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Role of melt behavior in modifying oxidation distribution using an interface incorporated model in selective laser melting of aluminum-based material

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A transient three dimensional model for describing the molten pool dynamics and the response of oxidation film evolution in the selective laser melting of aluminum-based material is proposed. The physical difference in both sides of the scan track, powder-solid transformation and temperature dependent physical properties are taken into account. It shows that the heat energy tends to accumulate in the powder material rather than in the as-fabricated part, leading to the formation of the asymmetrical patterns of the temperature contour and the attendant larger dimensions of the molten pool in the powder phase. As a higher volumetric energy density is applied ($\geq 1300$ J/mm$^3$), a severe evaporation is produced with the upward direction of velocity vector in the irradiated powder region while a restricted operating temperature is obtained in the as-fabricated part. The velocity vector continuously changes from upward direction to the downward one as the scan speed increases from 100 mm/s to 300 mm/s, promoting the generation of the debris of the oxidation films and the resultant homogeneous distribution state in the matrix. For the applied hatch spacing of 50 $\mu$m, a restricted remelting phenomenon of the as-fabricated part is produced with the upward direction of the convection flow, significantly reducing the turbulence of the thermal-capillary convection on the breaking of the oxidation films, and therefore, the connected oxidation films through the neighboring layers are typically formed. The morphology and distribution of the oxidation are experimentally acquired, which are in a good agreement with the results predicted by simulation.

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I. INTRODUCTION

Selective laser melting (SLM), as an additive manufacturing (AM)/3D printing process capable of fabricating bespoke individual components,$^{1,2}$ has attracted a lot of attention in a large amount of remarkable structures in a fairly wide range of materials, including titanium biomedical scaffolds for hip replacements and knee arthroplasties,$^3$ a fairly wide range of materials, including titanium biomedical scaffolds for hip replacements and knee arthroplasties,$^3$ manufacturing (AM)/3D printing process capable of fabricating bespoke individual components,$^{1,2}$ has attracted a lot of attention in a large amount of remarkable structures in a fairly wide range of materials, including titanium biomedical scaffolds for hip replacements and knee arthroplasties,$^3$ designing rules of stainless steel for porous structures.$^4$ However, a number of materials, due to their intrinsic metallurgical properties, have been proved to be much more difficult to be successfully produced by the SLM process. Aluminum alloys, extensively applied in automotive and aerospace industries, are of a considerable interest in promoting the development in the SLM process.$^5$ Brandl et al. have concluded that the fatigue resistance of the SLM-processed AlSi10Mg parts were significantly sensitive to the platform heating and post heat treatment while the building direction has the negligibly effect on fatigue behavior.$^6$ The influence of the reinforcement weight fraction on thermal evolution behavior and fluid dynamics during SLM of TiC/AlSi10Mg is investigated by Gu and Yuan with the combined method of a numerical and experimental process.$^7$ Generally, SLM of aluminum materials, due to the limited absorption caused by the high reflectivity of laser energy and rapid dissipation of heat loss because of the high thermal conductivity,$^8$ is typically fabricated using high laser powers combined with low laser scanning speeds, ensuring the full melting and the resultant efficient spreading of the melt.$^9$ However, in this processing condition, it results in a disproportional increase in production costs, restricting the suitable applications in the wide and advanced industries. Nevertheless, another great challenge encountered in the SLM processing of aluminum is the rapid formation of adherent oxidation films on the free surface and/or bottom area of the molten pool.$^{10}$ Subsequently, the main obstacle of the oxidation films presented on the solid tracks and below the molten pool gives rise to a reduction in the wetting ability, and the oxidation stirred into the molten pool will therefore generate a series of weakness, such as residual unmelted powders and the resultant pores, within the as-fabricated parts.$^9$

Most researches pay high attention to the effects of oxidation films on aluminum processing related to the conventional manufacturing methods such as squeeze casting.$^{11,12}$ In comparison with the advantages of the SLM process due to the almost unlimited flexibility of complex geometry and the quick industrialization of this processing technology,$^{13–15}$ oxidation during SLM process has a detrimental influence on the wetting behavior of the molten pool with the previously solidified tracks, the surface morphology, the residual porosity, and the resultant relative density in the terminally solidified SLM-fabricated part, considerably restricting the industry performance lifetime.$^9$ Louvis has stated that the
major obstacle presented in the SLM processing aluminum is the formation of oxidation films within the molten pool, and it was experimentally concluded that the oxidation films on the top surface would be vaporized by the high energy, and the side/bottom ones would be broken by the stirring due to the Marangoni force.9 Therefore, it is necessary to have a thorough understanding of the breaking mechanisms of the oxidation films distributed in the different regions within the molten pool, promoting the feasibility of the SLM processing aluminum alloy with the combination of the excellent metallurgical behavior and the attendant high industrial performance. Consequently, numerical simulation approach, which can be applied to depict directly the thermal behavior and the mass transmission phenomenon within the molten pool, is chosen as a promising alternative to study the breaking mechanism of the oxidation films. Generally, most of the researchers have paid high attention to the velocity field within the molten pool by considering the interaction between the laser source and mere powder material in the SLM process.16–18 They numerically studied the influence of the processing parameters, e.g., the laser power and scan speed, on the melt dimensions, melting/evaporation of powder material and the resultant temperature change. It seems that the unique and tremendously different material physical properties on both sides of the molten pool irradiated by the laser spot are not taken into account in their mathematical model.

In this work, the numerical simulation taking into account the physical difference in both sides of the scan track, regarding the effect of different processing conditions on the molten pool dynamics and the response of the thermal-capillary convection in the SLM-processed AlSi10Mg powder, was presented, using commercially computational fluid dynamics (CFD) software. The fluid flow driven by the surface tension gradient influenced by the O element was considered in the physical model, and the temperature field, temperature gradient, and the thermo-capillary convection pattern in the molten pool were simulated, predicating the terminate distribution state of the oxidation films. Furthermore, in order to testify the accuracy of the developed simulation model, the oxidation films with its distribution state in SLM-processed parts predicted by the numerical simulation are compared with those acquired via experiments.

II. MODEL DESCRIPTION

A. Physical model

The schematic of the SLM process is depicted in Fig. 1(a). The laser energy used in this work is estimated as a heat flux, which is defined as a Gaussian function. The three dimensions of the numerical model are 100 μm × 140 μm × 80 μm in the X-, Y-, and Z-axis directions. During the SLM process, the interaction area irradiated by the laser beam is typically consisted of the as-fabricated part and powder feeding material. Therefore, the region connected by the two different materials is defined as an interface shared by the bulk/powder material with different physical properties (Fig. 1(a)). The heat loss from the six surfaces of the physical model is reasonably identified as the combination of the radiation and convection modes. The initial ambient temperature of the physical model is set as 300 K. The calculation was conducted with a structured mesh composed of 240 000 hexahedral cells. The convergence of the calculations is obtained when the sum of normalized residues was equal to 10⁻³. The solution convergence is obtained as the relative change of energy between successive iterations is lower than 10⁻⁶ for each node of the field.

B. Governing equations

The mathematical model, based on the Navier-Stokes equations, is employed in this work. The governing equations are composed of the conservation of the mass, momentum and energy, which are depicted in Eqs. (1)–(3)\(^{19}\)

\[
\frac{\partial \rho}{\partial t} + \nabla(\rho \vec{v}) = 0, \tag{1}
\]

\[
\frac{\partial (\rho \vec{v})}{\partial t} + \nabla(\rho \vec{v} \vec{v}) = -\nabla p + \mu \nabla^2 \vec{v} + \vec{S}, \tag{2}
\]

where \(t\) is the time, \(\rho\) is the density, \(p\) is the pressure, \(\vec{v}\) is the melt velocity vector in the X, Y, and Z directions, \(\mu\) is the viscosity, and \(\vec{S}\) is the source item.

During the SLM process, the initial powder material generally undergoes the melting/phase change/solidification process, playing a crucial role in determining the thermal behavior and the resultant temperature field. Therefore, the energy is identified, taking into account the sensible enthalpy\(^{20}\)

\[
\frac{\partial (\rho H)}{\partial t} + (\rho \vec{v} H) = \nabla(k \nabla T) + Q, \tag{3}
\]

where \(k\) is the thermal conductivity, \(T\) is the operating temperature, \(H\) is the enthalpy, and \(Q\) is the volumetric heat source. In the finite volume-based code Fluent, enthalpy of the material is computed as the sum of the sensible enthalpy and the latent heat\(^{21}\)

\[
H = h + \Delta H, \tag{4}
\]

\[
h = h_{\text{ref}} + \int_{T_{\text{ref}}}^{T} C_p dT, \tag{5}
\]

where \(C_p\) is the specific heat at constant pressure, \(h\) is sensible enthalpy, \(\Delta H\) is the latent heat, \(T_{\text{ref}}\) is the reference temperature, and \(h_{\text{ref}}\) is the reference enthalpy.

A Gaussian heat flux, as an input laser heat source, is employed in the physical model. Meanwhile, the heat flux input with the heat loss caused by the convection, radiation, and evaporation modes is expressed as\(^{20}\)

\[
-\kappa \left( \frac{\partial T}{\partial z} \right)_{\text{top}} = Q - h_c(T - T_\infty) - \sigma_c e (T^4 - T_\infty^4) - q_e, \tag{6}
\]

where \(h_c\) is the coefficient of the convective heat-transfer, \(T_\infty\) is the ambient temperature, \(e\) is the emissivity, \(q_e\) is the evaporation heat loss, and \(\sigma_c\) is the Stefan-Boltzmann constant. The heat source is identified as\(^{22}\)
where $P$ is the laser power, $d$ the laser beam penetration depth, $\omega_{f0}$ and $\omega_f$ are the beam focal radii at the surface and at depth, $A$ is the absorption of the material, $u(z) = 1$ for $0 \leq z \leq d$ and $u(z) = 0$ otherwise. In order to distinguish the different interaction behaviors of the melt and the powder material, the optical extinction coefficient is reasonably considered in the powder phase within the physical model, which is dependent on the material absorption and the particle packing.

Generally, the driving forces of the melt convection in the molten pool irradiated by the laser beam are the combination of surface tension, viscous and buoyancy forces. On the free surface of the molten pool, the shear stress caused by the temperature gradient is given by

$$\tau = \frac{\partial \sigma_T}{\partial f} \nabla_s T,$$

where $\frac{\partial \sigma_T}{\partial f}$ is the surface tension gradient and $\nabla_s T$ is the temperature gradient.

Generally, the powder material has a limited thermal conductivity and enhanced energy absorption compared to the bulk one. Therefore, the effective thermal conductivity of the powder bed is defined as

$$\frac{\kappa}{\kappa_f} = \left(1 - \sqrt{x}\right) \left(1 + \frac{(1-x)\kappa_r}{\kappa_f}\right) + \sqrt{x} \left[\frac{2}{1 - \frac{\kappa_r}{\kappa_s}} \left(\frac{1}{1 - \frac{\kappa_r}{\kappa_s}} \ln \left(\frac{\kappa_s}{\kappa_f}\right) - 1\right) + \frac{\kappa_r}{\kappa_f}\right].$$

FIG. 1. Interaction region in the physical model consisting the powder material and the bulk part (a) and the temperature distribution obtained in the simulation process (b).

### C. Numerical simulation

The simulation is carried out using the FLUENT commercial finite volume method (FVM) package (version 6.3.26) to simulate the thermal behavior, velocity field on both sides of the center along the scan direction and the resultant distribution of the oxidation films in the terminally solidified part. The temperature field obtained in the simulation process is shown in Fig. 1(b). The thermal physical properties of AlSi10Mg material are depicted in Fig. 2(a).

The SLM processing parameters depending on our previous research are shown: laser power ($P$) and the thickness of the powder layer ($l$) at 200 W and 30 μm, respectively. In order to change the processing conditions during SLM process, various scan speed ($v$) of $100 \text{ mm/s}$, $200 \text{ mm/s}$ and $300 \text{ mm/s}$ and scan line hatch spacing ($h$) of $40 \mu\text{m}$, $45 \mu\text{m}$, and $50 \mu\text{m}$ were identified by the system control program. Therefore, five different “volumetric laser energy densities” ($\eta$) of $1600 \text{ J/mm}^3$, $1480 \text{ J/mm}^3$, $1300 \text{ J/mm}^3$, $830 \text{ J/mm}^3$, and $555 \text{ J/mm}^3$, which were defined by

$$\eta = \frac{P}{lhv},$$

were used to estimate the laser energy input to the powder layer being processed.

### III. EXPERIMENTAL PROCEDURE

#### A. Powder materials and processing and characterization

The raw powder materials applied in this study is the 99.7% purity AlSi10Mg powder with a spherical shape and an average particle diameter of 30 μm (Fig. 2(b)). In order to have a thorough understanding of the thermo-capillary convection on the oxidation films evolution, the Al2O3 powder material is added 20 wt. %. The SLM apparatus consisted mainly of an YLR-500-WC ytterbium fiber laser with a power of $\sim500 \text{ W}$ and a spot size of $\sim70 \mu\text{m}$ (IPG Laser GmbH, Germany), an automatic powder spreading device, an inert argon gas protection system, and a computer system for process control. Details of the SLM process have been thoroughly summarized in the previous research paper. The rectangular specimens with dimensions of $10 \text{ mm} \times 10 \text{ mm} \times 5 \text{ mm}$ were built in a layer-by-layer manner until completion. Specimens

$$Q = \frac{AP}{\pi \omega_{f0} d} \left(\frac{\omega_f}{\omega_{f0}}\right)^2 \exp \left(-\frac{2(x^2 + y^2)}{\omega_f^2}\right) u(z),$$

(7)

where $\kappa_r = 4B\sigma_a T_p^3 D_p$, (10)

where $D_p$ and $T_p$ are the representation of the average diameter and operating temperature of the powder particles, respectively. $B$ is a view factor, approximately identified as 1/3.
for metallographic examinations were prepared according to the standard procedures and then etched with a solution composing HF (2 ml), HCl (3 ml), HNO₃ (5 ml), and distilled water (190 ml) for 10 s. The typical top surface morphology and microstructure study of the SLM-processed parts was performed using a S-4800 field emission SEM (FE-SEM) (Hitachi, Japan) at 5 kV.

IV. RESULTS AND DISCUSSION

A. Temperature distribution

Figure 3 shows the typical temperature distribution of the cross-sections within the molten pool using different processing parameters. It shows that the temperature contours are not symmetrical along the center of the scan direction, and the operating temperature obtained in the powder material is significantly higher compared to that produced in the as-fabricated part due to the tremendous difference in the thermo-physical properties, e.g., the absorption of the laser energy, the thermal conductivity, and the resultant heat loss rate. The heat energy has a tendency to accumulate in the powder material caused by its combination of high laser absorption and limited heat conductivity, leading to the formation of a higher operating temperature and the subsequent melting of the powder material. However, the bulk part, with an efficient thermal-conductivity (≈230 J/(mK)), considerably promotes the efficient loss of heat from the irradiated region to the ambient, resulting in a lower operating temperature obtained in the molten pool. As the application of scan speed \( v \) increases from 100 mm/s (\( \eta = 1600 \text{ J/mm}^3 \)) to 300 mm/s (\( \eta = 555 \text{ J/mm}^3 \)), the operating temperature obtained in the molten pool decreases from 1250 K (Fig. 3(a)) to 1050 K (Fig. 3(d)). The operating temperature obtained in the interaction region irradiated by the laser beam will be gradually reduced as the scan speed increases, playing a crucial role in the melt volume of molten pool and the resultant spreading behavior of the melt. On the other hand, as the hatch spacing increases to 45 \( \mu \text{m} \) (\( P = 200 \text{ W}, v = 100 \text{ mm/s}, \) and \( l = 30 \mu\text{m}, \eta = 1480 \text{ J/mm}^3 \)), more initial powder material would be therefore irradiated by the laser.
beam while the remelting rate of the as-fabricated part is generally reduced. As a result, the energy heat has a tendency to accumulate in the powder material with an elevated temperature value of 1300 K, and the operating temperature produced in the as-fabricated part is as low as the melting point, caused by the sufficient loss of heat through the thermal conduction method (Fig. 3(c)). As the hatch spacing further increases to 50 μm (P = 200 W, v = 100 mm/s and l = 30 μm, η = 1300 J/mm³), the operating temperature obtained in the powder material significantly increases to 1400 K while the remelting phenomenon in the as-fabricated part is seriously restricted derived from the limited operating temperature of 800 K (Fig. 3(e)), probably resulting in the production of poor metallurgical bonding ability between the neighboring tracks. It can be concluded that as the inappropriate hatch spacing is applied, the powder material is overheated while the bulk part is oppositely under-heated and it probably gives rise to the evaporation phenomenon and partial/non-melting of the as-fabricated part, which is detrimental to the successful SLM process.

B. Molten pool dimensions

In order to have a detailed understanding of the molten pool evolution and the metallurgical bonding ability between neighboring tracks and layers, the width and depth of the molten pool using different processing parameters are displayed in Fig. 4. It is apparent that the width and depth within the molten pool obtained in the powder side are larger than those produced in the bulk material using the same processing parameter. It can be analyzed that the energy from the laser spot is easily accumulated in the irradiated powder region due to the limited thermal conductivity of powder, and as a result, it generates larger dimensions of the molten pool, promoting the spreading and the attendant bonding behavior to the neighboring as-fabricated tracks/layers. As the scan speed is 100 mm/s (η = 1600 J/mm³), the width and depth obtained within the molten pool in the powder layer are 71 μm and 50 μm, respectively (Fig. 4), and those produced in the as-fabricated part are 45 μm and 42 μm, respectively. As the scan speed increases to 200 mm/s (η = 830 J/mm³), the interaction time in the irradiated region is generally reduced, giving rise to a slight decrease in the width (62 μm in powder and 39 μm in as-fabricated part) and depth (45 μm in powder and 37 μm in as-fabricated part), respectively. As the scan speed is 300 mm/s (η = 555 J/mm³), the width continues to decrease to as low as 51 μm in powder and 30 μm in as-fabricated part, and the depth of the molten pool falls to 40 μm in the powder and 35 μm in the as-fabricated part, respectively. It can be concluded that the dimensions of the molten pool have a negative relationship with the scan speed, caused by the restricted interaction time and the limited laser energy and material couple rate. As the hatch spacing increases to 45 μm (η = 1480 J/mm³), the width and depth within the molten pool in the as-fabricated part are sharply reduced to 26 μm and 20 μm, and these obtained in the powder material slightly decreased to 50 μm and 48 μm, respectively. The remelting depth in the as-fabricated part is accordingly reduced due to the combined influence of the enhancement of the hatch spacing and the lower operating temperature (Fig. 3(c)), resulting in the production of lower melting depth than the thickness of the powder bed. As the hatch spacing further increases to 50 μm (η = 1300 J/mm³), the width of the molten pool in the powder material radicals decreases to 35 μm in combination of a slight decrease in the as-fabricated part, as low as 20 μm (Fig. 4). The depth obtained within the molten pool slightly decreases to 46 μm in the powder material and 18 μm in the as-fabricated part. As more powder material is irradiated by the laser beam (h ≥ 45 μm), the irradiated powder region, due to the limited thermal conductivity, will be therefore seriously overheated and thus the melt has a high tendency to evaporate, resulting in the severe heat loss and the subsequent shrinkage of the dimensions of the molten pool in the powder. Meanwhile, the laser irradiated region in the as-fabricated part using larger hatch spacing is tremendously decreased, producing a low operating temperature (Fig. 3(e)) and the attendant reduction in the dimensions of the molten pool (Fig. 4). As a result, a poor metallurgical bonding ability between the neighboring tracks will be produced, caused by the combined effect of the evaporation phenomenon of the melt in the powder and the limited remelting behavior in the as-fabricated part.

C. Velocity field evolution

Figure 5 depicts the velocity field within the cross-sections of the molten pool using different processing parameters. It is obvious that the velocity vector in the top surface of the molten pool is typically shown in the inward pattern, which is driven by the surface tension derived from the temperature gradient. Meanwhile, it is worth noting that the high velocity vector of the melt is typically located near the interface in the powder material side, and the velocity vectors within the molten pool in the powder material and as-fabricated part are shown in the upward and downward direction pattern, respectively (Fig. 5). In order to have detailed information of the molten pool, the temperature gradient and the velocity distribution obtained along Y-axis direction within the molten pool are displayed in Fig. 6. It can be seen that the operating temperature obtained on both sides of the interface is of a significant difference, and the
A steeper temperature gradient is generally located in the powder material due to the restricted thermal conductivity and the resultant insufficient heat diffusion, reasonably leading to the formation of violent thermo-capillary convection in the powder material side (Figs. 5 and 6(b)). Meanwhile, it is obvious that the velocity obtained in the powder material side is typically higher compared to that produced in the as-fabricated part (Fig. 6(b)). As the scan speed is 100 mm/s ($v = 100 \text{ mm/s}$, $h = 40 \mu\text{m}$, $\eta = 1600 \text{ J/mm}^3$), the upward direction of the melt in the region, $Y \leq -20 \mu\text{m}$, is apparently produced with the average melt velocity of 5 m/s (Figs. 5(a) and 6(b)), implying the formation of the evaporation phenomenon caused by the overheating of the melt irradiated by the laser beam. As the scan speed increases to 200 mm/s ($v = 200 \text{ mm/s}$, $h = 40 \mu\text{m}$, $\eta = 830 \text{ J/mm}^3$), the evaporation region is slightly decreased ($Y \leq -30 \mu\text{m}$) (Fig. 5(b)), and the mean velocity obtained in the molten pool reaches 4 m/s (Fig. 6(b)). As the scan speed further increases to 300 mm/s, the velocity vector in the free surface of the molten pool is mainly downward (Fig. 5(d)) with the mean velocity of 3.5 m/s (Fig. 6(b)), indicating that the evaporation of the melt is fundamentally restrained and promoting the breaking of the oxidation appeared in the top surface. Therefore, it can be concluded that the velocity vector continuously changes from the upward direction to the downward one as the scan speed increases from 100 mm/s to 300 mm/s (Figs. 5(a), 5(b) and 5(d)), and as a result, the oxidation films are easily destroyed by the appearance of downward convection pattern. As the hatch spacing increases to 45 $\mu\text{m}$ ($\eta = 1480 \text{ J/mm}^3$), it is found that the area of the velocity vector upward direction is apparently expanded and shifted to the interface in the powder material with the maximum velocity of $\sim 14$ m/s, while the region of the convection flow produced in the as-fabricated part has a considerable shrinkage ($Y \leq 25 \mu\text{m}$ and $Z \geq 10 \mu\text{m}$) with the average velocity of $\sim 2$ m/s (Figs. 5(c) and 6(b)). It seems that a more serious melt evaporation phenomenon due to the high
instability of the melt is occurred in the current processing powder track, while the thermo-capillary convection is relatively limited in the as-fabricated track caused by the reduced laser energy input. As the hatch spacing is 50 μm (\( \eta = 1300 \text{ J/mm}^3 \)), the upward direction of the convection flow is produced within the severely narrow molten pool in the powder material with the maximum velocity of \(~15 \text{ m/s} \) (Figs. 5(e) and 6(b)), implying that a severe evaporation of the melt and the subsequent lack of efficient spreading of the melt will be produced. Meanwhile, the dimensions of the molten pool produced in the as-fabricated part are seriously decreased (Fig. 4) with a mean velocity of \(~2 \text{ m/s} \) in \( Y \leq 20 \mu \text{m} \) and \( Z \geq 15 \mu \text{m} \) (Fig. 6(b)), resulting in the formation of the inefficient melting of the previously fabricated track and the attendant poor metallurgical bonding behavior.\(^{15}\)

Therefore, the effect of the disturbance of the thermo-capillary convection on the oxidation films in the as-fabricated part is considerably restricted.

D. Oxide films distribution and the densification behavior

Figure 7 shows the schematic of the variation of the oxidation films morphology and distribution state using different processing parameters. It has been experimentally confirmed that the oxidation films are generally distributed in the edge of the molten/solid material (Fig. 7(a)), and it is theoretically found that the oxidation produced in the top surface of the molten pool tends to be vaporized as the powder material is irradiated by the sufficient energy input (Figs. 5(a)–5(c) and 5(e)).\(^{9}\) The oxidation located in the sides of the scan track is maintained, and thus, the unmelted powder particles cladding by the oxidation, due to the high melting temperature, are potentially remained in the interconnecting interface between powder and as-fabricated part (Fig. 7(a)), resulting in the production of weakness and porosity in the SLM-processed part. As the scan speed increases to 300 mm/s, the oxidation formed in the solid/melt is typically dispersed in the homogeneous distribution state (Fig. 7(b)). That is because that the turbulent melt stirring with a downward velocity vector has a capability to break the oxidation at the top surface into tiny pieces (Figs. 5(d) and 7(b)). Therefore, the oxidation debris, driven by the thermo-capillary forces, is fully rearranged with the formation of a uniform distribution state as dispersion strengthening reinforcement in the solidified part (Fig. 7(c)). Meanwhile, the laser energy in the upper region of the molten pool is able to diffuse along the Z direction to the bottom edge, leading to the successful remelting of the beneath part in the previously solidified layer. As a result, a fine metallurgical bonding ability is reasonably obtained. However, as a large hatch spacing of \(~50 \mu \text{m} \) is applied, the velocity vector obtained in the powder material is produced in the upward pattern (Figs. 5(c) and 5(e)), leaving the oxidation remained in the edge of the molten pool, and the operating temperature reaches as low as melting point (Figs. 3(c) and 3(e)). Therefore, the as-fabricated track/layer is partially melting and the oxidation in the edge remains intact, resulting in the formation of connected oxidation films through the neighboring layers along the side of the scan tracks (Fig. 7(d)), which has a detrimental effect on the finally solidified part.

E. Experimental verification

Optical microscopy (OM) of cross-section microstructures in the SLM-processed parts using different processing parameters is depicted in Fig. 8. In order to have a further understanding of the distribution and morphology of the oxide films in the solidified parts, the high magnification OM of cross-sections in the SLM-processed part using different processing parameters was shown in Fig. 9. Upon etching, the layerwise microstructure patterns were significantly apparent, due to the layer-by-layer additive manufacturing nature of the SLM process. Nevertheless, it was obvious that the microstructures on the cross-sections, e.g., the configuration of the solidified molten pool, oxide films distribution,
residual pores, and the obtained relative density, were sensitive to the applied volumetric laser energy densities. As the scan speed $v$ was 100 mm/s ($\eta = 1600$ J/mm$^3$), the cross-section of the as-fabricated part showed a seriously heterogeneous layerwise microstructure. It was obvious that a significant amount of the oxide films was apparently incorporated into the regions between the neighboring scan tracks, and the obtained oxide films were typically produced and interconnected with a continuous channels, resulting in the formation of large inter-track pores on a scale of 50 $\mu$m and the attendant limited relative density in the terminally solidified part (Fig. 8(a)). Meanwhile, it was shown that the oxide films on the top surface were broken into thin fragments while the bottom ones were kept in an entirety, significantly restricting the efficient melt/metal bonding ability (Fig. 9(a)). At the scan speed $v$ of 200 mm/s ($\eta = 830$ J/mm$^3$), an enhanced bonding ability of the inter-tracks, due to the presence of the tenuous oxide films free of severe aggregation, was consequently produced (Fig. 8(b)).

The oxide films in the bottom region were broken into tiny pieces, enhancing the reasonable inter-layer metal/metal bonding (Fig. 9(b)). As the scan speed $v$ further increased to 300 mm/s ($\eta = 555$ J/mm$^3$), it was interesting to find that a uniform distribution of oxide fragmentation rather than segregations between the adjacent scan tracks was successfully produced, leading to the formation of a homogeneous microstructure with a distinct additive material nature combined with a nearly full dense part (Fig. 8(d)). A large number of homogeneous oxidation pieces free of aggregation in the top/edge area within the solidified molten pool were thereafter obtained, realizing a coherent and reliable metallurgical bonding behavior between the neighboring scan tracks and layers (Fig. 9(d)). Therefore, it could be concluded that the oxide films played a crucial role in the spreading and penetrating of the molten pool, and the distribution of the oxide films and the resultant relative density in the solidified part were highly dependent on the scan speed. On the other hand, as the hatch spacing increased to 45 $\mu$m ($\eta = 1480$ J/mm$^3$), the obtained layer microstructure showed that a slight aggregation of the oxide films with a connecting texture was typically dispersed in the side regions of the solidified melt (Fig. 8(c)). Meanwhile, the residual pores had a tendency to gather together due to the connection of the inter-track gaps and the oxide films, considerably restricting the enhancement of the relative density due to the poor wetting behavior between the oxide films and aluminum melt (Fig. 9(c)). At the hatch spacing of 50 $\mu$m ($\eta = 1300$ J/mm$^3$), it was obvious that the obtained layer microstructure became significantly disorderied with the appearance of the horizontal oxide films on the top surface of the solidified melt (Fig. 8(e)). The overlap gaps between
the neighboring scan tracks were orientedly elongated in the building direction, generating the perforative residual pores (Fig. 9(e)). Generally, as an increase in the applied scan spacing was used, the laser energy absorbed by the previously fabricated track was reasonably reduced, and as a result, it limited the occurrence of the remelting or evaporation of the oxide films on the top surface of the previously processed materials. Therefore, it could be seen that the orientation of the oxide films was typically in the horizontal distribution.

The typical surface morphologies of the SLM-processed part influenced by the oxidation films using different processing parameters are displayed in Fig. 10. As the scan speed \( v \) used was 100 mm/s (\( \eta = 1600 \text{ J/mm}^3 \)), the laser scan tracks were severely disordered and the aluminum drops were produced by the formation of a series of humps on the top surface (Fig. 10(a)), and thus, a poor surface quality with a high roughness was produced which needed the post processing to moderate the industrial application. At the scan speed \( v \) of 200 mm/s (\( \eta = 830 \text{ J/mm}^3 \)), the formation of humps was suppressed while the top surface was yet produced with the appearance of scattered debris (Fig. 10(b)). As the scan speed \( v \) further increased to 300 mm/s (\( \eta = 555 \text{ J/mm}^3 \)), a flat top surface combined with continuous scan tracks and sound inter-track bonding response was generally realized, leading to the formation of a pretty relative density in the as-fabricated part (Fig. 10(d)). It seemed that the surface quality prepared in the SLM-processed part was highly sensitive to the scan speed due to the appearance of various morphologies and distributions of the oxidation films (Figs. 8(a), 8(b), 9(a), and 9(b)). The oxidation films distributed in the adjacent scan tracks had a detrimental impact on the surface morphology (Figs. 9(a) and 10(a)) while the broken and homogeneously dispersed oxidation was beneficial to the fabrication of fine surface quality (Figs. 9(d) and 10(d)). As the hatch spacing increased to 45 \( \mu \text{m} \) (\( \eta = 1480 \text{ J/mm}^3 \)), the top surface appeared loosely bonded with the formation of several apparent humps between which the thin hatch gaps were slightly produced in the as-fabricated parts (Fig. 10(c)), lowering the surface quality produced in the SLM processed part. As the hatch spacing further increased to 50 \( \mu \text{m} \) (\( \eta = 1300 \text{ J/mm}^3 \)), the continuous scan tracks were typically obtained, showing an apparent enhancement in the bonding situation. However, a seriously porous structure with a large amount of long and interconnected voids in the neighboring scan tracks significantly reduces the surface quality and the attendant relative density (Fig. 10(e)). These melting defects with the formation of deep valleys combined with the residual powder particles in the adjacent hatch gaps were found as the hatch spacing increased (\( h \geq 50 \mu \text{m} \)). It seemed that the limited hatch bonding ability and the response of poor
surface quality, such as the residual pores and deep valleys, had a tendency to provide a potential capacity channel for oxidation diffusion along the deep valleys (Figs. 7(d), 8(e), and 9(e)). As a result, it could be concluded that the oxidation films were more easily broken by the thermal capillary convection caused by the application of high scan speed ($v = 300$ mm/s), leading to the production of the dense part (Figs. 8(d) and 9(d)) and the fine surface quality (Fig. 10(d)). However, the oxidation films were typically dispersed in the adjacent scan tracks and diffusion along the deep valleys as an unreasonable hatch spacing was employed ($h = 50$ μm), resulting in the formation of porous microstructures (Figs. 8(e) and 9(e)) and poor surface morphology (Fig. 10(e)).

V. CONCLUSIONS

The simulation of the temperature field, temperature gradient, and velocity field in the molten pool, predicting the morphology and distribution of the oxidation films, during selective laser melting (SLM) AlSi10Mg material has been conducted, using a finite volume method (FVM), and the following conclusions can be drawn.

1. The operating temperature obtained in the molten pool is asymmetrical along the center of the scan direction, and the temperature gradient produced in the powder material is typically higher compared to that in the as-fabricated part. The significant difference in the operating temperature obtained in both sides of the interface irradiated by the laser beam is typically caused by the unique difference in the thermo-physical properties.

2. The width and depth within the molten pool obtained in the powder side are larger than these produced in the bulk material using the same processing parameter. It can be analyzed that the energy from the laser spot is easily accumulated in the irradiated powder region due to the limited thermal conductivity of powder, and as a result, it generates larger dimensions of the irradiated powder region compared to the neighboring as-fabricated tracks/layers.

3. The velocity vector in the top surface of the molten pool is typically shown in the inward pattern, which is driven by the surface tension derived from the temperature gradient. Meanwhile, the velocity vectors within the molten pool in the powder material and as-fabricated part are
shown in the upward and downward direction pattern, respectively. As the scan speed is low or the hatch spacing is large, the evaporation phenomenon tends to be produced in the powder material, restricting the efficient spreading of the melt. Meanwhile, the upward velocity vector cannot potentially break the oxidation appeared in the melt/solid interface.

(4) The microstructures on the cross-sections, e.g., the configuration of the solidified molten pool, oxide films distribution and residual pores were sensitive to the applied volumetric laser energy densities. At the scan speed of 300 mm/s \((\eta = 555 \text{ J/mm}^2)\), a uniform distribution of oxide fragmentation rather than segregations between the adjacent scan tracks is successfully produced, leading to the formation of a homogeneous microstructure with a distinct additive material nature combined with a nearly full dense part.

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