Particulate migration behavior and its mechanism during selective laser melting of TiC reinforced Al matrix nanocomposites

Pengpeng Yuan, Dongdong Gu*, Donghua Dai

College of Materials Science and Technology, Nanjing University of Aeronautics and Astronautics, Yudao Street 29, Nanjing 210016, PR China
Institute of Additive Manufacturing (3D Printing), Nanjing University of Aeronautics and Astronautics, Yudao Street 29, Nanjing 210016, PR China

Abstract
A transient three-dimensional model for describing fluid flow characteristics and particles migration behavior within the melt pool during selective laser melting (SLM) of TiC/AlSi10Mg nanocomposites was developed. The powder-solid transition, variation of thermophysical properties, and surface tension were considered in the model. The influence of laser energy per unit length (LEPUL) on heat and mass transfer, melt pool dynamics, and particles rearrangement was investigated. It showed that the Marangoni convection became more vigorous with an increase of LEPUL, accordingly enhancing the thermal capillary force. The high laser energy input induced a sufficient liquid formation and an improved wettability, lowering the friction force exerting on TiC solids. Under this condition, the reinforcing particles can be well mixed within the matrix. The experimental study on the distribution state of TiC reinforcement in the SLM-processed Al matrix was performed. The results validated that the dispersion of TiC reinforcement changed from a severe aggregation to a uniform dispersion in the matrix with the increase of LEPUL. The TiC reinforcement experienced a microstructural variation from the standard nanoscale structure with a mean particle size of 70–90 nm to the relatively coarsened submicron morphology with an average particle size of 134 nm as LEPUL increased.

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1. Introduction
Particulate reinforced metal matrix composites (MMCs) exhibit a favorable combination of metallic matrices and stiffer and stronger reinforcements [1]. Therefore, MMCs have been extensively studied in last two decades and are significant for numerous applications in the automobile, aerospace and military industries. Nano-sized ceramic particles reinforced aluminum matrix composites (AMCs) are characterized by their superior properties, including light weight, low thermal expansion coefficient, high specific strength and stiffness, and outstanding tribological properties [2]. Due to the superior properties, e.g., high elastic modulus, high hardness, low heat conductivity, and relatively high temperature stability, TiC has become an excellent reinforcement candidate in AMCs. Besides, the TiC ceramic particles exhibit good wettability and thermodynamic stability within the molten aluminum and are accordingly chosen as the reinforcing phases in AMCs [3]. Therefore, development of nano-scaled TiC particulate reinforced Al matrix nanocomposites is of great significance. However, processing problems such as gas entrapment and interfacial microcracks, especially aggregation of reinforcement, are very difficult to be solved through conventional methods. Meanwhile, some complex shaped Al matrix nanocomposites components are always difficult to be manufactured, due to the limitation of currently available moulds and tools.

In order to overcome these shortcomings, a newly developed Additive Manufacturing (AM) technique, i.e., Selective Laser Melting (SLM), has been introduced to produce TiC/AlSi10Mg nanocomposite parts [4–7]. Recent research efforts have demonstrated that SLM, due to its flexibility in feedstock and shapes, exhibits a promising potential for net-shape fabrication of three-dimensional (3D) nanocomposite parts with any complex configurations directly from nanopowders. However, due to the vigorous Marangoni convection induced by high thermal capillary forces [8], SLM suffers from the instability of the molten pool. The liquid material flow is mainly driven by gravity force, buoyancy force and surface tension in the molten region. Therefore, the rearrangement of solid particles and resultant dispersion states of the reinforcing particulates are significantly affected by the fluid flow. Nevertheless, it is difficult to understand the underlying mechanism of the movement of reinforcing particles through...
2. Numerical approach

2.1. Physical description of SLM process

SLM of the metallic powder generally involves a complex non-equilibrium physical and chemical metallurgical process, which exhibits multiple modes of heat, mass and momentum transfer, as well as the fluid flow. A schematic of SLM physical model considering some important physical phenomena, including melting and solidification, phase changes, and interactions between laser beam and powder particles, is shown in Fig. 1.

2.2. Model establishment

A three dimensional FVM model was established using the software FLUENT to simulate the heat transfer and fluid flow during the fabrication of TiC/AlSi10Mg parts by SLM. The three-dimensions of the numerical model of powder bed are chosen to be $3 \times 1 \times 0.08 \text{ mm}^3$ and the basic grid system used in the calculation contains 30,000 hexahedral cells.

2.3. Boundary conditions

During the SLM process, the Gaussian beam strikes the top surface of powder layer and moves through a predefined scan path with a certain velocity. As the top surface of powder layer is subjected to the incident laser source, the boundary condition can be defined as [12]:

$$-\kappa_{eff} \left( \frac{\partial T}{\partial z} \right)_{z=0} = \frac{2AP}{\pi\rho\alpha T} \exp \left( -\frac{2r^2}{\alpha \tau} \right) - h_e(T - T_0) - \sigma_e(T^4 - T_0^4)$$

(1)
Laser power, \( P \)

\[ P \text{ (W)} \]

Refers to the thermal conductivity due to the laser radiation, \( j \).

\( j \)

Marangoni coefficient, \( \eta \)

\[ \eta = -3 \times 10^{-6} \text{N/(m K)} \]

Stefan–Boltzmann constant, \( \sigma \)

\[ \sigma = 5.67 \times 10^{-8} \text{W/(m}^2 \text{K}^4) \]

Ambient temperature, \( T_0 \)

300 K

Powder layer thickness, \( \ell_p \)

50 \( \mu \)m

Hatching space, \( s \)

50 \( \mu \)m

Radius of the laser beam, \( \ell_o \)

35 \( \mu \)m

Laser power, \( P \)

100 W

Scan speed, \( V \)

100, 150, 200, 400 mm/s

The SLM process generally involves a direct interaction of powders with the laser beam. Dissimilar as dense materials, the absorption of powders to laser radiation has a great relation to the initial porosity of the powder system. Due to the change in thermal physical properties of powder, particle rearrangement, phase transformations and oxidation of the melt, the manufacturing process is considered to be time and process-dependent. In this study, the laser absorption of powder materials is determined by the volume fraction of each phase, which can be expressed by:

\[ A = \sum \chi_i A_i \]  

(2)

It is well known that the motion of the fluid generally follows three basic physical conservation laws, namely the conservation of mass, momentum and energy [12]. The liquid material flow is mainly dominated by the Marangoni convection, which is induced by the temperature difference and the attendant surface tension gradient. The surface tension (\( \gamma \), N/m) is considered in the simulation model and can be expressed by [14]:

\[ -\mu \frac{\partial u}{\partial z} = \frac{\partial \gamma}{\partial y} \frac{\partial T}{\partial x}; \quad -\mu \frac{\partial v}{\partial z} = \frac{\partial \gamma}{\partial y} \frac{\partial T}{\partial y} \]  

(3)

The heat transfer mechanisms of the four lateral surfaces of the powder bed include laser radiation, heat conduction and convection. The bottom of the powder layer is connected to the metal substrate. Therefore, the bottom is assumed to be conductive.

2.4. Properties of as-used materials and SLM processing parameters

During the SLM process, the laser power density is high enough to induce a powder-solid transition, causing a nonlinear variation of the thermal physical parameters of the as-used materials, especially the thermal conductivity and specific heat capacity (Fig. 2).

Rombouts et al. [16] have compared the experimental data with the model of discrete thermal resistances. It is found that the effective thermal conductivity of the powder system considerably depends on the laser wavelength, powder particle size and morphology, as well as the packing structure and density. The effective conductivity is mainly determined by the relative density of the powder bed and can be defined as [17]:

\[ \frac{K_{eff}}{K_g} = \left( 1 - \sqrt{1 - \Phi} \right) \left( 1 + \frac{\Phi \kappa_r}{K_g} \right) + \sqrt{1 - \Phi} \left( 2 \frac{1}{1 - \kappa_r} \ln \frac{\kappa_r}{\kappa_1} - 1 \right) + \frac{\kappa_r}{K_g} \]  

(4)

\( K_r \) refers to the thermal conductivity due to the laser radiation among particles and can be further expressed as:

\[ K_r = 4 \pi \sigma \varepsilon_0 T_p^4 D_p \]  

(5)

The properties of the as-used materials and SLM processing parameters are listed in Table 1. Based on a series of preliminary experiments, the laser power \( P \) was optimized at 100 W and the laser scan speeds \( V \) were set at 100, 150, 200 and 500 mm/s by the SLM control program, in order to vary the laser processing conditions during the experiments. Four different “laser energy input per unit length” (LEPUL) of 100, 667, 500 and 250 J/m, which were defined by [18]:

\[ \text{LEPUL} = \frac{\text{Laser power, } P}{\text{Scan speed, } V} \]  

(6)

were introduced to assess the laser energy input to the powder layer.

3. Experimental details

3.1. Materials

The 99.0% purity TiC nanopowder with a near spherical shape and an average equivalent spherical diameter of 50 nm and the 99.7% purity AlSi10Mg powder with a spherical shape and a mean particle size of 30 \( \mu \)m were used in this experiment. The two components were mixed according to TiC:AlSi10Mg weight ratio of 95:5 (i.e., the equivalent volume fraction of TiC of 5.8 vol.% and AlSi10Mg of 94.2 vol.%) in a Pulverisette 4 vario-planetary mill (Fritsch GmbH, Germany) at a rotation speed of main disk of 200 rpm for 4 h, with a ball-to-powder weight ratio of 1:1.
3.2. Processing and characterization

The SLM apparatus mainly consisted of a YLR-200-SM ytterbium fiber laser with a maximum output power of ~200 W and a spot size of 70 μm (IPG Laser GmbH, Germany), an automatic powder spreading device, an inert argon gas protection system, and a computer system for the process control. Details of SLM processing procedures have been addressed in [19]. Samples for metallographic examinations were cut, ground, and polished according to standard procedures. A solution consisting of HF (2 ml), HCl (3 ml), HNO₃ (5 ml) and distilled water (190 ml) was taken as an etching agent, with an etching time of 10 s. High-resolution microstructural study of the dispersions of TiC reinforcing particulates in the AlSi10Mg matrix was performed using a Hitachi S-4800 field emission scanning electron microscope (FE-SEM).

4. Results and discussion

4.1. Temperature and velocity fields

Fig. 3 represents the calculated temperature and velocity fields within the melt pool at different SLM processing conditions.

According to the temperature contour plots, it was clear that the temperature attained its maximum value at the center underneath the laser beam and decreased radially outward. Meanwhile, the temperature gradients were observed to be varied with the laser processing parameters. The recirculating flow of the molten materials is vividly presented on the right side of each picture in Fig. 3. It was interesting to note that the fluid was going radially outward on the surface and turned down when the flow approached the edge of the melt pool. After the flow moved to the center and came up to the surface again, a recirculation pattern was finally completed. Furthermore, by comparing the size of the velocity vectors and the contour legend, the magnitude of the recirculating flow was found to increase with the increase of LEPUL.

For a given laser power, a decrease in scan speed means more laser energy input into the powder bed, thereby a higher operating temperature can be obtained. Moreover, the laser power with a Gaussian density distribution is responsible for the appearance of large thermal gradients in the melt pool. The variation of surface temperature can induce a surface tension gradient, which is the main driving force pulling the molten materials from the center to the pool periphery. At the edge of the melt pool, the molten materials sink down due to the gravity force and the fluid at the bottom floats up to the surface again because of the buoyancy force.

Fig. 3. Simulated temperature and velocity fields within the melt pool in the cross-sectional view at different laser energy per unit length (LEPUL): (a) V = 400 mm/s, LEPUL = 250 J/m; (b) V = 200 mm/s, LEPUL = 500 J/m; (c) V = 150 mm/s, LEPUL = 667 J/m; (d) V = 100 mm/s, LEPUL = 1000 J/m. Fixed processing parameters are laser power 100 W, radius of the laser beam 35 μm, hatching space 50 μm and layer thickness 50 μm.
caused by density difference. Consequently, the outward convection of the recirculating flow within the molten pool is generated, which can also be called Marangoni convection or thermal capillary convection [8].

In order to visualize the whole flow field, more detailed information about the temperature and velocity profiles is depicted in Fig. 4. It was observed that the direction of the convection pattern was the outward convection. Because the surface tension usually decreases with increasing the temperature, the cooler liquid materials near the pool edge are inclined to draw the melt away from the pool center. In the longitudinal section (X-Z plane), two rotating vortices were formed within the molten region and the front one was observed more vigorous than the rear one. This phenomenon can be attributed to the effect of the moving laser source, and accordingly make the molten materials enter the melt pool from the front part and goes through the convection pattern, ultimately resolidifying on the trailing edge [20], as schematically shown in Fig. 5.

4.2. Cooling rate and liquid lifetime

Compared to the conventional manufacturing methods, SLM technique has great superiority due to its rapid melting and solidification nature. The cooling rates at solid–liquid interface can reach $10^5$ to $10^6$ K/s [21], which exhibits great flexibility in controlling the final microstructures and mechanical properties of SLM-processed parts. According to the curves of cooling rates, it was found that the cooling rates changed from negative to positive when the laser beam approached and moved away from the recorded point (Fig. 6). In fact, the negative cooling rates refer to the melting process, while the positive ones represent the solidification process. On the other hand, the calculated maximum cooling rate decreased from 4.71 × $10^5$ K/s to 1.43 × $10^5$ K/s as LEPUL increased from 250 J/m to 1000 J/m. This is because for a given laser power, a decrease in scan speed will shorten the laser powder interaction time. In this situation, the cooling rates of the molten materials will also decrease.

Moreover, it was noted that the region above the melting line signified the molten area and accordingly the liquid lifetime could be obtained. It was clear that the lifetime of the molten pool increased as an elevated LEPUL was applied, while the aspect ratio of the pool (i.e. melt pool length divided by melt pool width) decreased with the increase of LEPUL (Fig. 7). At a low LEPUL of 250 J/m, the liquid lifetime was only 0.73 ms, which was unfavorable to produce sufficient liquid to moisten the surrounding solid particles. With increasing the LEPUL to 677 J/m, the lifetime of the molten pool increased to 4.02 ms, giving rise to the improved liquid–solid wettability and favoring to increase the amount of
particle rearrangement. However, the existence time of the melt at a higher LEPUL of 1000 J/m could reach to 7.41 ms, which tended to induce excessive liquid formation and a large molten pool (Fig. 3d). In this situation, the vigorous Marangoni flow was difficult to control and it was also not expected. The particulate aggregation is detrimental to obtain the favorable microstructural homogeneity and the attendant good mechanical properties. The key factor accounting for this problem lies in the poor wettability between ceramics and metals. Therefore, a high cooling rate combined with sound liquid lifetime can effectively alleviate clustering of ceramic particles in SLM processed AMCs.

4.3. Molten pool dynamics

During the numerical study on SLM of TiC reinforced AlSi10Mg composites, the maximum operating temperature is far below the melting point of TiC (~3413 K), therefore the TiC particles remain solid or at most partially melt at the surface. It is easy to imagine that the solid TiC particles will migrate in the molten AlSi10Mg before the melt finally solidifies. Besides, the reinforcing particles are actually transported by the carriage of the recirculating flow. Accordingly, the magnitude of the recirculation of the fluid flow significantly affects the dispersion states of the reinforcing particulates.

The physical interpretation of Marangoni number \( (Ma) \) is the intensity of Marangoni flow and can be assessed by [22]:

\[
Ma = \frac{\Delta \gamma \cdot L}{\mu \cdot V}
\]  

(7)

where \( \Delta \gamma \) is the difference of surface tension (N/m), \( L \) signifies the characteristic length of the free surface and is taken as the molten pool length (m), \( V \) is kinematic viscosity \((m^2/s)\), which equals to the value of \( \mu \) divided by \( \rho \). Based on Takamichi and Roderick’s results [23], the dynamic viscosity of the melt \((\mu, Pa \ s)\) is defined by:

\[
\mu = \frac{16\gamma}{15} \sqrt{\frac{m}{kT}}
\]

(8)
where \( m \) is the atomic mass, \( k \) represents the Boltzmann constant and \( T \) is the liquid temperature. From Eq. (8), one can notice that either a higher \( T \) or a lower \( c \) leads to a smaller \( l \). Due to unavailability of the surface tension of AlSi10Mg, the surface tension of Al–Si alloy is chosen and can be expressed as follow [24]:

\[
\gamma = \frac{868}{C_0} \frac{1}{T_m} + \frac{138}{C_1} \frac{1}{T_m^3} > 873.2 \text{ K}
\]  

During SLM of TiC/AlSi10Mg composites, the thermal capillary force for the arrangement of TiC particles and the friction force due to the flow of Al liquid around TiC particles are the two main forces acting on TiC solids. According to Anestiev and Froyen’s study [14], the capillary force is found to be inverse proportion to \( \gamma \), while the friction force is proportional to \( l \). Consequently, a decrease in \( \gamma \) and \( l \) (Eqs. (7) and (8)), which can be realized by increasing the applied LEPUL, enhances the Marangoni force and, meanwhile, lowers the friction force (Fig. 8a). The combined effect of these two factors is beneficial to the sufficient rearrangement of TiC particles, thus tending to produce a homogeneous structure free of the apparent aggregation of TiC solids. On the other hand, the surface velocity of the melt at various LEPUL is provided in Fig. 8b. The maximum velocity of the liquid increased from 1.63 m/s to 2.66 m/s as the LEPUL varied from 250 J/m to 1000 J/m. The variation of velocity magnitude with the applied LEPUL further indicates that the Marangoni flow indeed becomes more vigorous with increasing the LEPUL.

4.4. Rearrangement of reinforcing particles

Due to the recirculation of the fluid, the solid reinforcing particles are believed to recirculate a great many times in the melt pool before the melt finally solidifies. This phenomenon significantly affects the dispersion state of the reinforcing particulates in the matrix. Therefore, estimating the detailed recirculating times of the solid particles within the molten region and studying their effect on the particulate distribution state are quite necessary.

A theoretic calculation approach estimating the intensity of fluid flow will be introduced in the following. The melt pool length and width at LEPUL of 667 J/m are measured from the temperature contour plot of the pool surface (Fig. 9a). Similarly, the molten pool dimensions at different SLM processing parameters are obtained. The effective diameter of the melt pool \( d_{eff} \) is calculated as the average value of molten pool length and width. The longest circumference within the melt pool is equal to the distance along solid/liquid interface. Thus, the effective circumference \( C_{eff} \) of the pool is attained through \( C_{eff} = \pi \cdot d_{eff}/2 \) and the variation of...
$C_{\text{eff}}$ with applied LEPUL is plotted in Fig. 9b. In the laser induced pool, the coupled motion of fluid flow and particles is mainly dominated by the surface tension. The movement distance of the melt can be acquired through multiplying liquid lifetime by the average melt velocity. Accordingly, the theoretical recirculating times of the solid particulates can be calculated as the value of movement distance of the melt divided by $C_{\text{eff}}$. The quantitative analysis is beneficial to understand the particles migration behavior within the melt pool. It can be concluded that higher recirculation times imply rapid rearrangement and the resultant uniform dispersion of the reinforcement in the ultimately solidified matrix.

Fig. 10 schematically presents the dispersion state of TiC reinforcement predicted by the numerical simulations. It is apparent that the dispersion homogeneity of the reinforcing particulates is significantly affected by the applied LEPUL. Selective laser melting at a low LEPUL of 250 J/m weakens the Marangoni convection and the attendant thermal capillary forces, thereby slowing down the fluid flow and particulates migration. The particles only recirculate three times combined with the flow according to the theoretic calculation. Consequently, the solid TiC particles tend to sink down to the bottom of the molten pool under the action of gravity force, resulting in a severe agglomeration of the reinforcing particulates (Fig. 10a). With increasing the applied LEPUL to 500 J/m, the recirculation is up to seven times and thus the distribution state of TiC reinforcement can be improved (Fig. 10b). Meanwhile, the mean particle size of TiC particles slightly increased to 82 nm under this condition (Fig. 11e). With a further increase in LEPUL to 667 J/m, the clusters of reinforcing particles underwent a significant deagglomeration, leading to a uniform dispersion of the TiC reinforcement throughout the solidified matrix (Fig. 10c). The average size of TiC reinforcing particles of 90 nm still maintained within the nanometer scale. However, as an elevated LEPUL of 1000 J/m was applied, although the homogeneously dispersed TiC particulates in the matrix could still be obtained, the mean particle size increased remarkably to 134 nm (Fig. 11e), indicating that the control. Meanwhile, the nanoscale TiC grains may become coarsened due to the excessive laser energy input (Fig. 10d), which is detrimental to the final microstructure and the resultant mechanical properties.

4.5. Experimental investigations of SLM-manufactured parts

Fig. 11 a–d presents the dispersion states and typical microstructures of TiC reinforcement on the cross-section of SLM-processed TiC/AlSi10Mg parts at various laser processing parameters. The average sizes of TiC reinforcing particulates are measured during the SEM analysis and the variation of mean particle sizes of TiC reinforcement with the applied LEPUL is depicted in Fig. 11e. It was clear that the dispersion state and particle size of the ceramic reinforcement were sensitive to the applied LEPUL. At a relatively low LEPUL of 250 J/m, significant agglomerations of TiC reinforcing particles were visible in the Al matrix, as selectively shown in Fig. 11a. Nevertheless, the average size of TiC reinforcement was quantitatively determined as 70 nm (Fig. 11e), exhibiting the ultrafine nanostructure. As LEPUL increased to 500 J/m, the distribution homogeneity of the TiC particles was improved, although some small-scaled particulate clusters were still found in the solidified aluminum matrix (Fig. 11b). Meanwhile, the mean particle size of TiC particles slightly increased to 82 nm under this condition (Fig. 11e). With a further increase in LEPUL to 667 J/m, the clusters of reinforcing particles underwent a significant deagglomeration, leading to a uniform dispersion of the TiC reinforcement throughout the solidified matrix (Fig. 11c). The average size of TiC reinforcing particles of 90 nm still maintained within the nanometer scale. However, as an elevated LEPUL of 1000 J/m was applied, although the homogeneously dispersed TiC particulates in the matrix could still be obtained, the mean particle size increased remarkably to 134 nm (Fig. 11e), indicating that the
TiC reinforcement has lost its original superior nanostructure after the SLM process.

It can be concluded that a proper increase in the applied LEPUL from 250 J/m to 677 J/m is beneficial to deagglomerate the clusters of the reinforcing particles and lead to a uniform distribution of the TiC reinforcement in the finally solidified Al matrix. However, an excessive laser energy input at 1000 J/m results in a pronounced coarsening of the TiC grains and resultant disappearance of the ultrafine nanostructure of the reinforcement. On one hand, the existence of the Marangoni convection within the melt pool induces thermal capillary force exerted on the liquid \[ \text{(Eq. (8))} \] and a higher \[ \text{Ma} \] \[ \text{(Eq. (7))} \], which facilitate a sufficient removal of the molten liquid in conjunction with the solid particulates. Consequently, the rapid rearrangement of reinforcing particles in the melt favors homogenizing the dispersion of nanoscale TiC reinforcement in the Al matrix. On the other hand, the rapid solidification nature of SLM induces extremely high undercooling degree and resultant solidification rate at solid–liquid interface, facilitating to shorten the growth time for TiC grains and, thus, making great contribution to the development of the nanostructured TiC reinforcement. With the increase of the applied LEPUL, the elevated thermalization of laser energy results in the

![Fig. 11. FE-SEM images showing dispersions and typical morphologies of TiC reinforcing particulates on the cross-section of SLM-processed TiC/AlSi10Mg parts under various laser processing conditions: (a) \( P = 100 \text{ W}, V = 400 \text{ mm/s}, \text{LEPUL} = 250 \text{ J/m} \); (b) \( P = 100 \text{ W}, V = 200 \text{ mm/s}, \text{LEPUL} = 500 \text{ J/m} \); (c) \( P = 100 \text{ W}, V = 150 \text{ mm/s}, \text{LEPUL} = 667 \text{ J/m} \); (d) \( P = 100 \text{ W}, V = 100 \text{ mm/s}, \text{LEPUL} = 1000 \text{ J/m} \). (e) Variation of average particle sizes of TiC reinforcement with the applied laser energy per unit length (LEPUL).](image)
significant heat accumulation within the melt pool. In this situation, the conduction of heat through the aluminum substrate or previously fabricated layers is significantly weakened. According to the numerical results, the maximum cooling rates ranged from $4.71 \times 10^6$ K/s to $1.94 \times 10^8$ K/s as the applied LEPUL increased from 250 J/m to 677 J/m (Fig. 6a–c), the average size of TiC reinforcement (70–90 nm) was well within the nanometer scale. However, as the LEPUL exceeded 1000 J/m, the maximum cooling rate markedly decreased to $1.43 \times 10^6$ K/s (Fig. 6d) and, accordingly, the nanoscale TiC grains tend to undergo a rapid growth to the relatively coarsened submicron structure with the mean particle size of 134 nm. Therefore, a proper care should be paid in the reasonable selection of laser processing conditions to determine a suitable process window, in order to yield moderate temperature and velocity fields to avoid the severe aggregation of the reinforcement in the metal matrix.

5. Conclusions

(1) Both the maximum SLM working temperature and velocity of molten material increased as a higher LEPUL was applied. The convection pattern of the fluid flow was radially outward and two symmetric vortices were formed on the cross-section of the melt pool.

(2) As an appropriate LEPUL of 667 J/m was applied, the elevated SLM operating temperature intensified the Marangoni convection and induced sufficient liquid flow with a low dynamic viscosity. The TiC reinforcement showed great tendency to become homogeneously dispersed in the melt due to the high recirculating times of the reinforcing particulates within the Al melt.

(3) As the applied LEPUL increased from 250 J/m to 1000 J/m, the TiC reinforcement in SLM-manufactured components experienced a microstructural variation from the standard nanoscale structure with a mean particle size of 70–90 nm to the relatively coarsened submicron morphology with an average particle size of 134 nm. The distribution of TiC reinforcement changed from a severe aggregation to a uniform dispersion in the finally solidified matrix with the increase of LEPUL.

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