Thermal evolution behavior and fluid dynamics during laser additive manufacturing of Al-based nanocomposites: Underlying role of reinforcement weight fraction

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(Received 15 September 2015; accepted 1 December 2015; published online 21 December 2015)

In this study, a three-dimensional transient computational fluid dynamics model was established to investigate the influence of reinforcement weight fraction on thermal evolution behavior and fluid dynamics during selective laser melting (SLM) additive manufacturing of TiC/AlSi10Mg nanocomposites. The powder-to-solid transition and nonlinear variation of thermal physical properties of as-used materials were considered in the numerical model, using the Gaussian distributed volumetric heat source. The simulation results showed that the increase of operating temperature and the resultant formation of larger melt pool were caused by the increase of weight fraction of reinforcement. The Marangoni convection was intensified using a larger reinforcement content, accelerating the coupled motion of fluid and solid particles. The circular flows appeared when the TiC content reached 5.0 wt. % and the larger-sized circular flows were present as the reinforcement content increased to 7.5 wt. %. The experimental study on surface morphologies and microstructures on the polished sections of SLM-processed TiC/AlSi10Mg nanocomposite parts was performed. A considerably dense and smooth surface free of any balling effect and pore formation was obtained when the reinforcement content was optimized at 5.0 wt. %, due to the sufficient liquid formation and moderate Marangoni flow. Novel ring-structured reinforcing particulates were tailored because of the combined action of the attractive effect of centripetal force and repulsive force, which was consistent with the simulation results. © 2015 AIP Publishing LLC.

http://dx.doi.org/10.1063/1.4937905

I. INTRODUCTION

With the ever-growing application demands in automotive, aerospace, and aircraft industries, aluminum (Al) and its alloys have been extensively applied because of their unique characteristics such as light weight, high strength-to-weight ratio, excellent formability, and corrosion resistance.1,2 Nevertheless, the limited mechanical strength of Al is regarded as a serious drawback for its applications. Aluminum matrix composites (AMCs) reinforced with harder and stiffer ceramics are accordingly developed to obtain higher specific strength of Al. TiC ceramic exhibits good wettability and thermodynamic stability within Al melt, making it a suitable candidate material to be used as a reinforcing phase for Al matrix.3 Recent studies have ascertained that decreasing the size of ceramic particulates, especially from micrometer to nanometer level, can lead to a substantial improvement in mechanical properties of AMCs.4 Therefore, the development of nanocrystalline TiC reinforced Al based composites, which are termed as nanocomposites, is of great significance. Furthermore, a uniform distribution of nanoparticles throughout metal matrix is essential to achieve the desirable mechanical performance of nanocomposites. Nevertheless, the uncontrolled aggregation of nanoscale reinforcement is a serious problem for the conventional processing methods.5,6 Therefore, it is imperative to find a novel manufacturing technology to obtain the tailored homogeneous nanostructure.

Selective laser melting (SLM), as an emerging additive manufacturing (AM) technique, exhibits a great flexibility in fabricating complex shaped components directly from loose powder.7–9 SLM process is characterized by an extremely rapid melting and solidification metallurgical nature. Therefore, SLM is capable of providing more opportunities for fabricating high-performance AMCs with unique microstructures.10–12 However, the laser-induced melt pool generally involves a non-equilibrium physical and chemical metallurgical process, which presents multiple modes of heat and mass transfer.13 Within the melt pool, the fluid flow is mainly dominated by thermo-capillary force, viscous drag force, and buoyancy. The rearrangement of reinforcing particles and their final dispersion state in matrix are influenced significantly by the dynamics of melt pool. Nevertheless, the migration path of particles is difficult to be observed through experiment methods, since the melt pool has extremely small dimensions of several tens of micrometres to hundreds of micrometres. Therefore, numerical simulation of SLM process is an effective approach to visualize the melt flow behavior and understand the underlying mechanisms of particle movement.

The monitoring of laser processes has been studied actively since 1980s in several institutes around the world.
Presently, monitoring has been commercially applied to the newest laser processes, including additive manufacturing and laser welding.\textsuperscript{14–16} Charged-couple device (CCD) or complementary metal-oxide semiconductor (CMOS) cameras with high resolution are usually used for process monitoring and real-time control. However, the dimensions of laser-induced melt pool are extremely small and, meanwhile, the duration of laser beam on any powder particle is considerably short, which is typically between 0.5 and 25 ms. Therefore, visualizing the melt pool during SLM by experimental methods is very difficult. Alternatively, some numerical methods have been developed to simulate the heat transfer and fluid flow within laser-induced melt pool. Kovalev et al.\textsuperscript{17} numerically investigated the convective heat and mass transfer within laser-induced melt pool. They pointed out that the multivortex fluid flows were mainly responsible for the movement of the reinforcing particles, which influenced the final distribution state of the reinforcing particles within the metal matrix. Cui et al.\textsuperscript{18} proposed a new model in which the surface tension, gravity, recoil force, and buoyancy were considered. The particle movement was totally coupled with the detailed fluid motion during laser welding process. Qiu et al.\textsuperscript{19} studied the interaction between laser beam and powder particles during SLM process through both high-speed imaging observation and modeling approaches. The formation of pores and rough surface were strongly associated with the unstable melt flow and splashing of melt. Up to now, little previous work has been focused on the study of fluid flow and resultant particle movement within melt pool during SLM process. The underlying mechanism for the development of the final distribution state of the reinforcing particles still needs to be clarified.

The previous studies have proved that laser AM of Al alloys is much more difficult than AM processing of nickel alloys, titanium alloys, or steels. This is mainly ascribed to the low laser absorption (only 9\%\%) of Al powder to laser beam.\textsuperscript{10} An effective approach to solve this problem is to add a proper weight fraction of reinforcing phase that has higher laser absorption than the matrix phase. In this study, the nanoscale TiC ceramic has been chosen as the reinforcing phase because of its high laser absorption nearly of 82\%\%.\textsuperscript{30} However, the existence of the considerably large van der Waals attractive forces is extremely easy to cause the uncontrolled aggregation of nanoparticles. The optimization of the weight fraction of TiC reinforcing phase is beneficial for obtaining microstructural homogeneity and, meanwhile, increasing laser absorption of the whole powder system.

In this work, a three-dimensional transient computational fluid dynamics (CFD) model was established to investigate the heat transfer and fluid flow during SLM of TiC reinforced AlSi10Mg nanocomposites. The influence of reinforcement weight fraction on the thermal evolution behavior and fluid dynamics was clarified by analyzing the obtained simulation results. The detailed flow field around reinforcing particles and the formation mechanism of novel ring-structure were disclosed through both numerical modeling and experimental validation.

II. MODELING APPROACHES

A. Physical model of SLM

The SLM process is schematically presented in Fig. 1(a), exhibiting a unique layer-by-layer incremental deposition manner. During SLM process, a Gaussian distributed laser beam moves through a predefined spatial domain with a certain velocity. The laser energy is absorbed by individual particles within the powder system. When the working temperature reaches the melting temperature of the matrix phase ($C_2^A_{873}$ K), the AlSi10Mg particles start to become melted. However, the reinforcing phase TiC remains in solid or, at most, becomes partially melted on the surface, due to a higher melting point ($C_2^A_{3413}$ K). Within laser-induced melt pool, the material flow is mainly driven by gravity, buoyancy force, and surface tension caused by the temperature gradient.\textsuperscript{20} The underlying mechanism of the coupled movement of fluid and reinforcing particles needs to be clarified using numerical methods.

B. Model setup and governing equations

To better understand the thermal fluid dynamics during SLM process, a mathematical model is established using the FLUENT software. The geometrical model is schematically shown in Fig. 1(b). The calculation is performed on a domain of dimensions $1 \times 1 \times 0.08 \text{mm}^3$ with a uniform structured mesh containing 30 000 hexahedral cells. The finite volume method is applied to solve the numerical problem through a set of conservation equations.
The motion of fluid generally follows three basic physical conservation laws, i.e., the conservation of mass, momentum, and energy. Based on the above assumptions, the mass, momentum, and energy conservation can be expressed in three-dimensional Cartesian coordinate system as follows:

Continuity equation:
\[
\frac{\partial \rho}{\partial t} + \nabla \cdot (\rho \mathbf{V}) = M_s. \tag{1}
\]

Momentum equation:
\[
\rho \left( \frac{\partial \mathbf{V}}{\partial t} + \mathbf{V} \cdot \nabla \mathbf{V} \right) = -\nabla p + \mu \nabla^2 \mathbf{V} + \mathbf{M}_s + \mathbf{F}. \tag{2}
\]

Energy equation:
\[
\rho \left( \frac{\partial T}{\partial t} + \mathbf{V} \cdot \nabla T \right) = \nabla \cdot (\kappa \nabla T) + S_H, \tag{3}
\]

where \( \rho, \kappa, \mu, \) and \( \rho \) represent density, thermal conductivity, dynamic viscosity, and pressure, respectively. \( M_s \) is a mass source, \( \mathbf{V} \) is the motion velocity of the melt, \( \mathbf{F} \) is the body force (e.g., gravity and buoyancy forces), \( S_H \) is the source item of energy equation and can be defined by
\[
S_H = -\rho \left( \frac{\partial}{\partial t} \Delta H + \nabla \cdot (\mathbf{V} \Delta H) \right), \tag{4}
\]

where \( \Delta H \) is the latent heat of phase change. Since SLM process involves rapid melting and solidification phenomena, the latent heat of phase change must be considered in the numerical model.

C. Boundary conditions

During SLM process, the loose powder is freely deposited on the previously fabricated layer, producing a high porosity (\( \sim 40\% \)). The typical particle morphology of the starting mixed powder is presented in Fig. 2(a). The laser beam penetrates into the voids between powder particles through the multiple reflection mechanism. The penetration depth approximately equals to several particle diameters and, then, the laser beam directly interacts with powder particles. A schematic of the laser radiation transfer within a powder layer is shown in Fig. 2(b). In the present simulation model, the volumetric heat source is applied and the boundary condition of the top surface is accordingly defined by

\[
-Q_{\text{eff}} \left( \frac{\partial T}{\partial z} \right)_{z=0} = \frac{3\pi p}{\pi \omega^2} \exp \left( -3 \frac{(x^2 + y^2)}{\omega^2} \right)_{z=0} - h_c(T - T_0) - \sigma_e v(T^4 - T_0^4), \tag{5}
\]

where \( A \) is the laser absorptivity of powder system. The laser absorption of powder system is determined by both physical properties of AlSi10Mg and TiC powder and can be calculated by
\[
A = \sum x_i A_i, \tag{6}
\]

where \( x \) represents the volume fraction of one phase; \( \sum x_i = 1, n \) is the phase number; and \( A_i \) is the laser absorption of the \( i \) phase. The nominal compositions of as-used AlSi10Mg alloy powder are listed in Table I.

When the heat is absorbed by powder material, the laser energy decreases as the beam penetrates deep into the powder bed. The energy input to the deposited powder layer depends on the material absorbance, particle packing, i.e., porosity (\( \Phi \)) and particle diameter (\( D_p \)). Assuming that the powder bed consists of spherical particles, the optical extinction coefficient is given by
\[
\beta = \frac{3}{2} \frac{1 - \Phi}{\Phi \frac{D_p}{2}}. \tag{7}
\]

where porosity \( \Phi = 1 - \pi D_p^3 \rho / 6 \). The optical thickness can be expressed by
\[
\lambda = \beta \cdot l_p = \frac{3}{2} \frac{1 - \Phi}{\Phi \frac{D_p}{2}}. \tag{8}
\]

The Marangoni force caused by \( \partial \gamma / \partial T \) is fully integrated in this model through the user defined functions (UDFs). The surface tension is defined by
\[
-\mu \frac{\partial u}{\partial z} = \frac{\partial \gamma}{\partial T} \frac{\partial T}{\partial x}; \quad -\mu \frac{\partial v}{\partial z} = \frac{\partial \gamma}{\partial T} \frac{\partial T}{\partial y}. \tag{9}
\]

| TABLE I. Nominal compositions of as-used AlSi10Mg alloy powder. |
|---------------------------|----------------|----------------|----------------|----------------|----------------|----------------|----------------|
| Element | Si | Fe | Cu | Mn | Mg | Zn | Ti | Al |
| Weight | 10.08 | 0.16 | 0.001 | 0.002 | 0.35 | 0.002 | 0.01 | Balance |
| fraction (%) | |

The optical thickness can be expressed by
\[
\lambda = \beta \cdot l_p = \frac{3}{2} \frac{1 - \Phi}{\Phi \frac{D_p}{2}}. \tag{8}
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\]
D. Thermal physical properties and SLM processing parameters

Due to a high laser energy input, SLM process involves a powder-solid transition phenomenon within an extremely short time. The thermal physical properties of as-used materials undergo nonlinear variations because of the temperature change and phase transformation. Some important physical properties, including thermal conductivity, specific heat capacity, density, and viscosity, are considered to be temperature-dependent. Some other thermal physical properties are taken as constants in order to simplify the computational complexity. The effective thermal conductivity of loose metallic powder is controlled by gas in pores and is about 100 times lower than that of the dense material. Considering the initial porosity and powder packing density, the effective thermal conductivity of powder system can be expressed by

\[
\frac{k_{\text{eff}}}{k_g} = \left(1 - \sqrt{1 - \Phi}\right) \left(1 + \frac{\Phi k_s}{k_g}\right) + \sqrt{1 - \Phi} \left(\frac{2}{1 - \frac{k_g}{k_s}} \left[\frac{1}{1 - \frac{k_s}{k_g}} \ln\left(\frac{k_g}{k_s}\right) - 1\right] + \frac{k_s}{k_g}\right),
\]

where the radiative heat transfer among powder particles is also considered in this model and can be defined by

\[
k_r = 4F_0 \sigma_c T_p^4 D_p,
\]

The effective thermal conductivity of powder system \(k_{\text{eff}}\) equals to \(k_{\text{powder}}\) when the working temperature is below the sintering temperature \(T_s\). During heating stage, \(k_{\text{eff}}\) increases linearly because of the transformation from powder phase to bulk phase when the temperature is between \(T_s\) and \(T_m\). Moreover, \(k_{\text{eff}}\) is taken as \(k_{\text{bulk}}\) when the temperature exceeds \(T_m\). Therefore, the variation of \(k_{\text{eff}}\) with temperature follows curve ABCD shown in Fig. 3. Some other thermal physical properties of as-used materials are given in Table II.

III. EXPERIMENTAL PROCEDURES

A. Powder preparation

The 99.0% purity TiC nanopowder with a near spherical shape and a mean particle diameter of 50 nm and the 99.7% purity AlSi10Mg powder with a spherical shape and an average particle size of 30 \(\mu\)m were used in this experiment. The two components were mechanically mixed according to TiC:AlSi10Mg weight ratios of 2.5:97.5, 5:95, and 7.5:92.5 in a Pulverisette 4 vario-planetary mill (Fritsch GmbH, Germany) using a ball-to-powder weight ratio of 1:1, a rotation speed of main disk of 200 rpm, and a mixing time of 4 h.

B. Laser processing and microstructural characterization

The SLM system used in this study mainly consisted of an IPG YLR-200-SM ytterbium fiber laser with a maximum output power of \(\sim\)200 W, a SCANLAB hurrySCAN 20 laser scanner, an inert argon gas protection system, an automatic powder spreading device, and a computer system for process control. Details concerning SLM processing procedures have been addressed in the previous paper. Based on a series of preliminary SLM experiments, the following suitable processing parameters were chosen to prepare specimens: laser power of 150 W, scan speed of 150 mm/s, spot size of 70 \(\mu\)m, powder layer thickness of 50 \(\mu\)m, and hatching space of 50 \(\mu\)m.

The samples for metallographic examinations were prepared according to the standard procedures and etched with a mixture of HF (2 ml), HCl (3 ml), HNO\(_3\) (5 ml), and distilled water (190 ml) for 10 s. High-resolution study of surface morphologies and interior ultrafine nanostructures of SLM-processed nanocomposite samples was performed using a Hitachi S-4800 field emission scanning electron microscope (FE-SEM).

![FIG. 3. Variation of effective thermal conductivity of as-used TiC/AlSi10Mg material with temperature.](image-url)
IV. RESULTS AND DISCUSSION

A. Influence of reinforcement weight fraction on thermal behavior

Figure 4 shows the plots of surface temperature contour of laser-powder interaction zone with different TiC contents. The black circle indicates the isotherm of the melting temperature of AlSi10Mg (873 K), which also represents the melt pool boundary. It was obvious that the maximum temperatures and melt pool dimensions changed with the TiC contents. At a low TiC content of 2.5 wt. %, the peak temperature of the melt pool was 1460 K, and the melt pool length and width were only 178.5 µm and 161.2 µm, respectively (Fig. 4(a)). As the TiC content increased to 5.0 wt. %, the maximum temperature of the molten materials reached 1580 K. Meanwhile, laser melting induced the formation of a larger melt pool with a length of ~200.8 µm and a width of ~200.8 µm (Fig. 4(b)). When 7.5 wt. % TiC was used, a considerably larger melt pool with 281.2 µm in length and 250.5 µm in width was generated (Fig. 4(c)). In this instance, the maximum operating temperature during SLM process increased to 1670 K. During SLM process, the laser energy is directly absorbed by solid particles through both bulk coupling and powder coupling mechanisms.26 However, due to the considerably high reflectivity (91%) of Al powder to laser beam,10 the input laser energy cannot be effectively absorbed by powder bed. Mixing the TiC ceramic phase, which has a high laser absorptivity (~82%),30 with the Al powder is a useful approach to improve the laser absorptivity of a whole powder system. The laser absorption of a powder system is determined by physical properties of both AlSi10Mg and TiC powder, which can be calculated by \[ A = \sum x_i A_i. \]

Therefore, an increase of the working temperature and the resultant formation of a larger melt pool are ascribed to the increase of weight fraction of the reinforcement.

Furthermore, it was noted that the plots of temperature contour stretched slightly to a comet-shaped tail along laser scanning direction. The isotherms were more closely spaced in the front of the ellipses (powder phase) than those in the rear part (dense phase). It accordingly indicated that the temperature gradients in front of the melt pool were larger than those in the back side, which was ascribed to the variation of thermal conductivity caused by powder-to-solid transition.

B. Influence of reinforcement weight fraction on fluid dynamics

When laser beam irradiates the powder layer and moves at a constant speed, strong Marangoni flows are formed due to the surface tension gradients within the transient melt pool.
under laser beam irradiation. In order to comprehensively understand the thermal and rheological properties during SLM process, it is necessary to investigate the characteristics of material flow, e.g., surface tension and dynamic viscosity. The distribution of surface temperature and surface tension on the top surface of powder layer using different amounts of TiC reinforcement is shown in Fig. 5. It was clear that the surface tension was inversely proportional to the working temperature of the melt pool. Therefore, the colder liquid with a high surface tension near the boundary of the melt pool tends to pull the fluid away from the pool center. In this situation, the fluid flow exhibits a radially outward flow pattern, as shown in Fig. 6. Figure 6 presents the variation of velocity field within the melt pool in the longitudinal section with TiC contents. It was found that increasing the TiC content from 2.5 wt. % to 7.5 wt. % enhanced the maximum velocity of the melt flow from 1.8 m/s to 2.6 m/s.

The convective intensity of Marangoni flow can be described by the dimensionless parameter Marangoni number \( (Ma) \), which is expressed by

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

where \( \Delta \gamma \) means the difference of surface tensions and \( L \) is the characteristic length of the melt pool surface and can be taken as the melt pool length. During SLM process, the dynamic viscosity \( \mu \) of the fluid flow is dependent on temperature \( T \) and can be assessed by

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

FIG. 5. Distribution of surface temperature and surface tension during SLM on the top surface of TiC/AlSi10Mg powder layer containing different amount of TiC reinforcement.

FIG. 6. Variation of velocity fields within the melt pool in the longitudinal section with TiC contents: (a) 2.5 wt. %, (b) 5.0 wt. %, and (c) 7.5 wt. %. (d) Variation of Marangoni number \( (Ma) \) with reinforcement weight fraction.
where \( m \) is the atomic mass, \( k \) represents the Boltzmann constant, \( \nu \) signifies kinematic viscosity (m\(^2\)/s) that equals to the value of \( \mu \) divided by \( \rho \). Due to unavailability of the surface tension of AlSi10Mg, the surface tension of Al-Si alloy is chosen and applied approximatively in the simulation model. The surface tension is proportional to the working temperature and can be expressed as follow:

\[
\gamma = [868 - 0.152(T - T_m)] \times 10^{-3} T > 873.2 \text{ K}. \tag{14}
\]

During SLM of TiC/AlSi10Mg powder system, increasing the weight fraction of TiC reinforcement can effectively enhance laser absorptivity of powder bed, producing a higher operating temperature within the melt pool. From Eq. (14), it reveals that an appropriate increase of the working temperature leads to a lower \( \gamma \) when \( T \) exceeds the melting temperature of AlSi10Mg alloy. Moreover, according to Eq. (13), a higher \( T \) or a lower \( \gamma \) favors a decrease of the dynamic viscosity of the melt, hence enhancing the Marangoni intensity (Eq. (12)). Figure 6(d) shows the variation of Marangoni number with reinforcement weight fraction. It reveals that the intensity of Marangoni flow increases with an increase of reinforcement content. Under this condition, the flow rate of the molten material is elevated, accelerating the infiltration of liquid phase into powder particles. Therefore, the interior pores can be effectively alleviated and the densification level of the finally solidified parts can be improved significantly.

On the other hand, the capillary force exerted on solid particles by the wetting liquid is inversely proportional to \( \gamma \). Decreasing \( \gamma \) is beneficial for increasing capillary force, thereby enhancing the particle rearrangement rate. Furthermore, due to the sound wettability and thermodynamic stability of TiC solid within Al melt, the clusters and aggregates of TiC particles can be alleviated efficiently.

C. Particle rearrangement mechanism

In order to investigate the influence of reinforcement weight fraction on particle rearrangement mechanisms, the typical characteristics of velocity vector plots around a TiC reinforcing particle within the melt pool at different TiC contents are depicted in Fig. 7. It was clear that the fluid flow pattern was sensitive to the applied reinforcement content. At a low TiC level of 2.5 wt. %, the Marangoni flow passed through the TiC particle with a maximum velocity of 1.76 m/s. The flow direction was almost along the same direction. At a higher TiC content of 5.0 wt. %, the maximum velocity of Marangoni flow reached 2.24 m/s. Moreover, it was interesting to find that a small circular flow was formed when the convective flow passed through the TiC particle. At an even higher TiC content of 7.5 wt. %, the Marangoni flow was intensified and a larger circular flow was present around the TiC particle.

![Velocity vector plots around a TiC reinforcing particle within the melt pool at different TiC contents: (a) 2.5 wt. %, (b) 5.0 wt. %, and (c) 7.5 wt. %](image-url)
Within laser-generated melt pool containing both liquid (Al-Si-Mg) and solid (TiC) phases, the fluid flow is mainly dominated by the thermo-capillary force. The solid particles tend to migrate with the coupled motion of the melt flow. The Marangoni force and viscous drag force are two key forces acting on TiC solids. By increasing the reinforcement content, the presence of a higher amount of absorbed laser energy leads to an enhancement of working temperature of melt pool and a larger amount of liquid formation having a lower dynamic viscosity. A stronger Marangoni convection within the melt pool is accordingly formed due to the larger temperature gradients and surface tension gradients. When using a low reinforcement content of 2.5 wt. %, the rearrangement of reinforcing particles is significantly restricted because of the limited liquid formation with a considerably high melt viscosity. At a higher reinforcement content of 5.0 wt. %, an elevated Marangoni force and a lower viscous drag force favor the movement of TiC particles within the melt. Meanwhile, a circular flow appears due to the interaction between the solid particles and the strong material flow (Fig. 7(b)). The TiC particles tend to be trapped into the circular flow, forming a unique ring structure in the finally solidified materials. By further increasing the reinforcement content to 7.5 wt. %, a larger circular flow appears around TiC particles, as depicted in Fig. 7(c). In this situation, a larger ring structure tends to be formed once the TiC particles enter and rotate with the circular flow due to the effect of centripetal force. However, the formation of an excessive amount of molten Al phase with a too low liquid viscosity tends to cause a typical metallurgical defect of SLM process, i.e., balling phenomenon.

D. Experimental investigations of SLM-processed parts

Figure 8 illustrates the characteristic surface morphologies of SLM-processed TiC/AlSi10Mg parts with variation of TiC additions. At a relative low TiC addition of 2.5 wt. %, the surface was considerably rough with the appearance of sintering necks and incoherent scan tracks, resulting in the formation of a large porosity (Fig. 8(a)). By increasing the TiC addition to 5.0 wt. %, a smooth and dense surface, which was free of apparent pores, was present in Fig. 8(b). At an even higher TiC addition of 7.5 wt. %, although the surface remained sufficiently dense, a large amount of small metallic balls were observed on the surface (Fig. 8(c)). It was reasonable to conclude that the “balling” effect initiated in this instance.

During SLM process, the laser energy is directly absorbed by the solid particles via both bulk coupling and powder coupling mechanisms. The weight fraction of the reinforcing phase that has a higher laser absorptivity influences the laser absorptance of a whole powder system. A lower TiC addition results in a lower SLM operating temperature and resultant limited liquid formation. Meanwhile, the molten materials possess a high dynamic viscosity in this instance, thereby hindering the sufficient flow of the melt. Under this condition, the insufficient amount of liquid phase cannot effectively fill the voids among solid particles, accordingly producing porosity in the finally solidified materials. By increasing the reinforcement content, the enhancement of liquid formation with a decreased liquid viscosity is beneficial for improving the spreading of the melt and the overall rheological performance of the pool. The formation of a dense and smooth surface is reasonable to support the above analysis. However, when the TiC content exceeds a threshold value, the intensive Marangoni streams within the melt pool consisting of an excessive amount liquid phase tend to cause the Rayleigh-Plateau capillary instability and the resultant “balling” effect. The obtained results, as shown in Fig. 8(b), reveal that the combination of a moderate Marangoni flow and a proper amount of molten materials, which can be realized using 5.0 wt. %
TiC reinforcement, leads to a sound densification level without an apparent balling effect.

In order to further verify the effect of reinforcement content on particle migration behavior, the typical microstructures on the polished sections of SLM-processed TiC/AlSi10Mg nanocomposite parts at different TiC contents are illustrated in Fig. 9. According to the previous study, the dispersed white reinforcing particles were TiC, while the surrounding dark matrix was Al. It was obvious that the reinforcement weight fraction had a significant effect on the distribution state of TiC reinforcing phase within the matrix. At a relatively low TiC content of 2.5 wt. %, the TiC particles showed a disordered distribution and some large aggregates of TiC reinforcement were observed within the matrix (Fig. 9(a)). As the TiC content increased to 5.0 wt. %, the TiC reinforcing phase exhibited a high tendency to become regularly distributed in the matrix, forming a novel ring-structure (Fig. 9(b)). The average diameter of the ring-structure approximated to 0.95 μm. With a further increase in TiC content to 7.5 wt. %, a series of ring-structured reinforcement with a larger average diameter of 1.65 μm was formed along grain boundaries of the matrix (Fig. 9(c)).

A combination of numerical and experimental results reveals that a weak Marangoni convection and an obstructive effect of the liquid phase having a high dynamic viscosity are detrimental to the movement of TiC particles within the melt pool. Therefore, the rearrangement of the reinforcing particles is significantly limited at a low reinforcement content, resulting in the agglomeration of TiC reinforcement. Figure 10 schematically depicts the mechanism for the formation of ring-structured reinforcement. It is clear that the intensified Marangoni flow and resultant lower liquid viscosity favor an efficient movement of TiC particles within the melt pool when using an elevated reinforcement content of 5.0 wt. %. Due to the appearance of the circular flow around TiC reinforcement in this instance, the TiC solids are trapped easily into the circular flow. Moreover, there exists a torque around a single particle due to the combined action of thermo-capillary force and viscous drag force on a non-spherical TiC particle. The presence of torque tends to rotate the reinforcing particles within the melt pool, accelerating the rearrangement of particles. The centripetal force of the circular flow and torque force continuously drive the reinforcing particles, pushing and gathering them around the center of the circular flow to form a ring-structure. Research by Anestiev et al. reveals that the repulsion forces will form between particles when a sufficient amount of Al liquid forms within the melt pool. The repulsive force between two adjacent particles will make the particles to maintain a certain distance from the center of the circular flow. Therefore, the TiC particles are regularly embedded in the finally solidified matrix as a novel ring-structure when the laser beam moves away. With a further increase of reinforcement content to 7.5 wt. %, the larger-sized circular flow contributes to

![FIG. 9. FE-SEM images showing typical microstructures of SLM-processed TiC/AlSi10Mg nanocomposites with different TiC contents: (a) 2.5 wt. %, (b) 5.0 wt. %, and (c) 7.5 wt. %.

![FIG. 10. Schematic of the formation mechanism of novel ring-structured TiC reinforcement within the matrix during SLM process.](image)
the formation of the ring-structured reinforcement with a larger internal diameter.

V. CONCLUSIONS

In the present work, the influence of the TiC reinforcement weight fraction on heat transfer and fluid flow during SLM of TiC/AlSi10Mg nanocomposites has been investigated using a transient three-dimensional CFD model. The SLM experiments were also conducted using the same laser processing parameters as those used in the simulation. The following conclusions were drawn.

(1) The increase of weight fraction of reinforcement was responsible for the increase of working temperature and resultant formation of larger melt pool, due to the enhancement of laser absorption of a whole powder system.

(2) Using a higher TiC reinforcement content intensified Marangoni convection and reduced liquid dynamic viscosity. The maximum velocity magnitude of the melt flow increased on increasing TiC content.

(3) The Marangoni force and viscous drag force were two main forces acting on TiC reinforcing particles at a low TiC content of 2.5 wt. %. Circular flows appeared at a higher reinforcement content and a larger circular flow was formed as the TiC content increased to 7.5 wt. %.

(4) A fully dense and smooth surface free of any balling effect and pore formation was obtained when the TiC reinforcement content was optimized at 5.0 wt. %. The distribution state of reinforcing particles changed from a significant aggregation to a regularly distributed ring-structure as TiC content increased from 2.5 wt. % to 5.0 wt. %. A novel ring-structure with a larger diameter was tailored when a higher reinforcement content of 7.5 wt. % was applied.

ACKNOWLEDGMENTS

The authors gratefully acknowledge the financial support from the National Natural Science Foundation of China (Nos. 51322509 and 51575267), the Outstanding Youth Foundation of Jiangsu Province of China (No. BK20130035), the Program for New Century Excellent Talents in University (No. NCET-13-0854), the Science and Technology Support Program (The Industrial Part), Jiangsu Talents in University (No. NCET-13-0854), the Program for New Century Excellent Youth Foundation of Jiangsu Province of China (No. NCET-13-0854), and the Priority Academic Program Development of Jiangsu Higher Education Institutions.