Selective Laser Melting Additive Manufacturing of Novel Aluminum Based Composites With Multiple Reinforcing Phases

The selective laser melting (SLM), due to its unique additive manufacturing (AM) processing manner and laser-induced nonequilibrium rapid melting/solidification mechanism, has a promising potential in developing new metallic materials with tailored performance. In this work, SLM of the SiCp/AlSi10Mg composites was performed to prepare the Al-based composites with the multiple reinforcing phases. The influence of the SLM processing parameters on the constitutional phases, microstructural features, and mechanical performance (e.g., densification, microhardness, and wear property) of the SLM-processed Al-based composites was studied. The reinforcing phases in the SLM-processed Al-based composites included the unmelted micron-sized SiC particles, the in situ formed micron-sized Al4SiC4 strips, and the in situ produced submicron Al4SiC4 particles. As the input laser energy density increased, the extent of the in situ reaction between the SiC particles and the Al matrix increased, resulting in the larger degree of the formation of Al4SiC4 reinforcement. The densification rate of the SLM-processed Al-based composite parts increased as the applied laser energy density increased. The sufficiently high density (~96% theoretical density (TD)) was achieved for the laser linear energy density larger than 1000 J/m. Due to the generation of the multiple reinforcing phases, the elevated mechanical properties were obtained for the SLM-processed Al-based composites. The SLM-processed Al-based composites, showing a high microhardness of 214 HV0.1, a considerably low coefficient of friction (COF) of 0.39, and a reduced wear rate of 1.56 × 10−3 mm3 N−1 m−1. At an excessive laser energy input, the grain size of the in situ formed Al4SiC4 reinforcement phase, both the strip- and particle-structured Al4SiC4, increased markedly. The significant grain coarsening and the formation of the interfacial microscopic shrinkage porosity lowered the mechanical properties of the SLM-processed Al-based composites. These findings in the present work are applicable and/or transferrable to other laser-based powder processing processes, e.g., laser cladding, laser metal deposition, or laser engineered net shaping. [DOI: 10.1115/1.4028925]

1 Introduction

SLM is a newly developed AM technology [1–5], which enables the quick production of the three-dimensional parts with the nearly full density without the postprocessing treatments [6–10]. SLM and direct metal laser sintering (DMLS) are the typical powder-bed-based laser AM processes. Generally DMLS is processed based on a liquid phase sintering mechanism involving a partial melting of the powder, i.e., a semisolid consolidation mechanism [11]. In contrast, SLM of metallic powder is through a complete melting/solidification mechanism. The idea of full melting is supported by the availability of fine metal powder specially designed for SLM and the significantly elevated laser processing conditions in recent years, e.g., higher laser power, smaller focused spot size, thinner layer thickness, etc [12]. The SLM technology can be used to manufacture the complicated parts with any geometry by layer-by-layer melting and consolidating of powder materials, based on the computer-aided design (CAD) data of the objects to be built [13–16]. The SLM process includes scattering and absorption of laser radiation in the powder, heat conduction, melting and coalescence of powder particles, formation of the melt pool, and its solidification [17,18]. During the SLM process, the interaction time between the laser beam and the powder particles is extremely short (typically between 0.5 and 25 ms) [19], resulting in the considerably high heating and cooling speed [20]. During the laser-induced rapid solidification process, a number of nonequilibrium metallurgical phenomena may occur, such as the microstructural refinement, the solid solution hardening, and the formation of metastable phases, favoring the improvement of the properties of the laser-processed materials [21–23]. Due to the laser beam with the high energy density applied during SLM process, the working temperature is high and the powder materials tend to melt completely [6,24,25]. The melted materials undergo a typical melting/solidification metallurgical process, leading to a sufficiently high densification behavior and attendant mechanical performance of the SLM-processed materials. The SLM technology offers the remarkable degree of the processing freedom over the traditional processing methods, thereby improving the process ability of the components with the complex shapes that are hard to be processed using the conventional methods [15,26–28]. Currently, a number of powder materials, such as stainless steel and titanium, have been successfully utilized in the SLM process [29].

For the last few decades, the aluminum alloys have been widely used in many industrial fields such as aerospace, architectural construction, and automotive, because of the nature of lightweight,
high corrosion resistance, high thermal conductivity, etc [13–15]. In order to meet the demand of the more comprehensive mechanical properties (e.g., the enhanced tensile strength and wear/tribological performance), the development of the particle reinforced aluminum matrix composites (AMCs), especially the SiC particle reinforced AMCs, attracts the considerable attention, exhibiting the inspiring prospects for development [30]. The SiC particle reinforced AMCs have the significant potential for the applications in the aircraft and automotive industries, because of the characteristics of lightweight, high strength, high elastic modulus, good wear resistance, and low coefficient of thermal expansion [31–33]. Currently, the parts manufactured from the SiC particle reinforced AMCs have been successfully applied in both military and civilian fields, such as aircraft ventral fin [34], wing panels [31], pistons and engine blocks [35]. Powder metallurgy (PM) and casting are the commonly used conventional processing methods to produce the SiC particle reinforced AMCs parts and the obtainable mechanical performance can mainly meet the requirements for the practical applications [32,36]. However, there are some significant drawbacks that severely hinder the further development of the AMCs by the conventional methods, such as the expensive PM/casting molds and the complex preprocessing and postprocessing treatments [15]. Furthermore, due to the limited wettability between the ceramic reinforcing phase and the aluminum matrix, the limited particle/matrix interfacial bonding ability that results from the low working temperature during the conventional process may further decrease the mechanical performance of the final products. Therefore, the development of the new processing method to produce SiC particle reinforced AMCs parts is driven by the desire to shorten the processing cycle and to manufacture parts with excellent performance. Thanks to the advances in the AM technology, the application of the SLM process in the production of the SiC particle reinforced AMCs is a promising choice. Nevertheless, the intrinsic nature of the aluminum, e.g., the high laser reflectivity, the high heat conductivity, and the elevated affinity to oxygen, is the detrimental factor that may lower the properties of AMCs manufactured by the SLM process [37,38]. Some previous researches have already been carried out regarding laser processing of AMCs. Simchi and Godlinski prepared the SiC/Al–7Si–0.3 Mg composite parts [37] and the SiC/AlSi56 composite parts [38] using the DMLS AM technology. Anandkumar et al. produced the SiC/Al–Si composite coatings on the casting aluminum substrate by laser cladding technology [39]. According to the previous studies, during laser processing, the aluminum tends to react with the SiC and the temperature is regarded as the key factor influencing the reaction procedure. In the temperature range from 940 K to 1620 K, the SiC reacts with aluminum to form Al4C3 and the reaction procedure is [39]:

\[
4\text{Al}(l) + 3\text{SiC}(s) \rightarrow \text{Al4C3}(s) + 3\text{Si}
\]  

When the temperature reaches above 1670 K, the reaction product is Al4SiC4 and the reaction procedure is as follows [37,39]:

\[
4\text{Al}(l) + 4\text{SiC}(s) \rightarrow \text{Al4SiC4}(s) + 3\text{Si}
\]  

It is noted that the Al4SiC4 (1200 HV) is much harder than the Al4C3 (250 HV), which is regarded as the favorable reinforcing phase for aluminum matrix. The Al4C3 is brittle and has a tendency to react with H2O in wet environments to form the amorphous aluminum hydroxide. Hydration of the aluminum carbide is achievable in laser-processed AMCs. From this point of view, setting the elevated manufacturing temperature above 1670 K may be a useful method to achieve the proper reinforcement in laser-processed AMCs.

In the present work, the novel AMCs with multiple reinforcing phases were prepared by SLM processing of the SiC/AlSi10Mg composite. The microstructure characteristics and phase development were studied for the SLM-processed AMCs using different processing parameters. The SLM densification mechanism of AMCs was analyzed and the mechanical properties, e.g., microhardness and wear resistance, were investigated, with a major target to develop new Al-based composite materials and to optimize the SLM process and resultant mechanical performance. The SLM-processed Al-based composites reinforced by in situ Al4SiC4 exhibited novel reinforcing structures and resultant elevated hardness/wear performance, which, to our best knowledge, have not been reported previously in the conventionally processed or laser additive manufactured Al-based composites.

2 Experimental Procedures

2.1 Powder Materials. The raw powder materials applied in this study consisted of the 99.7% purity AlSi10Mg powders with a spherical shape and a mean particle size of 30μm (Fig. 1(a)), and the 99.5% purity SiC powder with a polyangular structure and an average particle size of 7μm (Fig. 1(b)). The SiC and AlSi10Mg powder components, according to the weight ratio of 20:80, were mechanically mixed in a Fritsch Pulverisette 4 vario-planetary mill, using a ball-to-powder weight ratio of 1:1, a rotation speed of the main disk of 200 rpm, and a mixing time of 4 h. Since a high flowability of the powder is necessary for its smooth spreading during SLM process, the ball milling parameters in this work have been carefully determined to maintain the initial spherical structure of the AlSi10Mg matrix powder and, meanwhile, to realize the uniform dispersion of the SiC reinforcing particles in the powder mixture. The morphology and dispersion state of the mixed powder is revealed in Fig. 1(c). The spreading property of the mixed powder on powder bed was sufficiently favorable during SLM process.

2.2 SLM Process. The SLM process system used in this study was self-developed and mainly consisted of an IPG YLR-200-SM ytterbium fiber laser with a power of ~200 W and a spot size of 70μm, a SCANLAB hurrySCAN 30 scanner, an automatic powder spreading system, an inert argon gas protection system, and a computer software system for process control, as schematically depicted in Fig. 2(a). As the specimens were to be built, an aluminum substrate was fixed on the building platform and leveld. The building chamber was then sealed and the argon gas was fed inside, decreasing the O2 content below 20 ppm. Afterward, the SiC/AlSi10Mg composite powder was deposited on the substrate by the layering mechanism, with the powder layer thickness of 50μm. The laser beam was then controlled by the CAD data to scan the powder bed surface selectively, thereby producing a two-dimensional profile. A laser scan strategy of the alternate X–Y directions was applied and the hatch spacing between the neighboring scan lines was 50μm, as shown in Fig. 2(b). A series of SLM experiments using a wide range of processing parameters, i.e., laser powers of 70–190 W and scan speeds of 50–600 mm/s, were performed to testify the SLM processability. Among these parameters, we chose a fixed laser scan speed (v) of 100 mm/s and four various laser powers (P) of 80 W, 90 W, 100 W, and 110 W in the present study, in order to vary the SLM processing conditions. The laser linear energy density (η), which was defined by [41]

\[
\eta = \frac{P}{v}
\]

was used to describe the laser energy delivered into the powder layer in the unit length along the laser scanning direction, as revealed in Fig. 2(c). Four different η of 800 J/m, 900 J/m, 1000 J/m, and 1100 J/m were accordingly changed for SLM experimentation. The cubic specimens with dimensions of
6 mm × 6 mm × 6 mm were fabricated in a layer-by-layer manner until completion.

2.3 Microstructure Characterization and Mechanical Properties Tests. After the specimens were cut from the substrate, the samples were ground and polished according to the standard procedures for the metallographic specimen preparation and examination. The samples were successively ground with finer and finer SiC paper as the abrasive media. After grinding the specimen, polishing was performed. The woven polishing cloths were employed to produce a scratch-free mirror finish, free from smearing or pull-outs remaining from the metallographic preparation process.

The density (ρ) of the SLM-processed specimens was defined as mass divided by volume. The mass of the specimens was measured by the analytical balance and the volume of the specimens was determined using the Archimedes’ method. Archimedes’ principle shows that the apparent weight of an object immersed in a liquid decreases by an amount equal to the weight of the volume of the liquid that it displaces. Since 1 ml of water has a mass almost exactly equal to 1 g, if the object is immersed in water, the difference between the two masses (in grams) will equal (almost exactly) the volume (in ml) of the object weighed [42]. The experimental determination of the mass and the volume of an object allows us to calculate the density.

The densification behavior and microstructures of the samples were further observed using an Olympus PMG3 optical microscope. The phases of the specimens were identified by a Bruker D8 Advance X-ray diffractometer (XRD) with Cu Kα radiation at 40 kV and 40 mA, using a continuous scan mode at 4 deg/min. The microstructural features were characterized using a Hitachi S-4800 field emission scanning electron microscope (FE-SEM), after the samples were etched by a solution consisting of HF (1.0 ml), HCl (1.5 ml), HNO₃ (2.5 ml), and distilled water (95 ml). The chemical compositions were determined by an EDAX energy dispersive X-ray spectroscopy (EDX), using a super-ultrathin window sapphire detector.

The Vicker hardness of the samples was measured using a Buehler MicroMet 5101 microhardness tester at a load of 100 g and an indentation time of 20 s. Based on the ASTM G99 standard, the wear/tribological properties of the specimens were 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 with a test load of 230 g. The schematic of the working mechanism of the tribometer is illustrated in Fig. 2(d). The counterface material was GCr15 bearing steel ball with a diameter of 3 mm and a mean hardness of HRC 60. The friction unit was rotated at a speed of 560 rpm for 20 min, with the rotation radius of 2 mm. The COF of the specimens was recorded during wear tests. The wear volume (V) of specimens was calculated by \( \frac{M_{\text{loss}}}{\rho} \), where \( M_{\text{loss}} \) was the weight loss of the sample after wear test and \( \rho \) was the density of the sample. The wear rate (\( \omega \)) was identified by \( \omega = \frac{V}{WL} \), where \( W \) was the contact load applied in the test and \( L \) was the sliding distance.

3 Results and Discussion

3.1 Phases Identification. Figure 3 depicts the typical XRD spectra of the SLM-processed Al-based composite parts at the different laser linear energy densities (\( \eta \)). The strong diffraction peaks corresponding to α-Al were identified. The newly formed phase Al₆SiC₄ was also determined, which implied that the in situ reaction between aluminum melt and silicon carbide occurred at the temperature larger than 1670 K, as disclosed in Eq. (2). The characteristic peaks of the residual SiC phase and the new Mg₂Si phase were also detected. Due to the relatively high mass fraction of the SiC (20 wt.%) in the starting powder system, the existence of the SiC phase after SLM was reasonable, because of the incomplete in situ reaction during the laser-induced rapid melting/solidification process. The presence of the Mg₂Si phase indicated the formation of magnesium silicide by means of the solid solution of the silicon atoms into the magnesium lattice. It was worth noting that although the aluminum had a high affinity to oxygen, the XRD results showed that the characteristic peaks of aluminum oxide were not detected in the SLM-processed Al-based composites, due to the efficient protection of the argon atmosphere during SLM.
3.2 Microstructures and Compositions. Figures 4 and 5 illustrate the microstructural characteristics of the SLM-processed Al-based composite parts at various laser linear energy densities (g), using the different FE-SEM magnifications. Typically, three different reinforcing structures existed in the SLM-processed composites under various SLM conditions, i.e., the relatively large-sized reinforcing particles, the strip-structured reinforcement, and the relatively small-sized reinforcing particles. In order to investigate the chemical compositions of these three kinds of reinforcements in the SLM-processed composites, EDX characterization within the different structures, as indicated in Figs. 4(a), 4(c), and 5(c), was performed, with the results provided in Fig. 6. As shown in Fig. 6(a), EDX point scan within the large-sized reinforcing particles (point 1, Fig. 4(a)) revealed that the C and Si had a near equal atomic proportion. In this instance, the presence of a small amount of Al element (12.51 at.%) was caused by the attachment of the matrix element on the reinforcement during the grinding process of the metallographic samples. It was accordingly reasonable to consider that the first kind of reinforcement was the residual SiC particles, which typically exhibited the relatively large particle morphology. Furthermore, as revealed in Figs. 6(b) and 6(c), the presence of the Al, Si, and C elements with an approximate Al:Si:C atom ratio of 4:1:4 was detected within both the strip-structured reinforcement (point 2, Fig. 4(c)) and the small-sized reinforcing particles (point 3, Fig. 5(c)). Combined with the XRD results, it was concluded that the Al$_4$SiC$_4$ reinforcing phase was in situ formed during SLM via the reaction between aluminum and SiC. The in situ presented Al$_4$SiC$_4$ typically had two different morphologies, i.e., the relatively large-scale strip-structure and the ultrafine particle morphology. Generally speaking, the SLM processing of the present SiC/AlSi10Mg composite system led to the formation of three different reinforcements, i.e., the unmelted micron-sized SiC particles, the in situ formed strip-structured Al$_4$SiC$_4$, and the in situ produced ultrafine Al$_4$SiC$_4$ particles.

It was clear that the microstructures and growth mechanisms of the reinforcements were significantly influenced by the applied laser linear energy densities (g). At a relatively low g of 800 J/m, the irregular shaped SiC reinforcing particles with an average size of 4 µm were remained in the SLM-processed composites and
there was no apparent formation of the strip-structured $\text{Al}_4\text{SiC}_4$ reinforcing phase (Fig. 4(a)). In this instance, the in situ formed $\text{Al}_4\text{SiC}_4$ reinforcement was present in the near spherical structure within the Al matrix, with the considerably refined particle size of $\sim 350$ nm (Fig. 5(a)). As the applied $\eta$ increased to $900$ J/m, the volume fraction of the residual SiC reinforcing particles decreased and the particle shape became more regular and smooth. Furthermore, a small amount of the strip-structured $\text{Al}_4\text{SiC}_4$ reinforcing
During the SLM process, the laser energy is delivered into the powder layer via the interaction between the laser beam and the powder materials, within an extremely short duration of action. The laser energy is absorbed by the particles through both bulk coupling and powder coupling mechanisms [43]. Although the intrinsic absorptivity of aluminum to the laser is considerably low, the addition of 20 wt.% SiC with high laser absorptivity is favorable to offset this disadvantage. When the high-energy laser beam irradiates and heats up the powder particles, the temperature of the powder material enhances in an extremely short time. The aluminum with a low melting point of 933 K melts rapidly and the SiC particles having a significantly high melting point remain unmelted. Then, the molten pool consisting of the melted aluminum and the unmelted SiC particles is formed. The aluminum melt surrounds and wets the unmelted SiC particles due to the flow of the melt under the action of the Marangoni convection in the molten pool and resultant capillary forces of the liquid [44]. With the sufficient wetting of the liquid at the elevated temperature, the SiC particles tend to be melted partially, starting from the melting of the particle surface. Because of the in situ formation of the Al4SiC4 phase, the maximum SLM temperature in the molten pool exceeds 1670 K, as revealed in Eq. (2). At the elevated temperature, the Si and C atoms have the strong thermal motion ability and, consequently, they are easy to escape from the equilibrium position of the lattice or to be replaced by the Al atoms. Thus, the in situ reaction occurred by means of the diffusion of the Al, Si and C atoms, typically on the melted interface between the SiC and Al melt, hence generating the in situ Al4SiC4 reinforcing phase [40]. The Al4SiC4 initially forms as the faceted platelets on the interface between the solid and liquid [37], whereas the Si further dissolves into the liquid and forms the Mg2Si phase (Fig. 3). As revealed in Fig. 4, the laser linear energy density (η) has an important influence on the mechanism of the in situ formation of Al4SiC4 reinforcing phase and the attendant microstructural features of the SLM-processed Al-based composites. At a relatively low η of 800 J/mm, the energy absorbed by the powder is relatively low. In this situation, only a small amount of fine SiC particles or the edges of the irregular large-sized SiC particles...
tend to melt. The in situ reaction only performs partially around the surface of the melted SiC particles, leading to the formation of the particle structured Al4SiC4 reinforcement between the residual SiC particles (Fig. 5a). Because of the existence of a large amount of unmelted SiC particles in the molten pool, the viscosity of the melt is considerably high, resulting in a relatively low flowability of the melt [45,46]. Under this condition, the microstructural homogeneity of the SLM-processed Al-based composites, which consist of both large-scale SiC reinforcing particles and in situ formed Al4SiC4 reinforcing particles, is limited (Fig. 4a). When the applied η increases to 900 J/m, the laser energy that the molten pool is received increases, thereby enhancing the temperature of the pool. A larger degree of the melting of the SiC particles occurs and the activity of the in situ reaction between the SiC and Al increases, leading to the formation of both the strip-structured (Fig. 4b) and the particle-shaped (Fig. 5b) in situ Al4SiC4 reinforcement. The orientated Al4SiC4 strips are present on the melted surface of the SiC particles under the effect of the elevated SLM temperature and the significant liquid flow, by means of the heterogeneous nucleation of the Al4SiC4 nuclei and the subsequent grain growth. At a relatively higher η of 1000 J/m, the temperature of the molten pool increases significantly and a large fraction of the SiC particles decompose and react with the aluminum melt, producing a larger amount of the Al4SiC4 reinforcing phase. As the in situ reaction proceeds, the strip-structured Al4SiC4 phase grows continuously and, conversely, the SiC phase is exhausted gradually until it vanishes in the in situ reaction system. The in situ system accordingly enters a stable state and the in situ present gradient grows continuously and, conversely, the SiC phase is exhausted other kind of in situ Al4SiC4 phase, which is in the ultrafine particles morphology, can further behave as the reinforcement throughout the Al matrix to improve the mechanical properties of the SLM-processed composites.

3.3 Densification Behavior. The different relative densities and the cross-sectional microstructures of the SLM-processed specimens at the various laser linear energy densities (η) are provided in Fig. 7. It showed that the relative density of the SLM-processed composite specimens had the trend of increasing as the applied η increased. At a relatively low η of 800 J/m, the density of the specimen was only 89.2% of the TD of the material, and some voids with a relatively large average size of ~30 μm were observed on the cross section of the specimen. As the η increased to 900 J/m, the relative density of the sample increased to 93.4% and the amount of pores decreased, showing a reduction in the pore size. The SLM-processed Al-based composites with the considerably high relative density of 95.5% were produced at a high η of 1000 J/m, exhibiting a cross section with a small amount of voids having a very small size. On increasing the η to 1100 J/m, the relative density exhibited a slight increase to 96.1% and only few small pores existed on the cross section after SLM.

It is known that during the SLM process, the laser energy input has a significant impact on the viscosity of the melt, which further influences the melt flow within the pool. When the applied η is low, the laser energy input is limited and a large fraction of the unmelted SiC particles remain in the molten pool, hence increasing the viscosity of the melt and lowering the fluidity of the liquid in the molten pool. In this situation, the melt cannot efficiently wet the underlying previously processed layer and the surrounding solidified material in the currently processed layer, thereby producing the residual porosity between the neighboring scan tracks/layer and lowering the densification activity of the SLM-processed parts. At a relatively high η of 1000 J/m, the higher laser energy input significantly reduces the melt viscosity in the molten pool, due to the elevated SLM working temperature obtained, hence enhancing the wetting characteristics of the melt in the molten pool. Furthermore, according to Zhou and De Hosson’s work [48], the reaction between the metal matrix and the ceramic particles favors the improvement of their mutual wetting ability. As revealed in Figs. 4c and 5c, a large amount of the SiC particles are consumed and turned into the Al4SiC4 phase with the refined microstructures. As relative to the direct interfacial contact of the SiC and aluminum, the in situ chemical reaction tends to decrease the interfacial energy significantly and the change of the lattice structures on the interface becomes much tempered [48]. Therefore, the wetting properties, both the microscopic interfacial wetting and the macroscopic layer/layer wetting, tend to be improved, leading to an elevated densification response for the SLM-processed composite parts. As the applied η continues to increase to 1100 J/m, the relative density of the SLM-processed sample does not show any apparent decrease. Nevertheless, the apparent microscale shrinkage porosity is present...
played an enhancement from 196 HV 0.1 to 203 HV 0.1, accompanying a decrease in the fluctuation of the values. The formation of the shrinkage porosity between the coarsened phases/crystals, which is in a considerably microscopic scale, may not have a direct influence on the densification level. Nevertheless, it may exert some detrimental effects on other mechanical properties such as hardness and wear performance, which will be studied and presented in Sec. 3.4.

3.4 Hardness and Wear Performance. Figure 8 depicts the variations of microhardness of the SLM-processed Al-based composite specimens at the different processing parameters. It was clear that the microhardness and its distribution were influenced by the applied laser linear energy densities (η). With the increase of η from 800 J/m to 900 J/m, the mean microhardness values displayed an enhancement from 196 HV 0.1 to 203 HV 0.1, accompanied with a decrease in the fluctuation of the values. At a higher η of 1000 J/m, the average microhardness value further increased to 214 HV 0.1 and the fluctuation of the values was further weakened. However, when the applied η further increased to 1100 J/m, the microhardness suffered a reduction in the average value to 212 HV 0.1 and the fluctuation of the microhardness values became apparent. A close comparison revealed that for all the given SLM processing parameters, the SLM-processed Al-based composites reinforced with the multiple reinforcements demonstrated the superior hardness compared to the unreinforced AlSi10Mg parts fabricated using the same SLM conditions (the maximum microhardness of ~145 HV 0.1) [49], showing 26–32% enhancement upon the unreinforced AlSi10Mg alloy.

In general, due to the effect of the grain refinement strengthening in laser-processed materials, the SLM-processed parts typically exhibit the higher hardness than the parts manufactured by the conventional methods. Figure 8 reveals that the laser linear energy density (η) of 1000 J/m is a critical value corresponding to the different hardness performance. When the applied η is below 1000 J/m, the density of the SLM-processed Al-based composites is relatively low due to the presence of the residual interlayer pores, which decreases the obtainable microhardness values. Also, due to the insufficient in situ reaction between the SiC and aluminum, the microstructural homogeneity of the reinforcement in SLM-processed composites is limited, hence enhancing the fluctuation of microhardness distribution. The highest microhardness is obtained at the η of 1000 J/m, with the minimum fluctuation of hardness values. In this case, the densification level of the SLM-processed composites is sufficiently high (~96%), which acts as the underlying basis for the improvement of hardness. Furthermore, the multiple reinforcing structures are produced due to the in situ reaction during SLM process, having a positive influence on the hardness performance. The strip-structured Al2SiC4 reinforcing phase has a stronger ability to withstand the load than the matrix. Meanwhile, the ultrafine particle-shaped Al2SiC4 phase embeds throughout the Al matrix and plays an important role in hindering the dislocation movement, exhibiting the effect of the dispersion strengthening. Furthermore, during the laser-induced rapid solidification process, a large amount of Si atoms tend to diffuse into the lattice of the matrix metals, thereby working as the solid solution strengthening effect to further improve the microhardness [12]. However, the excessive increase of the η applied above 1000 J/m results in the decrease of the hardness, which is caused by the formation of the coarsened grains resulting from the significant grain growth (Figs. 4(d) and 5(d)).

Figure 9 depicts the variations of the COF values and resultant wear rates of the SLM-processed Al-based composite parts under the different SLM processing conditions. As the applied laser linear energy densities (η) increased from 800 J/m to 900 J/m, the mean COF values changed slightly from 0.45 to 0.47 and the attendant wear rates decreased from $4.27 \times 10^{-5}$ mm$^3$ N$^{-1}$ m$^{-1}$ to $3.38 \times 10^{-5}$ mm$^3$ N$^{-1}$ m$^{-1}$. As the applied η increased to
1000 J/m, the average COF value and the wear rate decreased markedly to 0.39 and 1.56 × 10⁻⁵ mm³ N⁻¹ m⁻¹, respectively, and the fluctuation of the COF values became insignificant, implying the excellent wear resistance of the SLM-processed Al-based composites in this instance. However, the wear performance of the SLM-processed Al-based composites at an even higher η of 1100 J/m showed a decreasing tendency; the average COF value increased to 0.52 with the elevated wear rate of 2.62 × 10⁻⁵ mm³ N⁻¹ m⁻¹.

Fig. 10 FE-SEM images showing the morphologies of the worn surfaces of the SLM-processed Al-based composite parts at the different laser linear energy densities (η): (a) η = 800 J/m; (b) η = 900 J/m; (c) η = 1000 J/m; (d) η = 1100 J/m

Fig. 11. High-magnification FE-SEM images showing the typical morphologies of the reinforcement on the worn surfaces of the SLM-processed Al-based composite parts at the different laser linear energy densities (η): (a) η = 800 J/m; (b) η = 900 J/m; (c) η = 1000 J/m; (d) η = 1100 J/m
In order to further study the wear mechanism at the different SLM processing parameters, the detailed FE-SEM studies of the corresponding worn surfaces are shown in Figs. 10 and 11, using the various magnifications for SEM images. At a relatively low \( \eta \) of 800 J/m, the rough, heterogeneous worn surface consisting of the parallel grooves and the aggregated, protruding particles was observed for Fig. 10(a). High-magnification FE-SEM image shows the presence of the particle-shaped fragments with the average size of \( \approx 500 \) nm at the edges of the grooves (Fig. 11(a)), which revealed the severe deformation and plowing of the surface during sliding. As the applied \( \eta \) increased to 900 J/m, the grooves on the worn surface became shallower (Fig. 10(b)) and the protruding particles became dispersed on the surface, showing an apparently decreased size of \( \approx 300 \) nm (Fig. 11(b)). The relatively smooth and dense worn surface was formed at the \( \eta \) of 1000 J/m, without the presence of the apparent protruding particles (Figs. 10(c) and 11(c)). At an excessive \( \eta \) of 1100 J/m, the worn surface suffered the apparent spalling, showing the broader parallel grooves on the surface (Fig. 10(d)). In this situation, the debris in the morphology of the aggregated small-sized particles appeared again on the worn surface (Fig. 11(d)).

During the wear tests, the counterpart ball slides against the surface continuously and the wear performance is primarily influenced by the microstructural features and hardness of the corresponding specimens [50,51]. When the applied \( \eta \) is low, some SiC particles with a high hardness remain unmelted due to the low SLM working temperature. During sliding, under the action of the strong compression effect from the load, the aluminum matrix with a relatively lower hardness performs a large deformation and, meanwhile, the residual SiC particles in the matrix tend to protrude from the worn surface under the shearing stress. Since these protruded SiC particles exist between the specimen surface and the sliding ball, the hard ceramic particles severely plow the specimen surface and, consequently, some deep grooves are formed on the worn surface (Fig. 10(a)). The mechanism of the abrasive wear is represented in this situation, resulting in the low wear resistance with the significantly high wear rate (Fig. 9). With the increase of the applied \( \eta \), the in situ reaction in the molten pool occurs more efficiently and the SiC particles are exhausted in this reaction system. Therefore, the plowing effect of the SiC particles on the testing surface decreases and the worn surface shows the shallower grooves with few protruding particles (Figs. 10(b) and 11(b)). Meanwhile, the in situ formed strip-structured and particle-shaped Al\(_4\)SiC\(_4\) with the considerably refined microstructures are generated in the SLM-processed composites, which have the excellent bonding strength with the matrix (Figs. 4(c) and 5(c)). As the worn surface experiences the sufficient plastic deformation at a temperature below its recrystallization temperature, the material strengthening occurs because of the dislocation movements within the crystal structure of the material, typically in the sliding-treated layer, which is known as strain-hardened tribolayer [52]. It is believed that the in situ formed homogeneous and refined Al\(_4\)SiC\(_4\) reinforcements are not easy to split during sliding but have a high tendency to adhere to each other and get strain-hardened, favoring the complete formation of tribolayer to enhance wear resistance (Figs. 9, 10(c), and 11(c)). The sufficiently high densification is active (Fig. 7) and the enhancement of hardness (Fig. 8) also contributes to the improvement of the wear performance of the SLM-processed composites in this case. It is thus reasonable to consider that the mechanism of material removal during sliding changes from the abrasion (Figs. 10(a) and 11(a)) to the adhesion of the tribolayer (Figs. 10(c) and 11(c)), thereby favoring the stabilization of COF values and the reduction in wear rate after sliding (Fig. 9). However, for the composites produced at an even higher \( \eta \), because the energy input and resultant SLM temperature are too high, the grains of the in situ strip- and particle-structured Al\(_4\)SiC\(_4\) are coarsened significantly (Figs. 4(d) and 5(d)), hence decreasing their strengthening effect upon the Al matrix. Moreover, the formation of a large amount of microscopic shrinkage porosity between the coarsened strip-structured Al\(_4\)SiC\(_4\) reinforcement and the matrix is also detrimental to the hardness and wear performance of the SLM-processed Al-based composites.

A close comparison of the phase constitutions (Fig. 3), microstructural characteristics (Figs. 4 and 5), and densification activity (Fig. 7) of the SLM-processed Al-based composites reveals that the limited hardness and wear performance of the part processed at a low laser linear energy density (\( \eta \)) is ascribed to: (i) the insufficient densification rate due to the formation of residual pores (Fig. 7) and (ii) the inhomogeneous reinforcing structure caused by the insufficient in situ chemical reaction and the residual SiC particles (Fig. 4(a)). For the SLM-processed composites at a high \( \eta \), although the sufficiently high densification level (Fig. 7) should favor the improvement in the mechanical performance, the significant coarsening of the in situ Al\(_4\)SiC\(_4\) reinforcements and the formation of the interfacial microscopic shrinkage porosity (Figs. 4(d) and 5(d)) weaken the hardness and wear property.

4 Conclusions

The SLM AM of the SiC/AlSi10Mg composite powder was performed to prepare the Al-based composites with the multiple reinforcing phases. The main conclusions were summarized as follows:

1. The reinforcing phases in the SLM-processed Al-based composites included the unmelted micron-sized SiC particles, the in situ formed micron-sized Al\(_4\)SiC\(_4\) strips, and the in situ produced submicron Al\(_4\)SiC\(_4\) particles. As the input laser energy density increased, the extent of the in situ reaction between the SiC particles and the Al matrix increased, resulting in the larger degree of the formation of Al\(_4\)SiC\(_4\) reinforcement with the more homogeneous microstructures.

2. The densification rate of the SLM-processed Al-based composite parts increased as the applied laser energy density increased. The sufficiently high densification response of \( \approx 96\% \) TD was achieved for the laser linear energy density larger than 1000 J/m.

3. Due to the generation of the multiple reinforcements after SLM, the elevated mechanical properties were obtained for the SLM-processed Al-based composites, having a high microhardness of 214 HV\(_{0.1}\), a considerably low COF of 0.39, and a reduced wear rate of \( 1.56 \times 10^{-3} \) mm\(^2\) N\(^{-1}\) m\(^{-1}\).

4. The grain size of the in situ formed Al\(_4\)SiC\(_4\) reinforcing phase, both the strip- and particle-structured Al\(_4\)SiC\(_4\), increased considerably using an excessive laser energy input. The significant grain coarsening of the reinforcement and the formation of the interfacial microscopic shrinkage porosity between the reinforcement and the matrix lowered the mechanical performance of the SLM-processed Al-based composites.

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