization of the steel takes place. Again, the fast cooling in this area is assumed to stabilize retained austenite, lowering hardness. Contradictorily, the reason of the high hardness of the steel in this area could be found in the fine subgrain structure, which is known to possess high dislocation density and strength. Underneath the strongly softened material, a moderately softened area can be understood as a result of a heat influence that is comparable with conventional tempering. In addition to martensite relaxation, precipitation of secondary carbides might be initiated, transforming the microstructure into a tempered condition. The size of the area, in which tempering takes place, corresponds to the size of 2–3 three layers, which can be derived from the microhardness profile obtained at the L-PBF-built samples (Figure 4). The obtained results are in good accordance with the work of Li et al., who found highly similar microstructural patterns of different HAZ. Including simulations of the thermal history of single spots, Li et al. also assumed the presence of differently tempered martensite in the respective HAZ.

The findings of the nanoindentation experiments generally support previous investigations of the macroscale hardness-tempering behavior of X65MoCrWV3–2 steel. As published in a previous study, the as-built steel does not show a hardness decrease in the tempering temperature range up to 400 \( ^\circ \text{C} \) (Figure 7). In contrast, the same material after an additional quenching (\( T_{\text{aust}} = 1050 \) \( ^\circ \text{C} \), \( t_{\text{aust}} = 20 \) min, quenching medium: water) possesses higher hardness after quenching, followed by a relatively steep decrease in hardness up to 400 \( ^\circ \text{C} \). Furthermore, the peak hardness of the as-built and tempered material exceeds the peak hardness of the quenched and tempered steel. Taking the results of the nanoindentation experiments into account, the stable hardness level of the as-built and tempered material is attributed to an accumulated tempering effect induced by the iterative heat insertion during the buildup. Although partly reaustenitization and martensite formation occur, the heat input in the L-PBF-process results in an overall...
4. Conclusions

Combining the results of the nanoindentation tests with the microhardness profiles obtained at L-PBF-built samples, several conclusions can be drawn. 1) The L-PBF-built X65MoCrWV3–2 tool steel shows heterogeneous microstructure and varying microhardness. The microstructure consists of martensite and tempered martensite. 2) During the L-PBF process, an in situ multistep heat treatment of already solidified material takes place. It is induced by the multiple heat input that results from the layer-wise buildup process. 3) Newly solidified material shows an untempered martensitic microstructure. The heat input by the laser beam leads to a partly reaustenitization of the already solidified material, followed by martensite formation. The previously solidified 2–3 layers however are tempered by the heat input produced by the laser beam. 4) Use of a base plate preheating increases the size of the HAZ and leads to a stronger impact of the heat input by the laser beam on the solidified material. 5) The local microstructure and local properties of L-PBF-built X65MoCrWV3–2 tool steel resulting from the process-specific conditions can explain the macroscopic hardness-tempering behavior of the material. During L-PBF, the microstructure of the steel is thermally stabilized, leading to a stable hardness level up to tempering temperatures of 400 °C. 6) As a result of the re-entrant heat input over 2–3 buildup layers, the L-PBF process causes tempering of previously hardened areas so that additional heat treatment is not mandatory to achieve high hardness.

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Conflict of Interest

The authors declare no conflict of interest.

Keywords

additive manufacturing, laser powder bed fusion, martensite, microstructure formation, tool steels

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[1] F. Schieck, C. Hochmuth, S. Polster, A. Mosel, CIRP J. Manuf. Sci. Technol. 2011, 4, 189.
[2] B. He, L. Ying, X. Li, P. Hu, Appl. Therm. Eng. 2016, 106, 1176.
[3] H. Hoffmann, H. So, H. Steinbeiss, CIRP Ann. 2007, 56, 269.
[4] H. Karbasian, A. E. Tekkaya, J. Mater. Process. Technol. 2010, 210, 2103.
[5] H. Steinbeiss, H. So, T. Michelitsch, H. Hoffmann, Prod. Eng. 2007, 1, 149.
[6] J. Sander, J. Hufenbach, L. Giebeler, H. Wendrock, U. Kühn, J. Eckert, Mater. Des. 2016, 89, 335.
[7] S. Kugaevskii, E. Pizhenkov, A. Gamberg, Mater. Today: Proc. 2019, https://doi.org/10.1016/j.matpr.2019.07.055.
[8] A. Armilotta, R. Baraggi, S. Fasoli, Int. J. Adv. Manuf. Technol. 2014, 71, 573.
[9] R. Hölker-Jäger, A. E. Tekkaya, Laser Additive Manufacturing (Ed: M. Brandt), Woodhead Publishing, Sawston, UK 2017, pp. 439–464.
[10] T. DebRoy, H. L. Wei, J. S. Zuback, T. Mukherjee, J. W. Elmer, J. O. Milewski, A. M. Beese, A. Wilson-Heid, A. De, W. Zhang, Prog. Mater. Sci. 2018, 92, 112.
[11] L. Parry, I. A. Ashcroft, R. D. Wildman, Addit. Manuf. 2016, 12, 1.
[12] A. Olleak, Z. Xi, Procedia Manuf. 2019, 34, 613.
[13] X. Li, Y. H. Tan, H. J. Willy, P. Wang, W. Lu, M. Cagirici, C. Y. A. Ong, T. S. Herng, J. Wei, J. Ding, Mater. Des. 2019, 178, 107881.
[14] H. Fayazfar, M. Salarian, A. Rogalsky, D. Sarker, P. Russo, V. Paserin, E. Toyserkani, Mater. Des. 2018, 144, 98.
[15] H. Berns, W. Theisen, G. Scheibelein, Ferrous Materials: Steel and cast iron, Springer, Berlin 2008.
[16] R. Casati, M. Coduri, N. Lecis, A. Andriano poli, M. Vedani, Mater. Charact. 2018, 137, 50.
[17] P. Krakhmalev, I. Yadroitseva, G. Fredriksson, I. Yadroitsev, Mater. Des. 2015, 87, 380.
[18] J. Boes, A. Röttger, C. Mutke, C. Escher, W. Theisen, *Addit. Manuf.*, 2018, 23, 170.
[19] M. Garibaldi, I. Ashcroft, M. Simonelli, R. Hague, *Acta Mater.* 2016, 110, 207.
[20] Y. Zhong, L. Liu, S. Wikman, D. Cui, Z. Shen, *J. Nucl. Mater.* 2016, 470, 170.
[21] Z. H. Liu, D. Q. Zhang, C. K. Chua, K. F. Leong, *Mater. Charact.* 2013, 84, 72.
[22] D. Bourell, J. P. Kruth, M. Leu, G. Levy, D. Rosen, A. M. Beese, A. Clare, *CIRP Ann.* 2017, 66, 659.
[23] A. Foroozmehr, M. Badrossamay, E. Foroozmehr, S. Golabi, *Mater. Des.* 2016, 89, 255.
[24] L. Liu, Q. Ding, Y. Zhong, J. Zou, J. Wu, Y.-L. Chiu, J. Li, Z. Zhang, Q. Yu, Z. Shen, *Mater. Today* 2018, 21, 354.
[25] K. Saeidi, X. Gao, Y. Zhong, Z. J. Shen, *Mater. Sci. Eng. A* 2015, 625, 221.
[26] W. E. King, H. D. Barth, V. M. Castillo, G. F. Gallegos, J. W. Gibbs, D. E. Hahn, C. Kamath, A. M. Rubenchik, *J. Mater. Process. Technol.* 2014, 214, 2915.
[27] J.-P. Kruth, G. Levy, F. Kocke, T. Childs, *CIRP Ann.* 2007, 56, 730.
[28] B. Ferrar, L. Mullen, E. Jones, R. Stamp, C. J. Sutcliffe, *J. Mater. Process. Technol.* 2012, 212, 355.
[29] R. Mertens, B. Vrancken, N. Holmstock, Y. Kinds, J.-P. Kruth, J. van Humbeeck, *Phys. Procedia* 2016, 83, 882.
[30] J. Krell, A. Röttger, K. Geenen, W. Theisen, *J. Mater. Process. Technol.* 2018, 255, 679.
[31] P. A. Hooper, *Addit. Manuf.* 2018, 22, 548.