Microstructure evolution during selective laser melting of metallic materials: A review

Cite as: J. Laser Appl. 31, 031201 (2019); https://doi.org/10.2351/1.5085206
Submitted: 10 December 2018 . Accepted: 13 May 2019 . Published Online: 31 May 2019

Xing Zhang, Christopher J. Yocom, Bo Mao, and Yiliang Liao
Microstructure evolution during selective laser melting of metallic materials: A review

Cite as: J. Laser Appl. 31, 031201 (2019); doi: 10.2351/1.5085206
Submitted: 10 December 2018 · Accepted: 13 May 2019 · Published Online: 31 May 2019

Xing Zhang, Christopher J. Yocom, Bo Mao, and Yiliang Liao

AFFILIATIONS
Department of Mechanical Engineering, University of Nevada, Reno, Nevada 89557

April 30, 2019

ABSTRACT
Selective laser melting (SLM) is an additive manufacturing technology that uses a laser beam to melt powder materials together layer by layer for three-dimensional (3D) solid part fabrication. Due to its superior rapid prototyping capability of three-dimensional structures, SLM has been used for widespread industrial applications including aerospace, automotive, electronics, and biomedical devices. As a state-of-the-art technology, ongoing investigations are being conducted to improve the efficiency and effectiveness of SLM. In particular, understanding of microstructure evolution during SLM is essential to achieve improved process control and ensure the performance of laser-fabricated components. This paper is to review the recent research and development progress in SLM of metallic materials with a focus on the process–microstructure relationship. The grain growth and porosity evolution as affected by laser processing parameters in the SLM process are discussed. Phase transformation in SLM of steel and titanium alloys is studied. The formation of precipitates in SLM of titanium, nickel, and aluminum/magnesium alloys is reviewed. The balling phenomenon and cracking behaviors during SLM are discussed. In addition, the recent development of computational modeling of microstructure evolution during SLM is investigated.

Key words: selective laser melting, microstructure evolution, grain growth, porosity, computational modeling

© 2019 Laser Institute of America.

I. INTRODUCTION
Selective laser melting (SLM) is an additive manufacturing technology that uses a laser beam to melt powder materials together layer by layer for three-dimensional (3D) solid part fabrication. At the beginning of the process, a 3D computer-aided design model describing both internal and external geometry features of the part is developed and sliced into a stack of 2D layers, which are loaded into the SLM equipment. A thin layer of powder material is then spread on a substrate plate via a powder delivery system and later melted together with a high-energy laser beam scanning over the selected area. By applying another thin layer of powder on top of the previously processed layer and repeating the scanning process, the successive layers are formed and metallurgically bonded with the previous layer. The designed part is, thus, fabricated layer by layer as shown in Fig. 1. Compared to conventional manufacturing techniques such as casting and forging, SLM holds advantages including (1) rapid prototyping, (2) the capability of producing components with customized or complex geometry, (3) the ability to produce novel structural designs for increased functionality, and (4) the joining of dissimilar metals through metallurgical means.

Due to these advantages, SLM has been used for widespread practical industrial applications, such as aerospace, automotive, electronics, and biomedical devices. Yakout et al. produced an airfoil blade of stainless steel 316L using SLM. Song et al. manufactured the metal die of an automobile deck part at a low cost and a short process cycle. Wits et al. utilized SLM to fabricate a low-weight and high-pressure micropump for small satellite applications. Hou et al., Wong et al., and Zenou et al. employed this technology to manufacture electronics such as printed circuit boards, heat sinks, and transparent conductors. SLM has also been applied in the medical field to manufacture customized bioimplants. The most recent application of SLM is to fabricate novel materials such as functionally graded materials and metal matrix nanocomposites.
However, SLM is a complicated process that suffers from the intense laser–material interaction, molten pool instability, large degree of shrinkage, and complex residual stress. The properties and performance of a component fabricated via SLM are closely related to microstructure characteristics and determined by various considerations. From the perspective of metallurgical bonding, specific considerations should be given to the wetting behavior, Marangoni effect, and capillary forces. In addition, rapid thermal cycles and directional solidification can generate microstructure that differs from conventionally processed materials.

More importantly, microstructure evolution is highly affected by laser processing conditions such as laser parameters and scanning strategy. Therefore, precision control of the microstructure evolution during SLM remains a major challenge, which significantly affects the performance of as-built parts.

With the increased demand for SLM methods in manufacturing, it is of primary importance to gain a better understanding of the process–microstructure relationship in order to achieve improved process control and ensure the performance of laser-fabricated components. This paper is to review the recent research and development progress in SLM of metallic materials with a focus on the process–microstructure relationship. The evolution of grain and pore structures are studied by reviewing previously conducted experiments, along with the effects of various process parameters. Phase transformation in SLM of steel and titanium alloys is studied. The formation of precipitates in SLM of titanium, nickel, and aluminum/magnesium (Al/Mg) alloys is reviewed. The balling phenomenon and cracking behaviors during SLM are discussed. In addition, computational modeling methodologies for predicting microstructure evolution during SLM are reviewed with a focus on the growth of the grain structure. The knowledge gained from this review paper could provide insights and guidance for optimization and further development of the SLM process.

II. GRAIN GROWTH IN THE SLM PROCESS

A. Mechanism of grain growth

During SLM of metallic materials, the evolution of the grain structure in terms of grain size, morphology, and orientation is considered as the key microstructure feature determining the mechanical performance of 3D-printed parts. Similar to traditional welding processes, the grain growth during SLM is a localized solidification process that is initiated by epitaxial growth proceeded by competitive growth toward the center of the melt pool. As the
traveling laser beam interacts with the material, a melt pool forms. After the laser beam leaves the area, solidification takes place via epitaxial growth of the partially melted grains from the substrate or previous layer. The shape of the liquid–solid interface in terms of planar, cellular, columnar dendritic, or equiaxed dendritic fronts is governed by the temperature gradient \( G \), solidification rate \( R \), undercooling \( \Delta T \), and solute diffusion coefficient \( D_s \) (Fig. 3).\(^{27-30}\)

According to the solidification theory, the grain morphology and substructure (fineness/size) are strongly affected by the ratios \( G/R \) and \( G-R \) (cooling rate), respectively.\(^{31,32}\) At the bottom of the melt pool where the solidification rate \( R = 0 \), the ratio of \( G/R \) is infinite, leading to a planar-dominated grain structure. As the ratio, \( G/R \) decreases until \( G/R < \Delta T/D_s \), the cellular solidification takes place near the planar region. Further decreasing of the \( G/R \) ratio leads to the columnar dendritic formation. The equiaxed solidification typically happens near the surface of the melt pool where the temperature gradient is relatively low.

In addition to the grain morphology, the grain growth directions are aligned in the thermal gradient direction as the solidification proceeds. As shown in Fig. 4, at the early stage of solidification, the epitaxial grains have multiple orientations. However, the crystallographic effect affects the grain structure by favoring growth along particular crystallographic orientations.\(^{31,32}\) For instance, the metallic materials with the face-centered cubic structure have a preferred growth along the (100) crystallographic orientations. Thus, the grains with the same or close orientation as the temperature gradient grow faster, while the grains with other orientations gradually stop growing due to the competitive growth of different grains.\(^{33}\) In addition, the horizontal grain growth is suppressed when the adjacent grains impinge on each other, and the high rate of elongated grain growth results in little chance of nucleation occurring in the liquid ahead of the columnar zone.\(^{34}\) As a result, this competitive growth phenomenon leads to the unidirectional columnar grain structure during solidification, which is a typical microstructure observed in SLM-built products. Nevertheless, by adjusting laser parameters and/or applying methods to promote nucleation (such as adding impurities like nanomaterials and utilizing dynamic approaches like arc oscillation), the equiaxed dendritic growth can take place due to the massive increase of multiorientation nucleation sites, resulting in finer grain structures.\(^{35-43}\)

**B. Effects of laser parameters and scanning strategies on grain growth**

In order to investigate the effect of laser parameters on microstructure evolution during SLM, researchers usually consider a combined parameter—energy density, which can be expressed as

\[
E = P/(\nu \times h \times t),
\]

where \( P \) is the laser power, \( \nu \) is the scanning speed, \( h \) is the hatch spacing, and \( t \) is the layer thickness. These laser parameters play important roles in determining the temperature gradient, cooling rate, and local heat flow direction, which in turn considerably affect the grain morphology, size, and orientation. Monroy et al.\(^{44}\) studied the impacts of laser power and scanning speed on the grain size during SLM of cobalt alloys, and Sun et al.\(^{45}\) conducted similar investigation on stainless steel 316L. It was found that the grain size increased with increasing laser power due to a higher energy-induced coarsening effect, and the grain size decreased with increasing scanning speed due to shorter exposure time and faster solidification. Yadroitsev et al.\(^{46}\) characterized the grain growth of stainless steel 316L during SLM in terms of primary cell spacing and misorientation angle, which are defined as the distance between adjacent grains and the angle between the laser traveling direction and the grain growth direction, respectively. It was reported that the cell spacing decreased by 25%–40% as the scanning speed increased from 0.08 to 0.28 m/s. In addition, two regions with different misorientation angles (58°–69° and 10°) were
found while processing with a low scanning speed. On the other hand, increasing the scanning speed resulted in only one misorientation angle of 58°–69°. Dezfoli et al.\textsuperscript{34} obtained similar conclusions describing the influence of scanning speed during SLM on grain size and grain orientation of Ti64 (Ti-6Al-4V). In addition, the effect of hatching spacing on grain morphology of SLM-manufactured CoCrMo alloys was analyzed by Olsen et al.\textsuperscript{47}

In the SLM process, small equiaxed grains are often found in the center of the melt tracks due to the constitutional undercooling. As the hatch spacing decreased, the new melt track would remelt the center area, leading to large epitaxial grain growth across the entire area.

Besides laser parameters, scanning strategies play another critical role in determining the heat flow direction and, thus, the grain growth direction. Generally, as illustrated in Fig. 5, four different scanning strategies\textsuperscript{48} can be applied in the SLM process: unidirectional, zigzag, crosshatching, and “island” scanning strategies. Thijs et al.\textsuperscript{49} investigated the grain growth during SLM of Ti64 as affected by scanning strategies including unidirectional, zigzag, and crosshatching. It was found that all methods produced identical grain growth directions. In specific, for the unidirectional scanning strategy [as observed in the side view of Fig. 6(b)], the elongated grains tilted 19° away from the building direction. For the zigzag scanning, a grain structure with a herringbone pattern was observed [Fig. 6(c)] due to the alternating scanning direction, and elongated grains [Fig. 6(d)] were found to be generally tilted along the build direction. For the crosshatching strategy, the grains became more equiaxed, and a grain structure of the grid plan pattern was observed [Fig. 6(e)] while two grain growth directions, parallel to the building direction and tilted at 25°, were identified [Fig. 6(f)]. Arisoy et al.\textsuperscript{50} conducted SLM experiments of nickel alloy IN625 using cross-hatching strategy with various rotation angles between consecutive layers. It was reported that compared to the crosshatching with a rotation angle of 90°, the scanning strategy with a rotation angle of 67° resulted in a reduction of the average grain size, and the grain growth directions were less affected by laser power and scanning speed. Carter et al.\textsuperscript{51} studied the
influence of island scanning strategy on the grain structure of SLM-fabricated nickel superalloy CM247LC. A bimodal grain structure with large elongated columnar grains in the building direction and fine grains in the edges of islands was observed. It was claimed that each scanning island was heated by the band heating effect instead of point heating, resulting in the appearance of fine grains to the higher cooling rate at the edges of islands due to surrounding cold material.

C. Effects of additives on grain growth

The growth of the grain structure during SLM can be significantly affected by adding additives into the powder bed, such as second phase nanoparticles. By providing a large amount of heterogeneous nucleation sites in front of the liquid–solid interface, suitable nanoparticles can considerably decrease the critical undercooling needed for producing ideal equiaxed structures.32 Such effect of nanoparticles on microstructure evolution was firstly observed in other laser-based techniques such as laser cladding and laser surface alloying. Li et al.33 compared the grain structure of cobalt-based alloy coatings manufactured by laser cladding with and without nano-Y2O3. It was found that for the as-built part without nanoadditives, the directional columnar dendrites growing along the temperature gradient direction can be easily observed at the liquid–solid interface [Fig. 7(a)]. Contrarily, by mixing cobalt powders with Y2O3 nanoparticles (0.5 wt. %), multiorientation equiaxed dendrites formed rather than columnar dendrites, as shown in Fig. 7(b). It was reported that such “columnar-to-equiaxed transition” became more evident by increasing nanoparticle quantities. For instance, the equiaxed grains can be observed across the entire section by adding 1.0 wt. % of nano-Y2O3. Zhang et al.34 conducted similar experiments to study the effect of CeO2 microparticles/nanoparticles on the grain growth of Ni-based alloy coatings fabricated using laser cladding. It was found that a coating layer with a finer grain structure can be achieved by adding nanoadditives, as compared to use microadditives. Later on, nanoparticles have been introduced to the SLM process to manufacture nanocomposites with excellent mechanical properties. Li et al.35 studied SLM of TiB2 nanoparticle-decorated A1Si10Mg alloys. It was concluded that by adding TiB2 nanoparticles, the obtained average grain size (2 μm) was much smaller than that of the as-built sample without nanoparticles (~10 μm). Recently, a high-strength aluminum alloy (AA) with a fine equiaxed grain structure was 3D-printed using SLM, particularly through mixing AA micro-powders with hydrogen-stabilized zirconium nanoparticles.36 Ma et al.37 employed SLM to manufacture Ni/Al2O3 nanocomposite, and it was proved that nanoparticles could also prevent the grain coarsening within the heat-affected zone (substrate or previous layer) by altering thermal conditions. However, it is also worth mentioning that some studies have shown that nanoparticles do not necessarily lead to grain refinement. For example, Streubel et al.38 found that the addition of nano-Y2O3 did not affect the grain morphology of as-built steel part, which was attributed to the poor wettability and lattice mismatching between the nanoparticle and metal matrix. Clearly, the role of nanoparticles in microstructure evolution during SLM still needs to be further investigated.

D. Controlled grain growth by other strategies

Coupling SLM with dynamic approaches (such as electromagnetic vibration, ultrasonic vibration, transverse arc oscillation, etc.) might serve as an alternative methodology to tune the grain structure of SLM-fabricated components. For instance, Huang et al.39 obtained Ti64 with finer grains by integrating SLM with electromagnetic vibrations. It was claimed the electromagnetic vibration-induced electromagnetic force effect, Peltier effect, and Joule heating effect may all contribute to the grain refinement. To date, little effort has been put into the investigation of coupling SLM process with dynamic approaches. However, it is worth mentioning that researchers have successfully conducted experiments on similar processes like laser surface melting and welding, which present a promising future for introducing various dynamic approaches to SLM. Zhou et al.40 examined the effect of coupling laser surface melting with an electromagnetic field (EMF) on the microstructure evolution of AZ91D magnesium alloys. As shown in Fig. 8, it was observed that without EMF, the melted layer mainly consisted of columnar dendritic and cellular grains; while with EMF, the melted layer had a uniform microstructure with finer equiaxed grains. Xu et al.41 obtained fine grain structure through ultrasonic vibration-assisted laser surface melting. It was found that with the increase of vibration amplitude, more dendrites were broken up and functioned as nuclei due to ultrasonic cavitation and acoustic streaming stirring. As the vibration amplitude approached 80%, the columnar crystal almost disappeared. Yuan et al.42 conducted welding experiments with transverse arc oscillation to refine the grain structure of AZ31B magnesium alloys. In this work, the effects of oscillation amplitude,
frequency, and torch travel speed on grain refining were investigated. It was claimed that the grain refinement was attributed to the dendrite fragmentation induced by the remelting of dendrite arms rather than the mechanical breakup.

Due to the rapid solicitation process, directional columnar dendrite is the typical grain morphology in SLM-built parts, leading to anisotropic mechanical properties. Recent research indicates that altering the powder composition could serve as an alternative method to obtain ultrafine grains and address the anisotropic effect in SLM. For example, Montero-Sistiaga et al. investigated the effects of Si on the grain size distribution during SLM of Al7075, as shown in Fig. 9. It was found that the grain size was reduced from 50–100 to 5 μm by adding 4 wt. % Si. Further investigations are needed to elucidate the mechanism responsible for the powder composition effect on the grain growth in SLM.

III. EVOLUTION OF PORE STRUCTURES IN THE SLM PROCESS

In addition to the evolution of grains, the formation of pore structures during SLM needs to be investigated in order for achieving 3D-printed parts with desired properties. Considerable research and development effort has been put on SLM process design and optimization to eliminate or minimize the porosity, since a high material density ensures the mechanical integrity, which is critical for widespread industrial applications, particularly for those used as load-bearing structural components. On the other hand, SLM of porous materials attracts attention in recent years, since the pore structures with designed pore size and density are promising for applications in thermal management and mass/energy transport. Therefore, it is of specific importance to gain an understanding of the mechanism responsible for the formation of pore structures during SLM of metallic materials.

A. Porosity-formation mechanism during SLM

Similar to conventional metal casting processes, the formation of pore structures in SLM is mainly attributed to the liquid–solid phase transformation during solidification, which is strongly affected by various factors including the thermal history (cooling rate), material intrinsic properties, part geometry, etc. In a study conducted by Thijs et al., it was found that the porosity evolution in SLM of Ti64 was strongly affected by the accumulation of surface roughness across the layers. While for SLM of AlSi10Mg, deep keyholes were observed at the start/ending points of the scanning track due to the heat accumulation. These keyholes with a low thermal stability collapse during laser processing, resulting in the formation of large pores. Vilaro et al. contributed the small spherical pores in SLM-fabricated Ti64 to the existence of gas that may dissolve in the melt pool and remain there due to the rapid solidification. Qiu et al. observed open pores on the melting surface of SLM-fabricated AlSi10Mg and...
claimed that the formation of open pores was attributed to insufficient feeding of molten material to solidification fronts and/or the incomplete remelting of the localized surface. They also argued that the violent interaction between laser and material would make the melt pool become turbulent, causing splashing, and resulting in high porosity. Panwisawas et al. concluded that for SLM of Ti64, the spherical or ellipsoidal pore formation depended on the interacting domain area and flow velocity direction, and the flat or elongated internal pore formation was due to the incomplete bonding between layers. In addition, Gong et al. claimed that the excessive temperature would induce serious evaporation and form gas bubbles. The bubbles far beneath the surface may not have sufficient time to rise and escape during laser processing, leading to the formation of pores.

B. Effects of laser parameters on porosity evolution

It is well accepted that the porosity is greatly influenced by the energy density related parameters, such as laser power, scanning speed, hatch spacing, and layer thickness, since these parameters have significant impacts on the temperature evolution and liquid/gas flow dynamics within the processed area. As shown in Fig. 10, there are optimal windows for laser power and scanning speed to obtain minimized porosity during SLM of Ti64. It was observed that given a low laser power or a high scanning speed (low-energy density), irregular pores or crevices were formed due to insufficient melting and the balling phenomenon [Fig. 11(a)]; while given a high laser power or a low scanning speed (high-energy density), spherical pores were generated due to evaporated material and dissolved inert gases becoming trapped in the melted material during the solidification process [Fig. 11(c)]. In addition, the porosity typically increases with the widening of hatch spacing and thickening of the powder layer. Aboulkhair et al. investigated the effect of changing hatch spacing from 50 to 250 μm on the porosity during SLM of AlSi10Mg. The gaps between adjacent tracks were clearly observed once the hatch spacing exceeded 150 μm due to the reduced intralayer bonding. Stoffregen et al. found that the porosity of SLM-built stainless steel part increased from 2.0% to 11.4% as the layer thickness increased from 20 to 80 μm due to the insufficient penetration depth of laser energy. Ma et al. suggested that the powder layer thickness should be less than 100 μm such that the residual gas at the bottom of the melt pool can escape in time during the rapid solidification.

C. Porosity evolution as affected by scanning strategies

Scanning strategy is another key factor that affects the porosity in SLM-built parts. As aforementioned, pores can be formed within the unstable melt pool during solidification via various mechanisms. For the unidirectional and zigzag scanning strategies, a longer scanning distance increases the instability of the melt pool and leads to the accumulation of defects like porosity. To address this issue, a double-scanning strategy was developed to reduce the porosity by remelting the solidified layer or melting the remaining powder particles. Alcoutlab et al. investigated the effect of double-scanning strategy on porosity during SLM of AlSi10Mg. It was found that the second scanning reduced porosity by 4% at most. However, the double-scanning could introduce excessive thermal energy, leading to the generation of additional pores. An alternative solution, called the “refill” or “knitting” strategy, is proposed to reduce porosity within the overlapping zone. During the SLM process, the surface tension tends to drive the geometry of the melt material toward a spherical shape, leading to the porosity between scanning tracks. The knitting strategy is utilized to refill the pores/gaps in the overlapping area by scanning the next layer at an offset position. The cross-hatching strategy employs a similar technique as the method of repairing defects between adjacent scanning tracks in the previous layer, resulting in a stronger interlayer bonding and a lower porosity. The island-scanning strategy, which was developed to effectively decrease the residual stress at first, may reduce the porosity during the SLM process as well because of short scanning distances. It was found that the effectiveness of reducing porosity using island-scanning strategy was strongly affected by the individual island size.

D. Porosity evolution as affected by powder characteristics

In addition to laser parameters and scanning strategy, powder characteristics in terms of size distribution, morphology, and impurities have considerable influences on the formation of pore structures during SLM. For instance, surfaces of SLM-built parts fabricated using different sizes of M2 high speed steel powders are shown in Fig. 12. Micropores were clearly observed in SLM-built parts if the...
powder size was too large (>150 μm), since the laser energy input was not sufficient to fully melt powder particles to form a dense structure [Fig. 12(b)]. Moreover, coarse powders can lead to large voids among particles.77 On the other hand, micropores were also easily found in the parts if using the powders with a size too small (<38 μm) [Fig. 12(c)]. This is attributed to the enhanced laser–material interaction induced by higher surface energy or aggravating gas entrapment, and dewetting behavior induced by a higher level of oxygen.19,76 Therefore, it was concluded that a proper powder size needs to be selected in order for minimizing the porosity during SLM. Li et al.78 investigated the densification behavior of gas-atomized 316L stainless steel powders during SLM. It was found that the samples fabricated using gas-atomized powders exhibited less porosity as compared to those produced from water-atomized powders. This was accredited to the lower oxygen content and higher packing density of gas-atomized powder.

IV. PHASE TRANSFORMATION AND PRECIPITATION

A. Phase transformation during SLM of steels and titanium alloys

For SLM of steels and titanium alloys, the phase transformation plays another important role in determining the performance of laser 3D-printed parts. The formation and distribution of various phases in as-built parts are governed by the rapid heating and cooling cycles during laser processing.

1. SLM of steels

During SLM of steels, the phase transformation is highly determined by the heating-induced austenitization and cooling-induced self-quenching effect. Normally, SLM of austenitic steel powders, such as 304L and 316L, leads to a fully austenitic microstructure due to the rapid melting and cooling cycles.79,80 However, Yadollahi et al.81 observed a small volume fraction of δ ferrite (5–10 vol. %) distributed homogeneously within the austenite grains for direct laser deposited 316L stainless steel. It was claimed that the generation of ferrite phase was attributed to the combing effect of rapid solidification and chemical composition. For martensitic steels, it was reported that different amounts of metastable austenite were observed after SLM.82–85 Faccgini et al.82 identified 72 vol. % metastable austenite and 28 vol. % twinned martensite in the parts fabricated by SLM using precipitation-hardening stainless steel powders (17-4PH). Given different laser processing parameters, LeBrun et al.83 found 36 vol. % austenite in the SLM-fabricated 17-4PH samples. It was claimed that the
residual thermal stress in the bulk material was the reason for the formation of the mechanically stabilized metastable austenite phase. SLM of AISI 420 stainless steel was conducted by Krakhlmalev et al., and it was found that the austenite volume fraction varied with processing regions due to different temperature fields during laser processing. The upper layers exhibited 21 vol. % retained austenite, while the lower layers had a higher amount of austenite (57 vol.%) due to austenite reversion or growth of retained austenite, which was induced by in situ carbon partitioning and the diffusion process. A similar phenomenon in terms of retained austenite was also observed in SLM of other martensitic steels such as 18-Ni300 (Ref. 84) and AISI420.85

In addition, the effect of powder composition on phase transformation during SLM of steels was studied by Jerrard et al.86 By varying the ratio of 316L and 17-4PH stainless steel powder mixtures, it was found that with the increase of 17-4PH content, the precipitation of \( \alpha \)-phase ferrite and cementite (Fe,C) appeared due to the existence of interface boundary between austenitic and martensitic grains. Once the weight percentage of 17-4PH powders was increased to 15 wt. % Nb, both \( \alpha \)-phase ferrite and cementite (Fe,C) appeared due to the faster cooling condition.

For SLM of titanium and its alloys, various phases including \( \alpha \) (hcp), \( \beta \) (bcc), \( \alpha^\prime \) (hcp martensite), and \( \alpha^\prime\prime \) (f-c-orthorhombic) can be manufactured.86 Chen et al.87 conducted SLM of Ti64 using the commercial Ti64 powders, consisting of hcp \( \alpha \)-Ti phase and bcc \( \beta \)-Ti phase. It was found that the \( \beta \) phase was transformed into martensitic \( \alpha^\prime \) phase after laser processing. The similar results were also reported in Ref. 88. Gu et al.89 studied the effect of laser scanning speed on the phase transformation during SLM of pure Ti. It was reported that a high laser scanning speed (400 mm/s) resulted in the formation of martensitic \( \alpha^\prime \) phase. Xu et al.90 investigated how the phase transformation of SLM of Ti64 was affected by the layer thickness, focal offset distance, and energy density. It was found that the reduction of layer thickness resulted in finer \( \beta \) grains due to the faster cooling condition. By decreasing the focal offset distance or increasing the energy density, an ultrafine lamellar \( \alpha^\prime \) structure, rather than \( \alpha^\prime \) martensite, was gradually prevailed. For SLM of Ti-Nb alloys, Wang et al.91 studied the effect of Nb content on the phase transformation. Given a Ti-Nb powder mixture of 15 wt. % Nb, both \( \alpha^\prime \) and \( \beta \) phases were observed in the SLM-built parts. As the Nb content increased, the stability of the \( \beta \) phase was enhanced and martensitic transformation (\( \beta \rightarrow \alpha^\prime \)) was restricted. Once the Nb content increased to 25 at. %, only the \( \beta \) phase was observed.

2. SLM of titanium and its alloys

For SLM of titanium (Ti) and its alloys, various phases including \( \alpha \) (hcp), \( \beta \) (bcc), \( \alpha^\prime \) (hcp martensite), and \( \alpha^\prime\prime \) (f-c-orthorhombic) can be manufactured.86 Chen et al.87 conducted SLM of Ti64 using the commercial Ti64 powders, consisting of hcp \( \alpha \)-Ti phase and bcc \( \beta \)-Ti phase. It was found that the \( \beta \) phase was transformed into martensitic \( \alpha^\prime \) phase after laser processing. The similar results were also reported in Ref. 88. Gu et al.89 studied the effect of laser scanning speed on the phase transformation during SLM of pure Ti. It was reported that a high laser scanning speed (400 mm/s) resulted in the formation of martensitic \( \alpha^\prime \) phase. Xu et al.90 investigated how the phase transformation of SLM of Ti64 was affected by the layer thickness, focal offset distance, and energy density. It was found that the reduction of layer thickness resulted in finer \( \beta \) grains due to the faster cooling condition. By decreasing the focal offset distance or increasing the energy density, an ultrafine lamellar \( \alpha^\prime \) structure, rather than \( \alpha^\prime \) martensite, was gradually prevailed. For SLM of Ti-Nb alloys, Wang et al.91 studied the effect of Nb content on the phase transformation. Given a Ti-Nb powder mixture of 15 wt. % Nb, both \( \alpha^\prime \) and \( \beta \) phases were observed in the SLM-built parts. As the Nb content increased, the stability of the \( \beta \) phase was enhanced and martensitic transformation (\( \beta \rightarrow \alpha^\prime \)) was restricted. Once the Nb content increased to 25 at. %, only the \( \beta \) phase was observed.

B. Precipitation during SLM

Precipitation can take place simultaneously with phase transformation during SLM of titanium alloys. It is also commonly observed in SLM-built nickel, aluminum, and magnesium alloys. The type and characteristics of precipitates which affect the mechanical properties of as-built parts can be controlled by adjusting laser processing parameters and/or the powder chemical composition.

For SLM of titanium alloys, Chlebus et al.92 observed uniformly distributed precipitates of the \( \beta \)-AlNbTi2 phase in SLM-fabricated Ti-6Al-7Nb alloys. It was claimed that such uniform distribution of precipitates might be attributed to the microsegregation of solute in \( \beta \) grains, leading to the enhanced tensile and compressive strengths of the as-built parts due to the precipitation-hardening effect. Li et al.93 claimed the precipitation kinetics in SLM of Ti-45Al-2Cr-5Nb alloys as \( \beta \rightarrow \alpha_1 \) (Ti3Al) + \( \gamma \) (TiAl) and disorder–order transformation (\( \beta \rightarrow B_2 \)). It was found that by increasing the laser scanning speed, the contents of B2 and \( \gamma \) precipitates were increased with a decrease of the content of the \( \alpha_2 \) matrix phase.

In the study of SLM of Ni-based superalloys, Idell et al.94 and Jia and Gu95 both identified the nucleation of \( \gamma \) phase precipitates [Ni3(A1,Ti)], while the unexpected \( \delta \)-phase precipitates (Ni6Nb) were also found in the matrix due to the significant microsegregation of niobium caused by rapid heating and cooling cycles. It was claimed that the nucleation of \( \delta \)-phase precipitates would be restricted when the local content of niobium was less than 6.8% molar fraction in the SLM process. It was also concluded that by increasing the laser energy, the coarsening effect led to a larger size of \( \gamma \) precipitates, and as the sample building up, the previous layer experienced an aging treatment, which further promoted the formation of the \( \gamma \) phase.96

For SLM of Al/Mg alloys, the final products usually consist of primary \( \alpha \)-Al/\( \alpha \)-Mg and intermetallic phases depending on the chemical composition of alloy powder.97–99 Sun et al.96 optimized the laser energy in SLM of Al-8.5Fe-1.3V-1.7Si alloys to obtain high cooling rates (>10⁵ °C/s), leading to the formation of \( \alpha_1 \) phase rather than a brittle \( \beta'-\)Al4FeSi2 phase, which forms at low cooling rates (<10³ °C/s). The Al-20Si-5Fe-3Cu-1Mg alloy was manufactured using SLM by Ma et al.93 It was found that only \( \delta \)-AlFeSi phase was observed in the as-built part, while the as-cast alloy typically consisted of both \( \beta'_1 \)-Al2FeSi and \( \delta \)-AlFeSi2. Shuai et al.94 attributed the precipitation of Mg7Zn3 in the SLM-fabricated Mg alloy ZK60 to the rapid solidification, which hindered the decomposition of this high temperature phase (Mg7Zn3 → \( \alpha \)-Mg + MgZn). It was claimed that the uniformly distributed Mg7Zn3 phase in ZK60 could contribute to the improvement of strength.

In addition to the aforementioned different precipitates, the rapid cooling rate of the SLM process can also lead to the formation of a supersaturated solid solution and suppress the precipitation, especially for Al-Si and Al-Sc alloys. Maehshima and Oh-ishi100 conducted SLM experiments of AlSi10Mg material and found that the Si content in \( \alpha \)-Al was 2.2 at. %, which significantly exceeded the maximum solubility of 1.4 at. % in the equilibrium state of AlSi10Mg. Zhang et al.101 investigated the microstructure evolution during SLM of Sc modified Al–Mg alloys. It was found that Al7Sc2Zr precipitates were obtained at the bottom of molten pool at a scanning speed of 1800 mm/s, while increasing the scanning speed to 3000 mm/s resulted in the formation of supersaturated solid solution and no precipitates were observed. Moreover, the as-built supersaturated alloys can be used for subsequent heat treatment to improve mechanical properties, which deserves further investigation.
V. BALLING AND CRACKING

Apart from the evolution of grain structure, pore structure and metallic phases during SLM, various microstructure defects need to be studied, as those defects could result in poor quality, low dimension accuracy, and limited mechanical properties of as-built parts.

A. Balling phenomenon

Balling effect 102–111 as shown in Fig. 13, is a common phenomenon during SLM, which is detrimental to part formation. It can lead to discontinuous tracks and incomplete interline bonding, as well as the porosity, and even delamination. Generally, balling phenomenon takes place with a decrease of surface energy, but the formation mechanisms are rather complicated.

Gu and Shen 110 conducted an experimental study on SLM of 316L stainless steel, in which the balling phenomenon was considered as two categories based on the ball size. The coarsened balls were caused by the limited liquid formation due to a low laser energy input, which resulted in a high viscosity of the liquid–solid mixture and, thus, hindered the liquid flow and particle arrangement, eventually leading to the aggregation of particles and the formation of balls in the scale of laser spot size. On the other hand, the micrometer-scaled balls (∼10 μm) were formed by small droplets splashing from the surface of unstable molten tracks when laser processing was conducted with a high scanning speed. Niu and Chang 106 concluded that a high laser scanning speed during SLM would promote the breaking up of molten tracks into balls. Zhou et al. 100 studied the balling phenomenon in SLM of tungsten. It was claimed that the slow wetting and spreading speed of tungsten, caused by its high surface tension and viscosity, were the main reasons for the large balling tendency. In addition, the oxidation of tungsten melt pool also affected the surface tension gradient and the thermocapillary convection, which favored the balling initiation. Das 109 investigated how the balling phenomenon was affected by the wettability. It was concluded that the oxidation on the substrate can significantly reduce the wettability of liquid metal. As a result, the molten material tended to form spheres to decrease surface energy. SLM of 316L stainless was also investigated experimentally by Shen et al. 112 and it was observed that the balls formed during the initial scanning track were larger than the ones formed during the subsequent scanning tracks. Such a phenomenon was explained by the sharp temperature gradient between the heated area and unprocessed area for the initial scan, and lower temperature gradient generated for the subsequent scanning due to the overlap between scanning tracks. The effects of laser power, scanning speed, layer thickness, and oxygen content on balling phenomenon were discussed thoroughly by Gu and Shen 110 and Li et al. 113 In general, by adjusting laser processing parameters properly, such as increasing the laser power or reducing the scanning speed, layer thickness, hatch spacing, and/or oxidation, the balling effect during SLM process can be alleviated.

B. Cracking

Cracking is another major concern associated with the rapid heating and cooling cycles, large solidification shrinkage, and high coefficient of thermal expansion. 19,43,113,114 Given its complicated forming mechanisms, it is usually difficult to eliminate cracks only by the optimization of SLM processing parameters. 115,116 Carter et al. 116 characterized the cracks in SLM-fabricated Ni-based superalloy CM247LC into two categories: solidification cracks growing both longitudinally and transversely to the build direction [Fig. 14(a)] and grain boundary cracks lying along grain boundaries [Fig. 14(b)]. Three mechanisms, including liquation cracking, strain-age cracking, and ductility-dip cracking, were proposed to explain the formation of grain boundary cracks. The effects of laser power intensity, scanning speed, and hatch spacing on the cracking density and morphology were studied. It was reported that a high laser intensity led to the solidification cracking while a low laser intensity led to the grain boundary cracking. Additionally, the solidification cracking was increased by decreasing the hatch spacing. Wang et al. 117 studied the crack growth as affected by different scanning strategies during SLM of molybdenum powders. By rotating the laser scanning direction between adjacent layers, the columnar grain structure transferred into the interlocked structure, which greatly shortened the growth path for cracks and caused the crack deviation. Chemical compositions were identified as another important factor affecting cracking during SLM of Hastelloy-X alloys. 118,119 It was found that the minor elements, such as Mn, Si, and S, could lead to a decrease of solidification temperature and a microsegregation at grain boundaries, resulting in the crack initiation. Cloots et al. 120 investigated how cracking was affected by a laser beam profile. It was reported that for SLM-built Ni-based superalloy IN738LC, the crack density was reduced from 0.27 to 0.11 mm/mm² by using a laser beam with a Doughnut profile rather than a Gaussian profile. This was
attributed to the difference in melt shape and grain structure when applying different beam profiles.

C. Approaches to eliminate or minimize balling and cracking

The balling effect and cracking behavior are significant issues restricting the applications of SLM. Some researchers claim that it is difficult to completely eliminate balling and cracking by simply tuning the laser parameters and scanning strategy. Alternative approaches to tackle the balling and cracking issues are of critical needs. To alleviate balling effects, improving the wettability of the powder material has been proved to be an effective method. For instance, Agarwala et al.121 developed a multiphase powder approach by prealloying phosphorus with a copper alloy to improve the wetting characteristics, resulting in the as-built sample with smooth surface finish, as shown in Fig. 15. To eliminate cracking during SLM, preheating substrate is commonly recommended for stainless steels and Ti alloys.122,123 However, it is ineffective for metals like aluminum alloys, due to the severe level of cracking induced by the large solidification temperature range, high thermal expansion coefficient, and large solidification shrinkage. Chemical additives serve as another possible approach for preventing cracking. As shown in Fig. 16, it was found by Zhang et al.124 that the addition of 2 wt.% Zr to Al-Cu-Mg alloy powder significantly reduced and eventually eliminated the crack formation during SLM, which was attributed to the formation of Al3Zr precipitates and grain refinement. Further systematic investigations are necessary to identify and optimize approaches for eliminating or minimizing balling and/or cracking in SLM of metallic materials.

VI. COMPUTATIONAL MODELING OF MICROSTRUCTURE EVOLUTION IN THE SLM PROCESS

Due to the rapid heating and cooling cycles during the SLM process, in situ observation of microstructure evolution during laser processing is of specific difficulties. While the parametric study by experiments is time- and cost-consuming, the computational modeling offers the exceptional capability of bridging the process–microstructure relationship and, therefore, provides guidance and insights for process design and optimization. For SLM modeling, the primary focus has been placed on the grain structure evolution. In the literature, the grain growth during SLM can be simulated using three computation approaches including Monte Carlo (MC), cellular automaton (CA), and phase field (PF) simulation.

A. Monte Carlo simulation

As one of the earliest methods for microstructural modeling, MC simulation is widely used in casting and welding processes. It utilizes the change in the system energy as a criterion to generate the probability of grain growth, i.e., \( P = \exp(-\Delta E/k_BT) \) if \( \Delta E > 0 \) or \( P = 1 \) if \( \Delta E \leq 1 \), where \( \Delta E \) is the change in total energy, \( k_B \) is Boltzmann’s constant, and \( T \) is the local temperature.125 Rodgers et al.126 developed a modified kinetic MC 3D model incorporating the molten zone movement to predict the grain evolution in multilayer SLM. The shape of molten zone and heat-affected zone, the temperature gradient, and the scanning direction were taken into account in this model. The temperature-dependent grain boundary mobility was incorporated into the probability function. It was reported that the simulated results were in good agreement with experimental findings, as illustrated in Fig. 17. MC method
provides the exceptional capability of large scale simulation with a low computational cost. However, such a probability-based method lacks physical principles and hence is unable to predict detailed microstructure characteristics such as the dendrite front.

B. Cellular automaton simulation

For CA simulation, each computational cell has a solid fraction $f_s$ representing the local state (liquid if $f_s = 0$, solid if $f_s = 1$, and interface if $0 < f_s < 1$), as illustrated in Fig. 18. The interface consists of one layer of the CA cell that separates solid and liquid. Generally, a decenter square algorithm is used for the CA model, and the microstructure evolution is controlled by the interface growth rate obtained from coupled mass or heat equations. Many researchers have adopted a CA-finite element coupled model to predict the microstructure evolution in a large region. The temperature profile induced by laser energy is usually acquired first on the macroscale. Then, on the microscale CA model, the density of nucleus and the interface growth rate (tip velocity) are calculated based on the real-time undercooling, and the nucleation sites are updated to the grids at each time step. Yin and Felicelli used the CA model to predict the effects of processing parameters (including the laser scanning speed, layer thickness, and substrate size) on the dendrite morphology during SLM of a single layer Fe-0.13 wt. % alloy. Nie et al. simulated the microstructure evolution during SLM of Nb-bearing nickel-based superalloy IN718. The coefficient relating the growth angle and the preferential crystallographic orientation was introduced to capture the growth of dendrites. Lopez-Botello et al. further applied this method to simulate a multilayer SLM of AA2024. As one of the most advanced models, CA is coupled with lattice Boltzmann (LB) method for SLM simulation, where the temperature history is predicted using the LB method with consideration on the capillarity, wetting behavior, and melt pool dynamics, as shown in Fig. 19. The CA method is exceptional due to its highly computationally efficient and capability of simulating anisotropic effects during solidification. However, it fails to provide a detailed description of dendritic morphology, because the tip curvature and...
secondary dendrite arm are greatly affected by grid meshing since the interface only consists of one layer of cells. It is also not suitable to simulate complex processes such as solidification of the multicomponent alloy due to the constraints of the growth kinetics formulation.

C. Phase field simulation

In PF simulation, an order parameter ($\varphi$) is used to represent the state of grids. For the solid phase, $\varphi = 1$, and for the liquid phase, $\varphi = 0$. For the interface, $0 < \varphi < 1$, which usually consists of a few cells and, thus, provides a smooth transition of material properties across the interface. The microstructure evolution is obtained by the solution of a set of functions including the phase field equation, solute, and/or temperature conserved equations. In the work of Cao and Choi, laser processing of H13 tool steel was simulated by coupling PF microscale model with the macroscale temperature field obtained from a volume of the fluid based algorithm. The microstructure evolution was predicted by solving the phase field equation, composition equation, and energy equation with dimensionless temperature and composition. The solidification of pure metal and binary alloy was both analyzed, and the effects of microcell location and undercooling on the microstructure evolution were studied. Keller et al. assumed a local steady-state condition during SLM by using an imposed temperature field. $T(x,t) = T_0 + G(x-V_s t)$ as an input for the phase field equation, where temperature gradient $G$ and velocity $V_s$ were considered to be constant. The microstructure evolution of Ni-Nb binary alloys was simulated by solving the phase field equation and the composition field equation. Fallah et al. used a similar method to predict the microstructure evolution by solving a phase field equation and a supersaturation field equation. The dendrite growth morphology, especially the dendrite arm spacing, during directional solidification of Ti-Nb alloys was studied. Kundin et al. successfully predicted the grain structure during SLM of IN718 alloys. The model was formulated with a general phase field equation and multiple diffusion equations for each component. The temperature-dependent parameters used in the diffusion equations including the equilibrium concentration in liquid and solid phases were recalculated in each time step. The modeling results were validated by the steady-state growth problem taking solute trapping into account. The effects of undercooling on dendritic tip radius and velocity were predicted for multicomponent alloy solidification.

Recently, Acharya et al. used a computational fluid dynamics model to describe the characteristics of molten pool and to introduce the laser heat source as affected by the scanning speed, $V_s$, in the phase field equation, composition equation, and energy equation. This model was used to predict the evolution of microstructure features during SLM of IN718 alloys, including the dendritic fragmentation, dendritic size, and dendritic orientation. As shown in Fig. 20, the dendritic growth was simulated in both longitudinal and transverse cross sections. The effect of scanning speed was analyzed via simulation results and further discussed in terms of multilayer build. The PF method does not require the tracking of interface position, and hence the number of governing equations is minimized. It is suitable to simulate complex microstructure and multicomponent alloy. However, it is restricted due to the small scale simulation with a relatively high computational cost.

VII. CURRENT CHALLENGES AND FUTURE RESEARCH DIRECTIONS

As a promising 3D printing technology, SLM has been extensively applied to different fields and experimented on various materials. However, the rapid processing makes the microstructure evolution complicated with issues such as anisotropic mechanical behaviors, balling, and/or cracking.

The anisotropic mechanical properties of SLM-built parts are mainly attributed to anisotropic grain orientations aligned with the thermal gradient directions as the solidification proceeds. The heat flow direction and temperature gradient as affected by various process parameters play major roles in determining the grain growth during the SLM process. Although the effects of laser power and scanning speed on the grain growth have been extensively studied for different materials, a few efforts have been put on understanding effects of other laser parameters (hatching spacing and layer thickness151,152 and powder characteristics (size distribution and chemical composition). Additives and/or dynamic parameters...

approaches integrated into the SLM process provide potential solutions to address the issue of anisotropic mechanical properties by generating equiaxed grains all over the melting area. Nevertheless, the uniform distribution of nanoadditives in the powder bed remains a challenge, while the integration of dynamic approaches with SLM brings difficulties in manufacturing process design. Altering the powder composition could serve as an alternative method to obtain ultrafine grains and address the anisotropic effect in SLM; however, further investigations are needed to elucidate the fundamental mechanism responsible for the powder composition effect. On the other hand, in order to tackle the balling and/or cracking challenges, in addition to design and optimization of laser parameter and scanning strategy, alternative approaches are of critical needs, such as improving the wettability of the powder materials, employing preheating substrate, and adding chemical additives. However, the lack of systematic investigations makes these approaches less effective.

Porous structures offer a variety of advantages, such as low thermal conductivity, low density, short diffusion path, and large surface area of reaction. The SLM-fabricated porous structure has shown its promising future due to the combination of advantages of laser-based rapid prototyping and unique properties of porous material toward applications in mass, momentum, and energy transport. Recently, a few investigations have been conducted with a focus on porous structure fabrication using SLM by partial melting,143–145 or adding pore-forming agent;146 however, the properties of as-built porous parts have been rarely studied. It is envisioned that future studies aiming at advancing the understanding of process–microstructure–property relationship of SLM-fabricated porous structures will open new avenues for applications of SLM technology.

For SLM of metallic materials with phase transformation and/or precipitation phenomena, it is feasible to control the phase transformation or precipitation during SLM by adjusting processing conditions, leading to tunable bulk properties such as hardness, yield strength,147 and magnetic performance.148 However, due to the rapid melting and solidification, the phase transformation and precipitation in the SLM process can be very different as compared to those in conventional manufacturing processes like casting and heat treatment. Thermal dynamics and physical metallurgy studies on phase transformation and precipitation during SLM will be the first investigation step toward a fundamental understanding of process–structure–property relationship during SLM of metallic materials, for which the phase volume fraction and/or precipitate characteristics play critical roles on determining their properties and multiphysics performance. Computational methods exhibit their advantages in process optimization and microstructure prediction. The physics-based modeling efforts nowadays, such as CA and PF, have mainly focused on the microstructure evolution during the single scan. However, the SLM process involves remelting of previous scanned tracks and solidified layers, leading to the grain reformation. Therefore, it is necessary to further develop models for multiscan/multilayer to effectively predict the entire microstructure in 3D components.149–151 Moreover, the microstructure simulation typically needs temperature history as the input. Currently, the temperature profile is often acquired from a simple heat conduction model by considering the powder bed as a bulk material. Nevertheless, the material properties such as thermal conductivity and absorptivity can change as the powder particles melt/consolidate together. In addition, when the laser beam is directed onto the powder bed, the loose powder particles would result in multiple absorptions and scatterings of the beam energy,152 rather than a simple absorption as a bulk material. It would be more accurate for predicting the microstructure evolution by improving the temperature model with consideration on these important facts. Furthermore, researchers have been dedicating to simulating the grain growth during SLM, while there is very little research that involves the simulation of porosity formation,153 phase transformation,154 or precipitation during SLM.

Considering the aforementioned current challenges in SLM process design and optimization, future investigations are needed to address the following knowledge and technology gaps:

1. The SLM process design and optimization toward 3D printing of porous structure is of great importance. 3D-printed components with controlled porosity find many applications including water filtration, thermal storage, vacuum insulation, and bioactive implant.134 However, for the SLM process, most researchers have focused on manufacturing “full dense” or “pores-free” part via adjusting process parameters or post-treatment conditions. To date, there are few works on manufacturing porous structure with SLM process,151,143,144,155 let alone the properties of the as-built porous parts. The capability and feasibility of fabricating porous components using SLM is promising and needs to be further investigated.

2. Adding chemical additives into commercial alloy systems has been proved to be an effective method for tuning the grain size, generating the strengthening second phase, and alleviating balling or cracking. In addition, the formation of new phases could affect the melt pool by altering the temperature distribution and thermal behavior, such as dynamic viscosity, capillary instability, and surface tension, leading to beneficial effects on reducing porosity and residual stress. Furthermore, the additions make SLM possible for processing some unprintable metallic materials, which is of enormous significance to expand the material library for SLM and hence enlarge the application areas. Future research with a focus on adding chemical additives into alloy systems will further advance understanding, control, and optimization of the SLM process.

3. Computational methods are powerful tools for process design since it is difficult to directly observe the microstructure evolution during the SLM process. The future effort is necessary to further develop models for multiscan/multilayer simulation to effectively predict the entire microstructure in 3D-printed components. Moreover, the prediction of temperature profile during SLM can be further improved by considering the powder size effect on the absorption and scattering of laser beam energy.

VIII. CONCLUSION

In this paper, the microstructure evolution during SLM of metallic materials is reviewed. The effects of processing conditions (laser parameters, scanning strategies, additives, and dynamic
approaches) on the grain growth during SLM are reviewed in terms of grain morphology, size, and orientation. The formation of pore structures during SLM as affected by laser parameters, scanning strategies, and powder characteristics is discussed. In addition, the phase transformation during SLM of steels and titanium alloys is reviewed. The precipitation during SLM of titanium, nickel, and Al/Mg alloys is studied. Moreover, since the balling and cracking are complicated and difficult to avoid during SLM, the formation mechanisms and classification are discussed in detail. Finally, various computational methods including MC, CA, and PF approaches to predict the grain structure evolution during the SLM process are introduced. This review summarizes the current advances in understanding of the process–microstructure relationship during SLM of metallic materials. It provides guidance for process design and optimization in SLM of metallic materials.

ACKNOWLEDGMENT

The authors appreciate financial support by start-up funding from the Department of Mechanical Engineering at the University of Nevada, Reno.

REFERENCES


D. Sun, X. Li, and W. Tan, presented in the 28th Annual International Solid Freeform Fabrication Symposium, Austin, TX, 7–9 August 2017.


