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Dongdong Gu, Hongqiao Wang, Donghua Dai, et al.

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Densification behavior, microstructure evolution, and wear property of TiC nanoparticle reinforced AlSi10Mg bulk-form nanocomposites prepared by selective laser melting

Dongdong Gu,a) Hongqiao Wang, Donghua Dai, and Fei Chang
College of Materials Science and Technology, Nanjing University of Aeronautics and Astronautics, Yudao Street 29, 210016 Nanjing, People’s Republic of China

Wilhelm Meiners, Yves-Christian Hagedorn, Konrad Wissenbach, Ingomar Kelbassa, and Reinhart Poprawe
Fraunhofer Institute for Laser Technology ILT/Chair for Laser Technology LLT, RWTH Aachen, Steinbachstraße 15, D-57074 Aachen, Germany

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Selective laser melting (SLM), due to its unique additive manufacturing processing philosophy, demonstrates a high potential in producing bulk-form nanocomposites with novel nanostructures and enhanced properties. In this study, the nanoscale TiC particle reinforced AlSi10Mg nanocomposite parts were produced by SLM process. The influence of “laser energy per unit length” (LEPUL) on densification behavior, microstructural evolution, and wear property of SLM-processed nanocomposites was studied. It showed that using an insufficient LEPUL of 250 J/m lowered the SLM densification due to the balling effect and the formation of residual pores. The highest densification level (>98% theoretical density) was achieved for SLM-processed parts processed at the LEPUL of 700 J/m. The TiC reinforcement in SLM-processed parts experienced a structural change from the standard nanoscale particle morphology (the average size 75–92 nm) to the relatively coarsened submicron structure (the mean particle size 161 nm) as the applied LEPUL increased. The nanostructured TiC reinforcement was generally maintained within a wide range of LEPUL from 250 to 700 J/m and the dispersion state of nanoscale TiC reinforcement was homogenized with increasing LEPUL. The sufficiently high densification rate combined with the uniform distribution of nanoscale TiC reinforcement throughout the matrix led to the considerably low coefficient of friction of 0.38 and wear rate of 2.76 × 10^{-5} mm^3 N^{-1} m^{-1} for SLM-processed nanocomposites at 700 J/m. Both the insufficient SLM densification response at a relatively low LEPUL of 250 J/m and the disappearance of nanoscale reinforcement at a high LEPUL of 1000 J/m lowered the wear performance of SLM-processed nanocomposite parts. © 2014 Laser Institute of America. [http://dx.doi.org/10.2351/1.4870877]

Key words: additive manufacturing, selective laser melting (SLM), nanocomposites, aluminum matrix composites, wear

I. INTRODUCTION

In the past few years, there is an increasing demand for the lightweight materials with high strength and stiffness in modern industries.1 Particle reinforced aluminum matrix composites (AMCs), because of its low density, high strength, low coefficient of thermal expansion, and outstanding abrasion resistance, can well meet these requirements and, accordingly, have been widely used in the industrial fields of aerospace, automotive, microelectronics, etc.2–4 As the potentially feasible reinforcement for AMCs, TiC exhibits a number of favorable characteristics, such as high elastic modulus, high hardness, and especially its good wettability and thermodynamic stability within the molten aluminum.5 Normally, the large sized reinforcing particles are used in the conventional processing of TiC particle reinforced AMCs, ranging from several tens of micrometers to hundreds of micrometers. However, due to the limited interfacial wettability and bonding coherence between ceramics and metals, the large ceramic particles are prone to cracking during mechanical loading, resulting in the poor ductility and premature failure of AMCs.6 It has been revealed that the strength and ductility of AMCs are significantly affected by the particle size of the reinforcement; both tensile strength and ductility increase with lowering the particle size.7 Therefore, the application of the refined ceramic reinforcing particles can efficiently favor the improvement of mechanical properties of AMCs, e.g., the reduced particle cracking and the enhanced strengthening. Previous studies have confirmed that the mechanical properties of AMCs can be further enhanced by decreasing the size of ceramic reinforcing particles from micrometer to nanometer levels.8 Such materials are known as nanocomposites.9 The production of nanocomposites with unique microstructural and mechanical properties has inspired considerably research interest. To
obtain the desired properties of nanocomposites, the nanoparticles as the reinforcement should be uniformly dispersed throughout the matrix. Nevertheless, it is rather difficult to achieve a uniform distribution of nanoscale reinforcement throughout a metal matrix. Because of the extremely large van der Waals attractive force between adjacent nanoparticles, the nanoparticles tend to agglomerate into the coarsened clusters, resulting in the microstructural inhomogeneity and even disappearance of original nanostructures. Therefore, it is urgent to find a novel method to avoid particulate aggregation and grain coarsening, ensuring the homogenization of nanoparticles throughout the matrix.

Selective laser melting (SLM), as a newly developed direct digital manufacturing technology, has been treated as one of the most effective powder-based additive manufacturing (AM) methods for metal parts.\textsuperscript{10–17} During SLM process, the 3D parts are built by selectively fusing and consolidation of the thin layers of powder using the high-energy laser beam in a layer-by-layer manner according to the computer-aided design (CAD) models of the parts. Because of the unique processing mechanisms of SLM, e.g., a full melting of powder materials followed by a rapid solidification at a rate up to $10^6$–$10^8$ K/s,\textsuperscript{18} the molten materials tend to experience a particular nonequilibrium metallurgical process during SLM processing, thereby providing a capacity to produce unique aluminum-based nanocomposite parts. However, the previous studies have shown that the feasibility of processing aluminum alloy parts by SLM is considerably lower than that of titanium alloys, nickel alloys, or steels.\textsuperscript{19–21} This is mainly attributed to the unique physical properties of the aluminum powder including (i) the considerably low absorptivity (only 9\%) of the aluminum powder to the laser beam;\textsuperscript{19} (ii) the high heat conductivity of 237 W/(mK)\textsuperscript{22} that is 11-fold compared to Ti and fivefold compared to Fe; and (iii) the high affinity to oxygen and elevated oxidation kinetics. The rapid dissipation of heat away from the molten pool through the previously solidified material and the formation of oxide layer on the top of the pool in the deposited tracks hinder the diffusion of melt and accordingly reduce the wettability and laser processability. Nevertheless, the parts with densities exceeding 97\% can be acquired successfully via SLM process, typically using the casting alloy AlSi10Mg as the starting powder.\textsuperscript{23,24} The AlSi10Mg is selected for SLM because of its favorable weldability, sufficient hardenability, and excellent corrosion resistance. The sound weldability is attributed to the near-eutectic composition of Al and Si, which leads to a narrow solidification range. The addition of small amount of Mg (0.3–0.5 wt. \%) gives rise to Mg$_2$Si precipitates by natural or artificial ageing treatments, therefore, inducing the hardening of aluminum alloys.

In the present work, the material as investigated was broadened from the conventional AlSi10Mg alloy to the novel AlSi10Mg based nanocomposites. The bulk-form TiC/AlSi10Mg nanocomposites with the uniformly distributed nanoscale reinforcement were successfully prepared by SLM process. The microstructural evolution of nanoscale reinforcement in SLM-processed parts at different laser processing parameters was studied, and the attendant densification level and wear/tribological property were assessed. A material–microstructure–property relationship was established to enable the successful production of AMCs parts with nanoscale reinforcement and elevated mechanical performance.

II. EXPERIMENT

A. Powder preparation

The 99.0\% purity TiC nanopowder with a near spherical shape and a mean particle size of 50 nm and the 99.7\% purity AlSi10Mg powder with a spherical shape and an average particle diameter of 30 \( \mu \)m were used. The powder properties were values from the data sheets of the suppliers. The TiC and AlSi10Mg powder components consisting of 5 wt. \% TiC were mechanically mixed in a Pulverisette 4 vari-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 duration of 4 h.

B. Laser processing

The self-developed SLM system consisted mainly of a YLR-200-SM ytterbium fiber laser with a power of \~200 W and a spot size of 70 \( \mu \)m (IPG Laser GmbH, Germany), an automatic powder spreading device, an inert argon gas protection system, and a computer system for process control. The SLM procedures for fabricating bulk-form nanocomposites parts are depicted in Fig. 1. As the specimens were to be built, an aluminum substrate was fixed on the building platform and leveled. The building chamber was then sealed and the argon gas with an outlet pressure of 30 millibars was fed inside, decreasing the O$_2$ content below 20 ppm. Afterward, the nanocomposite powder was deposited on the substrate by the layering mechanism, with a powder layer thickness of 50 \( \mu \)m. The laser beam was then controlled by the CAD data to scan the powder bed surface selectively, forming a two-dimensional profile. A simple linear raster scan pattern was used, with a scan vector length of 4 mm and a hatch spacing of 50 \( \mu \)m. The laser power ($P$) was optimized at 100 W and the scan speeds ($v$) were set at 100, 143, 200, and 400 mm/s by the SLM control program, in order to change the processing conditions during experiments. Four different “laser energy per unit length” (LEPUL) of 1000, 700, 500, 250 J/m, which was defined by\textsuperscript{25}

$$\text{LEPUL} = \frac{P}{v}$$  

(1)

was used to assess the laser energy input to the powder layer being processed. The rectangular specimens with dimensions of 20 mm $\times$ 10 mm $\times$ 5 mm were built in a layer-by-layer manner until completion.

C. Microstructural and mechanical properties characterization

Specimens for metallographic examinations were prepared according to the standard procedures, and etched with a solution composing HF (2 ml), HCl (3 ml), HNO$_3$ (5 ml), and
distilled water (190 ml) for 10 s. A PMG3 optical microscope (OM) (Olympus Corporation, Japan) was used to observe the low-magnification microstructures of SLM-processed specimens. High-resolution study of the ultrafine nanostructures was performed using a S-4800 field emission SEM (FE-SEM) (Hitachi, Japan) at 5 kV. Chemical compositions were determined by an EDAX energy dispersive x-ray spectroscopy (EDX) (EDAX, Inc., USA). The Archimedes principle was used to measure the density ($\rho$) of the specimens. The Vickers hardness was measured using a MicroMet 5101 microhardness tester (Buehler GmbH, Germany) at a load of 0.2 kg and an indentation time of 20 s. The tribological property of specimens was estimated by the dry sliding wear tests conducted in a HT-500 ball-on-disk tribometer (Lanzhou ZhongKe KaiHua Sci. &Technol. Co., Ltd., China) in air at room temperature. The counter material was GCr15 bearing steel ball with a diameter of 3 mm and a mean hardness of HRC60, using a test load of 3 N was applied. The friction unit was rotated at a speed of 560 rpm for 15 min, with the rotation radius of 2 mm. The coefficient of friction (COF) of the specimens was recorded during wear tests. The wear volume ($V$) was determined gravimetrically using

$$V = \frac{M_{\text{loss}}}{\rho},$$

(2)

where $M_{\text{loss}}$ was the weight loss of the specimens after wear tests. The wear rate ($\omega$) was calculated by

$$\omega = \frac{V}{WL},$$

(3)

where $W$ was the contact load and $L$ was the sliding distance.

III. RESULTS AND DISCUSSION

A. Phases identification

Typical XRD spectra of SLM-processed nanocomposite parts obtained within a wide 2$\theta$ range (20°–90°) are depicted in Fig. 2. The strong diffraction peaks corresponding to Al9Si (JCPDS Card No. 65-8554) and stoichiometric TiC (JCPDS Card No. 65-8805) were identified, revealing the general formation of TiC reinforced AMCs at all given LEPUL. Furthermore, the Mg2Si (JCPDS Card No. 35-0773) precipitates were formed by natural ageing, and the detected diffraction peaks were relatively weak. Although the SLM process was performed in an inert argon atmosphere, the atmosphere may still contain 0.1%–0.3% residual oxygen. Consequently, the XRD spectra showed that the Al2O3 oxide was still formed at the lowest LEPUL of 250 J/m. Nevertheless, the Al2O3 diffraction peaks disappeared gradually when the applied LEPUL increased from 500 to 1000 J/m. Based on the research of Louvis et al. on the oxidation phenomenon of aluminum parts during SLM process, it revealed that the oxide films tended to be formed both on the top of the molten pool and on the surface of the solidified region of the track. Normally, the temperature of the upper molten surface under the laser beam was high enough to vaporize the oxide film on the top of the molten pool. However, the oxide films on the surface of the solidified region in the deposited track could not be easily broken if the insufficient laser power was applied. In the present study, the XRD characterization results revealed that as the applied

\[\text{FIG. 2. XRD spectra of SLM-processed TiC/AlSi10Mg parts at different LEPUL.}\]
LEPUL was higher than 500 J/m, the input laser energy was high enough to disrupt the oxide films efficiently, since the Al₂O₃ diffraction peaks vanished from the XRD spectra.

B. Densification behavior

The influence of laser energy input on the cross-sectional microstructures and densification behavior of SLM-processed TiC/AlSi10Mg nanocomposite parts is given in Fig. 3. Upon etching, the layerwise microstructure features became visible, due to the additive manufacturing nature of SLM. At a relatively low LEPUL of 250 J/m, the cross-section of SLM parts showed a relatively heterogeneous layerwise microstructure with the formation of the irregular shaped interlayer pores on a scale of several tens of micrometers [Fig. 3(a)]. The corresponding densification rate was about 92.5% theoretical density (TD) [Fig. 3(e)]. FE-SEM characterization revealed that there were a large amount of metallic balls with the diameters of several tens of micrometers formed on the SLM-processed surface [Fig. 3(f)]. It was accordingly concluded that the balling effect, which was a typical metallurgical defect associated with powder bed based SLM process, initiated in this instance. As the LEPUL increased to 500 J/m, although a small amount of interlayer pores were still present on the cross-section, the size of residual pores became considerably small [Fig. 3(b)], leading to an apparent increase of the relative density to 95.3% TD [Fig. 3(e)]. When the LEPUL further increased above 700 J/m, the cross-section showed a homogeneous microstructure with the evenly distributed layers, showing the coherent interlayer bonding ability free of any residual pores [Figs. 3(c) and 3(d)]. In these situations, the near fully dense (>98% TD) TiC/AlSi10Mg parts were generally produced after SLM.

During SLM, the TiC/AlSi10Mg nanocomposite powder is melted line-by-line by the laser beam, forming a mobile molten pool with a continuous liquid front. The amount of the liquid formed in the molten pool influences the densification and the resultant microstructure of the SLM-processed parts by changing the thermokinetic and thermocapillary characteristics, such as viscosity, wettability, and liquid–solid rheological properties. The dynamic viscosity (μ) of the melt in the pool consisting of the Al–Si–Mg liquid and the TiC reinforcing particles can be defined by 27:
where $\mu_0$ is the base viscosity that includes temperature terms, $\phi_l$ is the volume fraction of liquid phase, and $\phi_m$ is a critical volume fraction of solids above which the mixture has essentially infinite viscosity. In SLM process, the densification activity of the melt is controlled by the base viscosity ($\mu_0$), which is a strong temperature-dependent viscosity. This viscosity decreases with increasing the operative SLM temperature,\(^{28}\) thereby leading to the better wetting characteristics and improved densification response.

For the TiC/AlSi10Mg system being investigated, the presence of TiC reinforcing particles in aluminum alloy liquid tends to increase the viscosity of the melt significantly, hence hindering the sufficient flow of the melt and decreasing the overall rheological performance of the composite melt. Meanwhile, using a lower LEPUL results in a limited SLM temperature and, accordingly, elevates the base viscosity ($\mu_0$) of the melt. The combined influence of the material nature and the insufficient laser energy input below 500 J/m tends to reduce the wettability of the melt within the pool, hence decreasing the densification response after solidification of the melt. Furthermore, during high-energy laser scanning, a steep temperature gradient is developed between the center and edge of the pool across the surface. The temperature gradients in the pool give rise to surface tension gradients and resultant Marangoni convention.\(^{29}\) Because of the combined effect of Marangoni flow and the poor flowability of the melt caused by the high viscosity, the melt tends to flow radially inward toward the melt pool center, instead of spreading outward on the underlying surface.\(^{30}\) Consequently, the instable melt track breaks up into several spherical agglomerates to achieve the equilibrium state, which is termed as “balling” effect. During layer-by-layer SLM process, balling effect is detrimental to the uniform deposition of the fresh powder on the previously processed layer. As the laser beam scans over such an uneven powder layer, especially at a high scan speed (i.e., a low LEPUL), the melting/solidification front of the moving molten pool undergoes a significant disturbance and even interruption, causing the residual pores between the uneven SLM layers and the limited densification response after solidification [Figs. 3(a) and 3(b)]. It is accordingly reasonable to conclude that the high melt viscosity and limited wetting characteristics caused by an insufficient laser energy input are the key factors in producing balling effect and interlayer pores, hence weakening the densification activity of SLM-processed TiC/AlSi10Mg nanocomposite parts.

C. Microstructural evolution

Figures 4 and 5 illustrate the typical microstructures of SLM-processed TiC/AlSi10Mg nanocomposite parts. The quantitative determination of the corresponding average particle size of reinforcement is depicted in Fig. 6. The quantitative EDX elemental analysis elucidated that the chemical compositions of reinforcement formed at various LEPUL were Ti and C elements with the atomic ratio very close to 1:1. Combined with XRD results (Fig. 2), it was reasonable to consider that the reinforcement in SLM-processed nanocomposites was stoichiometric TiC. Although the TiC reinforcement had similar microstructure to the starting particulate morphology before SLM, its dispersion state and particle size were significantly influenced by the applied LEPUL. At a relatively low LEPUL of 250 J/m, the ultrafine TiC reinforcing particles with a mean size of 75 nm were formed [Figs. 5(a) and 6]. However, the nanostructured TiC

![FIG. 4. FE-SEM showing dispersion states of TiC reinforcement in SLM-processed TiC/AlSi10Mg parts at various LEPUL: (a) 250 J/m; (b) 500 J/m; (c) 700 J/m; (d) 1000 J/m.](image-url)
reinforcement showed a high tendency to aggregate into clusters, resulting in the formation of micron-sized agglomerates which consisted of several nanoparticles, as selectively indicated in Fig. 4(a). As the applied LEPUL increased to 500 J/m, the low-magnification FE-SEM characterization revealed that the distribution state of TiC reinforcing particles was improved [Fig. 4(b)]. Nevertheless, the high-magnification analysis showed the presence of some inhomogeneous agglomerates of the nanoscale TiC reinforcement in the matrix [Fig. 5(b)]. The average particle size of TiC reinforcement slightly increased to 78 nm in this instance (Fig. 6). With a further increase in LEPUL to 700 J/m, a uniform distribution of the TiC reinforcement throughout the matrix was obtained [Fig. 4(c)]. Even observed in a high magnification, there was no apparent aggregation of TiC nanoparticles in a microscopic scale within SLM-processed structure [Fig. 5(c)]. The average size of TiC reinforcing particles, 92 nm, was well within the nanometer scale (Fig. 6). At an even higher LEPUL of 1000 J/m, although the uniform distribution state of TiC reinforcement was still maintained [Fig. 4(d)], the mean particle size of TiC reinforcement increased markedly to 161 nm [Figs. 5(d) and 6], which indicated that the TiC reinforcement lost its favorable nanostructure after SLM process. It was accordingly concluded that the microstructural development of the nanoscale TiC reinforcement during SLM was sensitive to the laser energy input.

Typically, laser melting offers a high heating/cooling rate, contributing to the development of the nanostructured TiC reinforcement by shortening growth time for TiC grains. In the present study, it is found that a reasonable increase in the applied LEPUL from 250 to 700 J/m can homogenize the dispersion state of the nanoscale TiC reinforcing particles; however, an excessive laser energy input at 1000 J/m resulted in the significant coarsening of TiC grains and attendant disappearance of the nanostructure of the reinforcement. On one hand, the formation of the Marangoni convention within the molten pool tends to induce the liquid capillary force, which in turn acts on the TiC reinforcing particles and promotes their rearrangement in the pool. Based on the results of Arafune and Hirata, the dimensionless Marangoni number \((M_a)\) can be used to evaluate the intensity of Marangoni flow

\[
M_a = \frac{\Delta \sigma L}{\mu l v_k},
\]

where \(\Delta \sigma\) is the surface tension difference of Marangoni flow, \(L\) is the length of the free surface, \(\mu\) is the dynamic
viscosity, and $\nu_k$ is the kinematic viscosity. As a relatively high LEPUL is applied, the significantly enhanced SLM temperature causes a considerable decrease in $\mu$, thereby intensifying the Marangoni flow in the molten pool [Eq. (5)] and the attendant magnitude of capillary force. Consequently, the rearrangement rate of the nanoscale TiC reinforcing particles in the melt is elevated, thereby avoiding the aggregation of TiC nanoparticles and homogenizing their dispersion state throughout the finally solidified matrix [Figs. 4(c) and 5(c)]. On the other hand, the laser energy input influences the microstructural evolution of the nanoscale TiC reinforcement by changing the operative undercooling degree and attendant solidification rate. At a relatively high scan speed $v$ and resultant low LEPUL, the solidification rate of the liquid front within the molten pool increases, thereby enhancing the temperature gradients within the pool and the attendant undercooling degree of the melt. The grain growth of nanoscale TiC reinforcement is inhibited effectively, due to the insufficient time supplied for grain coarsening, hence remaining the favorable nanostructured TiC reinforcement after SLM. With the increase of the applied LEPUL, the thermalization of laser energy is elevated, resulting in a significant thermal accumulation and a higher SLM temperature within the molten pool. In this situation, the effect of quenching that occurs by conduction of heat through the substrate is not significant. The activity of the grains within the nanoscale TiC reinforcing particles becomes significantly elevated in the molten pool having a slower cooling rate and, accordingly, the TiC grains tend to experience a rapid growth. As a result, the TiC grains become coarsened continuously as the applied LEPUL increases above 1000 J/m and eventually the TiC reinforcing particles lose the initial standard nanostructure after SLM [Figs. 5(d) and 6].

D. Hardness and wear performance

Figure 7 depicts the microhardness and its distribution measured on the cross-sections of SLM-processed TiC/AlSi10Mg nanocomposite parts. With the increase of the applied LEPUL from 250 to 700 J/m, the average microhardness experienced a continuous increase from 157.3 HV$_{0.2}$ to 177.6 HV$_{0.2}$; however, as the LEPUL further increased to 1000 J/m, the mean microhardness decreased to 166.7 HV$_{0.2}$. Furthermore, the microhardness of the part processed at a low LEPUL of 250 J/m fluctuated significantly with the change of the measurement positions. This was mainly contributed to the inhomogeneous distribution of TiC reinforcement throughout the matrix at a low laser energy input [Fig. 4(a)]. This situation was considerably improved by increasing the LEPUL to 700 J/m, due to the homogeneous incorporation of nanoscale TiC reinforcing particles throughout the matrix. However, a further enhancement of the applied LEPUL to 1000 J/m did not lead to a further elevation of the obtainable microhardness, which indicated that the significant grain coarsening and the resultant disappearance of nanostructured TiC reinforcement [Figs. 5(d) and 6] tended to lower the hardness performance. Nevertheless, a close comparison revealed that the SLM-processed AlSi10Mg parts reinforced with nanoscale TiC particles, for all the given SLM parameters, demonstrated the superior hardness compared to the unreinforced AlSi10Mg parts fabricated with the same SLM conditions (maximum at approx. 145 HV$_{0.1}$).

Figure 8 shows the influence of SLM parameters on COFs and wear rates of SLM-processed TiC/AlSi10Mg nanocomposite parts. The FE-SEM studies of the
corresponding worn surfaces are provided in Fig. 9. At the initial stage of the wear tests, due to the existence of hard oxide film on the surface of samples, the COF values were relatively high and also showed an apparent fluctuation. The oxide film could be broken easily and shifted by the direct sliding with the counterface after which the COF values decreased significantly and reached a relatively steady-state with little local fluctuation [Fig. 8(a)]. Nevertheless, the large differences in COF values and attendant wear rates among various parts indicated that the wear performance of SLM-processed parts was significantly influenced by the applied laser processing conditions. At a relatively low LEPUL of 250 J/m, the average COF reached a high value of 0.63 [Fig. 8(a)], resulting in a considerably elevated wear rate of $7.46 \times 10^{-5} \text{ mm}^3 \text{ N}^{-1} \text{ m}^{-1}$ [Fig. 8(b)]. The wear surface was disruptive and consisted of large-sized agglomerates which contained some ultrafine nanoscale particles [Fig. 9(a)]. The presence of irregular shaped fragments revealed the local severe deformation and delamination of the worn surface occurred. On increasing the LEPUL to 500 J/m, although the abrasive fragments still existed on the worn surface, there were no apparent agglomerates of reinforcing nanoparticles in the worn matrix [Fig. 9(b)], thereby decreasing the mean COF and attendant wear rate to 0.57 [Fig. 8(a)] and $5.89 \times 10^{-5} \text{ mm}^3 \text{ N}^{-1} \text{ m}^{-1}$ [Fig. 8(b)], respectively. Interestingly, at a reasonable LEPUL of 700 J/m, the worn surface of SLM-processed sample became rather smooth and the reinforcing nanoparticles were found to be inserted within the adhesive tribolayer [Fig. 9(c)]. The lowest COF of 0.38 [Fig. 8(a)] and wear rate of $2.76 \times 10^{-5} \text{ mm}^3 \text{ N}^{-1} \text{ m}^{-1}$ [Fig. 8(b)] were obtained in this instance. It was accordingly reasonable to consider that as the nanostructured TiC reinforcement with a homogenous dispersion state was formed in SLM-processed parts, the mechanism of material removal during sliding was changed from the abrasion to adhesion of the tribolayer. Such a transition was expected to homogenize the COF values and reduce the wear rate after sliding tests. However, at an even higher LEPUL of 1000 J/m, the parallel, deep grooves were observed on the worn surface [Fig. 9(d)], which revealed the severe deformation and plowing of the surface during sliding. In this situation, the wear mechanism was microploughing rather than the adhesive wear. Some relatively large-sized reinforcing particles were also observed at the edges of grooves on the worn surface. The obtained average COF value of 0.54 [Fig. 8(a)] and wear rate of $5.83 \times 10^{-5} \text{ mm}^3 \text{ N}^{-1} \text{ m}^{-1}$ [Fig. 8(b)] showed a significant increase in this case.

In order to disclose the underlying mechanisms contributed to the wear performance enhancement, a comprehensive relationship of SLM process, microstructures, and mechanical properties is determined. At the relatively low laser energy input, the SLM densification response is limited due to the pore formation [Figs. 3(a) and 3(e)], and meanwhile, the TiC nanoparticles aggregate into clusters within the laser processed matrix, resulting in a poor microstructural homogeneity. The pores and nanoparticle aggregates are prone to crack because of the concentration of stress during sliding. The debris is thus formed and split from the wear surface when the expanded cracks are connected to each other [Fig. 9(a)], resulting in a significant increase in the COF and wear rate. When the proper LEPUL is applied, the nanostructured TiC reinforcement with a uniform distribution is formed throughout the matrix [Figs. 4(c) and 5(c)]. During sliding tests, the homogeneously dispersed ultrafine TiC nanophase has a high tendency to stick to each other, due to considerably small interfacial stresses of nanoscale reinforcement with the matrix.32 As the sliding friction acts on the surface, the worn surface experiences a sufficient plastic
deformation and, therefore, a strain-hardened sliding-treated layer consisting of the significantly refined TiC reinforcing nanoparticles is produced on the wear surface [Fig. 9(e)], which favors the improvement in wear property. However, at a relatively high LEPUL, although the sufficient densification level is maintained [Figs. 3(d) and 3(e)], the TiC reinforcement loses its nanostructure after SLM process [Fig. 5(d)]. The coarsened TiC reinforcing particles are prone to split and pull out from the matrix during sliding. The plowing between the segregated TiC reinforcement and the matrix tends to occur with the action of the tangential force, thereby producing the deep grooves on the worn surface with a limited wear performance [Fig. 9(d)].

IV. CONCLUSIONS

(1) The densification behavior of TiC/AlSi10Mg nanocomposite parts was significantly affected by the laser energy input during SLM. Using an insufficient LEPUL of 250 J/m lowered the SLM densification rate due to the formation of residual pores. The highest densification level (>98% theoretical density) was achieved for SLM-processed nanocomposite parts processed at the LEPUL of 700 J/m.

(2) The TiC reinforcement in SLM-processed TiC/AlSi10Mg nanocomposite parts experienced a structural change from the standard nanoscale particle morphology (the average size of 75–92 nm) to the relatively coarsened submicron structure (the mean particle size of 161 nm) as the LEPUL increased. The nanostructure of TiC reinforcement was successfully maintained within a wide range of LEPUL from 250 to 700 J/m. The dispersion state of nanoscale TiC reinforcement was homogenized from the apparent agglomerates to the uniform distribution with the increase of LEPUL to 700 J/m.

(3) The densification rate of SLM-processed TiC/AlSi10Mg nanocomposite parts and the particle morphology and dispersion state of TiC reinforcement in the matrix played the key roles in determining the wear property. The sufficiently high densification rate combined with the uniform distribution of nanoscale TiC reinforcement throughout the matrix LEPUL to the considerably low COF of 0.38 and wear rate of 2.76 × 10^{-5} \text{mm}^3 \text{N}^{-1} \text{m}^{-1} for SLM-processed nanocomposites at LEPUL of 700 J/m. The insufficient densification of SLM parts at a relatively low LEPUL of 250 J/m and the disappearance of nanoscale reinforcement at a high LEPUL of 1000 J/m generally lowered the wear performance of TiC/AlSi10Mg nanocomposite parts.

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