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Microstructure, Processability, and Strength of SiC-Reinforced AlSi9Mg Composite After Laser Surface Remelting and Post-Heat Treatment

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Abstract

The present study investigated the microstructure, processability, and mechanical strength of an AlSi9Mg-20vol.%SiC composite to assess its processing and mechanical performance during the laser powder bed fusion process. A simple laser surface remelting approach was adopted to simulate laser-based rapid solidification. The results revealed that this composite generally exhibited good laser processability, and the samples with the highest laser energy density and lowest scan speed possessed the best processability owing to the elimination of microcracks and pores. After laser processing, all the samples displayed a fine Al-Si cellular structure accompanied by in-situ formed fine needle-shaped Al₄SiC₄ particles. Increasing laser energy density considerably increased the area fraction of the Al₄SiC₄. The T5 aging treatment preserved the fine cellular structure and promoted the precipitation of a large number of Si nanoparticles and MgSi precipitates. During T6 solid solution treatment, the Si networks were broken down into coarse Si particles, disintegrating the cellular structure and reducing the strength. The T5 treatment was identified as the most suitable post-heat treatment for enhancing the microhardness and strength of the composite. Compared to conventionally laser-processed AlSi10Mg alloys, the AlSi9Mg-20vol.%SiC composite exhibited a significant increase in microhardness and yield strength.

Keywords: AlSi9Mg-SiC composite; laser surface remelting; laser energy density; microstructure; mechanical strength



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1. Introduction

In recent years, there has been a notable increase in the demand for high-performance aluminum matrix composites (AMCs) in the automotive, aerospace, and optical industries [1–4], owing to their outstanding combination of specific strength and stiffness, impressive wear resistance, and low coefficient of thermal expansion. Various ceramic reinforcements, including SiC, Al₂O₃, TiC, TiB₂, and ZrO₂ [3,5,6], have been employed to fabricate AMCs owing to their exceptional hardness and modulus. Among all of them, SiC is a preferred choice for fabricating AMCs owing to its low density, high hardness, high elastic modulus, low thermal expansion and low cost, rendering SiC-reinforced AMCs a significant contender for utilization in various industrial sectors [7]. For example, a commercial permanent mold cast Duralcan™ Al-SiC F3S.20S AMC with 20 vol.% discontinuously dispersed SiC particles in the T6 temper is known to have a density of 2.77 g/cm³, a

hardness of 77 HRB, an elastic modulus of 98.6 GPa, and a yield strength of 338 MPa [8]. However, in conventional approaches to the fabrication of AMCs, such as casting and powder metallurgy, the manufacture of complex geometrical components poses a significant challenge. J. Cyboron et al. [9] investigated the structural characteristics of an AlSi7Mg/SiC composite manufactured through the stir casting process. However, the SiC was found to be agglomerated throughout the material's volume. Dehghan Hamedan [10] explored how the parameters of stir casting affect the microstructure and mechanical properties of A356-1wt% SiC nanocomposites. Utilizing these optimized processing parameters, the nanocomposite attained an ultimate tensile strength of 173 MPa and a yield strength of 5.38%, respectively. Furthermore, the utilization of expensive customized molds and complex post-processing techniques in powder metallurgy unavoidably raises production costs and prolongs lead times [7].

Laser powder bed fusion (LPBF) is one of the most advanced and sophisticated additive manufacturing technologies that has gained significant popularity in producing intricate metallic components [11–14]. LPBF, which operates as a powder-bed-based manufacturing process, constructs bulk-form parts from the initial loose powder in a methodical layer-by-layer manner, aligned with the Computer Aided Design (CAD) information of the desired components [15–18]. Owing to its sequential track-by-track and layer-by-layer construction approach, LPBF provides the ability to fabricate components with a significant level of geometric flexibility. This capability can overcome the limitations of traditional manufacturing processes without incurring additional costs, such as for tooling [12,19]. Additionally, because of the transient interaction between the laser beam and the powder bed, the LPBF process experiences a significant temperature gradient (up to 10^5 K/m) and rapid cooling ($\sim 10^5$ K/s) [13,20,21]. Recently, Kaščák et al. reported [22,23] that the presence of significant thermal gradients, along with the rapid solidification process of the component in the LPBF method, can lead to not only elevated residual stresses but also considerable thermal distortion, ultimately resulting in part failure or diminished mechanical properties of the component. Consequently, the LPBF process exhibits exceptional potential for refining the microstructure and, hence, increasing the mechanical properties. However, the employment of the LPBF technique for AMC materials is accompanied by several drawbacks. These include insufficient fluidity and spreading of powders, which can lead to the emergence of defects and reduction of the material strength [24].

Recently, there has been a surge in interest in the utilization of LPBF for the production of AMCs that are reinforced with micron-sized SiC particles [25,26]. In composites reinforced with ceramic particles, it is known that when they are subjected to tensile stress, cracks tend to initiate and propagate at the interfaces between the ceramic particles and the matrix, resulting in a reduction in the strength and ductility of the composites [3,27–29]. In addition, aluminum matrix composites incorporating different SiC particle sizes and volume fractions were studied by Subrata et al. [30]. Their work clearly indicated that shrinkage stress and thermal stress during LPBF processing are the primary factors leading to cracks. When the SiC particle content exceeds 15%, both the crack density and wear resistance of the composite sample are enhanced. Moreover, Xue et al. [3] reported that the interface bonding in SiC-reinforced AMCs can be significantly improved through the utilization of an elevated temperature during LPBF. Despite the implementation of these conditions, the strength (342 MPa) and elongation (3.1%) were deemed unsatisfactory. The premature failure of these AMCs was attributed to the existence of numerous pores and cracked SiC particles. Additionally, the sharp edges of the SiC particles act as a severe stress concentration, thereby serving as a potential origin of cracks during loading [7,31]. On the contrary, some studies [15,32] have demonstrated that the mechanical properties of AMCs can be improved with smaller-sized SiC particles. It is also worth noting that reactions

often occur between SiC and the matrix during high-temperature LPBF processing, leading to the formation of brittle phases of Al_4C_3 and Al_4SiC_4 [2,7]. Similarly, Sercombe et al. [33] discovered that needle-like Al_4C_3 and residual Si particles are readily generated at the junction of the interlayer interface. As the laser energy density increases, the formation of the needle-like Al_4C_3 structure also rises. Such reaction phases in the molten pool, influenced by different laser energy inputs, can significantly affect the mechanical properties of AMCs. On the other hand, at high energy density, the wettability of SiC in the aluminum melt was enhanced with the increase in molten pool temperature, which suppressed the formation of the brittle Al_4C_3 phase and promoted the conversion of reaction products into Al_4SiC_4 . The intensified Marangoni convection and the decrease in melt viscosity contributed to a well-dispersed eutectic structure and improved mechanical characteristics [7]. Therefore, it is necessary to understand the impact of different processing parameters on the microstructure, defects, and mechanical properties of LPBF-fabricated AMCs.

The laser surface remelting (LSR) method is a cost-effective and simple approach to studying various phenomena that occur during laser-based additive manufacturing processes [34–38]. This process involves the utilization of a high-power laser to scan the metal surface, and the laser expeditiously raises the surface temperature above the liquidus temperature and generates a shallow molten pool [36] to simulate the laser-based rapid solidification. The aim of this work is to study the correlations between laser processing parameters and microstructure, processability, and mechanical properties of AlSi9Mg-20%SiC composites using the LSR method. Moreover, based on the existing literature, there is limited information regarding the processability with various linear energy densities and the impact of post-heat treatments on the development of reinforcing phases and precipitates in this composite. Therefore, the prime novelty of this study is the investigation of the effects of post-heat treatments (T5 and T6) on the aging response for different times for the first time and the related microstructures of laser-remelted composites. The specimens that were subjected to laser surface remelting (LSR) and peak aging under T5 and T6 conditions are examined using scanning electron microscopy (SEM) and transmission electron microscopy (TEM) to understand the formation of various reinforcing phases and their impact on strength.

2. Materials and Methods

The starting material was an AlSi9Mg-20vol.%SiC composite cast ingot with a D50 SiC size of 13 μm (known as Duralcan™ Al-SiC F3S.20S metal matrix composite [8], supplied by the Arvida Research and Development Centre of Rio Tinto Aluminium in Saguenay, Quebec. The aluminum matrix of this composite is a 359 foundry alloy, with nominally 9% Si, 0.55% Mg, up to 0.2% Cu, and up to 0.2% Fe (wt%) [8]. The starting material was first melted in an electric resistance furnace with a melting temperature of 740 °C and then poured into a copper permanent mold, which had a cooling rate of ~20 °C/s to obtain thin-wall cast plates measuring 110 mm $\times$ 110 mm $\times$ 4 mm. The samples with 4 mm thickness were cut from the cast plate, ground using SiC paper, and polished with diamond suspension to maintain the same surface roughness. These samples were placed on the substrate of an SLM SOLUTIONS 125 machine (SLM Solutions Group AG, Lubeck, Germany) for laser surface remelting (LSR), utilizing a reciprocating scanning strategy. The methodology of casting and laser surface remelting was adopted from [12,37].

During the LSR process, high-purity argon (99.99%) continued to flow in the forming cabinet to prevent the sample oxidation. The laser scans were conducted with a spot diameter of 100 μm and a hatch distance of 100 μm . By utilizing two laser powers and three scan speeds, a total of six process conditions with different linear energy densities were used on the cast plate samples to replicate 20 adjacent melt tracks at each condition.

This approach establishes a groundwork for exploring the effect of the process parameters on the processability and alloy properties of Al-SiC composites under conditions that are similar to the LPBF process. Figure 1 shows a schematic of the LSR process on the AlSi9Mg-20vol.%SiC cast plate, which was later metallographically prepared for microstructure characterization and microhardness measurement. The linear energy density (ED , J/mm) was calculated using Equation (1) [24], and the results are listed in Table 1.

$$ED = \frac{P}{V} \quad (1)$$

where P is the laser power (W), and V is the scan speed (mm/s).

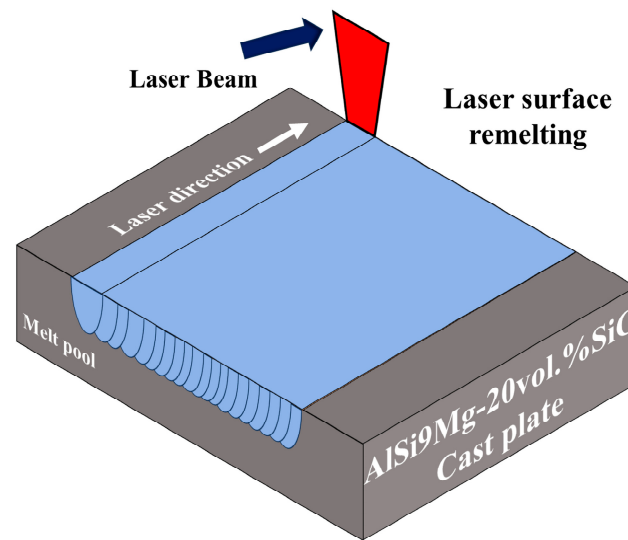


Figure 1. Schematic of the laser surface remelting on the cast plates.

Table 1. LSR process parameters and the laser linear energy densities for AlSi9Mg-20vol.%SiC composite.

Sample ID	Laser Power (W)	Scan Speed (mm/s)	Linear Energy Density (J/mm)
A	300	500	0.60
B	300	700	0.42
C	300	900	0.33
D	400	500	0.80
E	400	700	0.57
F	400	900	0.44

To study the aging response and the potential of precipitation strengthening after the LSR processing, the samples were subjected to post-heat treatments corresponding to T5 and T6 conditions. For the T5 treatment, the LSR samples were directly aged at 160 °C for up to 24 h. On the other hand, the T6 treatment involved a solution treatment at 530 °C for 2 h, followed by water quenching and then artificial aging at 160 °C for up to 24 h.

For microstructure analysis, the cross sections of laser-remelted samples were metallographically prepared and observed using the optical microscope (OM, Nikon, Eclipse ME600, Nikon Instruments Inc., Melville, NY, USA) and scanning electron microscope (SEM, JEOL-6480LV, JEOL USA Inc., Peabody, MA, USA) after etching the cross sections with Keller's reagent for a duration of 20 s. To examine the presence of different nanoscale phases and precipitates, a transmission electron microscope (TEM, JEM 2100, JEOL USA Inc., Peabody, MA, USA) operating at 200 kV was employed. Samples for the TEM study were ground mechanically, and 3 mm circular disks were carefully extracted from the laser-

remelted area, which underwent thinning using a Gatan 691 Precision Ion Polishing System (Gatan Inc., Pleasanton, CA, USA). All TEM images were taken in the vicinity of the [100] zone axis. Vickers hardness measurements using an NG-1000 CCD machine (NextGen Material Testing, Vancouver, BC, Canada) were carried out on all LSR heat-treated samples. The indentations were carefully placed between SiC particles to measure hardness values in the aluminum matrix only. For each condition, a total of 10 measurements were performed at a 25 g load with 20 s dwell time.

3. Results

3.1. Microstructures After Casting and Laser Surface Remelting

The as-cast microstructure of the copper permanent mold casting is presented in Figure 2. It consisted of equiaxed Al dendrite grains and fine Al-Si eutectics, accompanied by randomly distributed SiC particles (Figure 2a). The average Al grain size was measured to be 35–40 μm . Most SiC particles were distributed within the Al-Si eutectic (Figure 2b).

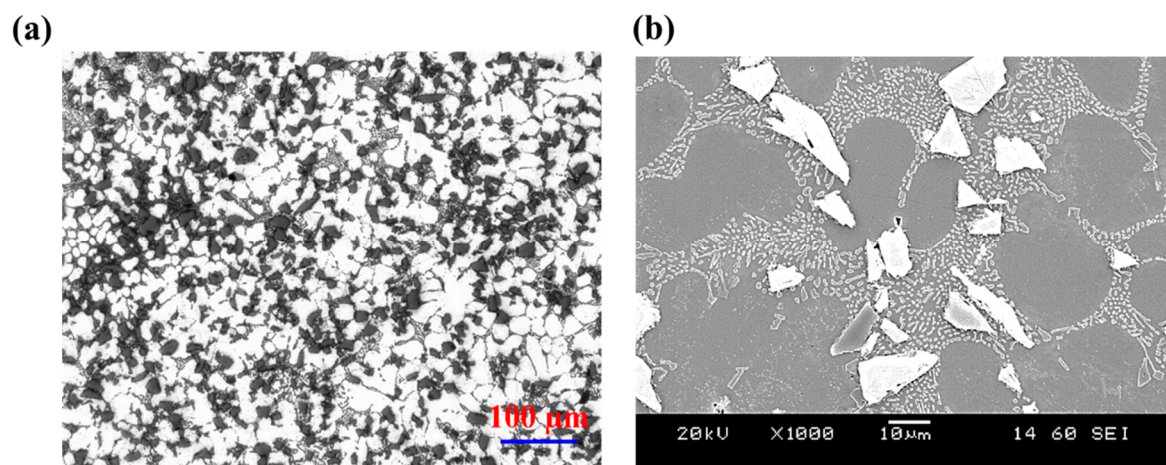


Figure 2. As-cast microstructure of the copper permanent mold casting for AlSi9Mg-20%SiC composite: (a) optical, (b) SEM images.

Figure 3 shows cross-sections of the laser-scanned areas of the AlSi9Mg-20%SiC composite for various processing conditions (Table 1), where the upper section represents the laser-remelted areas, and the bottom section shows the original as-cast microstructure. The red lines illustrate the melt pools. In all processing conditions, no macro solidification cracks were found, indicating generally good laser processability for the AlSi9Mg-20%SiC composite. The process of LSR involves the dissipation of absorbed heat from the molten pool by surrounding materials to simulate the laser-based additive manufacturing process. The applied laser energy input has a significant effect on the development of the melt pool cross-section. The widths of the melt pools were determined to be 320, 210, 185, 410, 310, and 220 μm for the A, B, C, D, E, and F processing conditions, respectively. It is evident that the width of the melt pools clearly increases with increasing linear energy density.

Figure 4 shows enlarged cross-sections of melt pools for various process parameters. The melt pools of all samples consisted of fine Al-Si cellular structures with eutectic Si networks in the intercellular regions, a long needle-shaped Al_4Si_4 phase, and polygonal SiC particles. The SiC particles in the melt pools were distributed randomly and uniformly in the macro scale during the LSR process. As the laser linear energy density increased, both the morphology and size of the SiC particles slightly changed. The particles underwent a partial decomposition, and the morphology of the SiC particles changed from polygonal particles with sharp corners to more rounded particles, resulting in a slight decrease in their size.

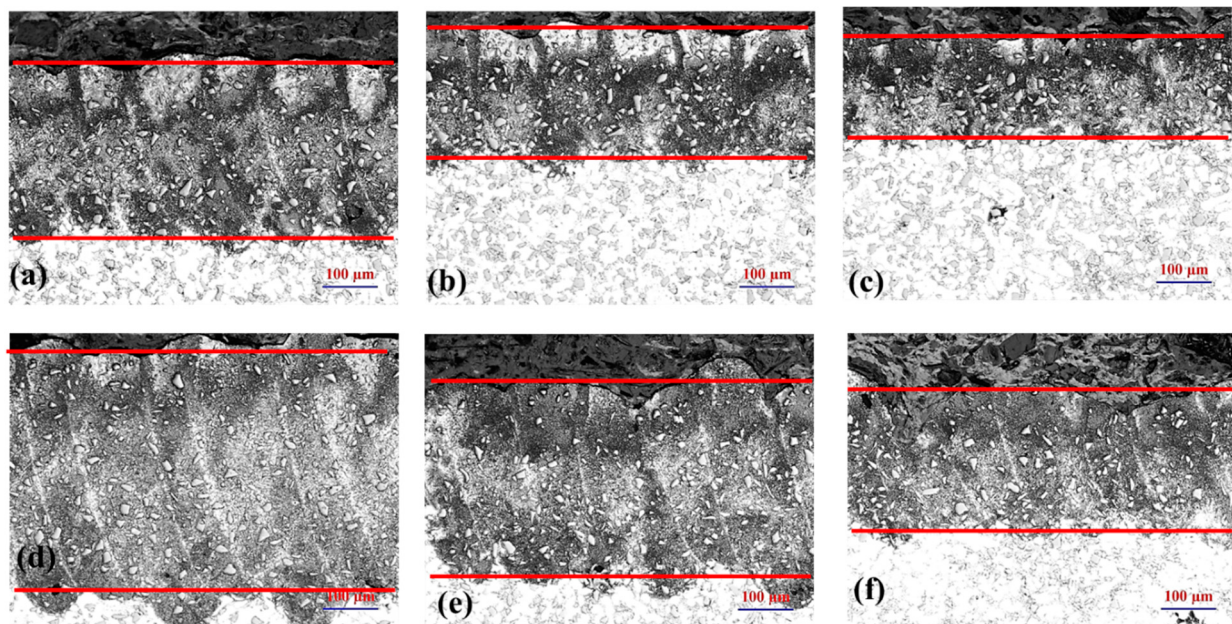
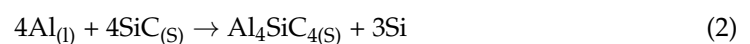


Figure 3. Optical cross-sectional microstructures after LSR for (a) A, (b) B, (c) C, (d) D, (e) E, (f) F conditions for AlSi9Mg-20%SiC composite. The red lines indicate the widths of the melt pools in each condition.

At the late stages of melt pool solidification, microcracks emerged within SiC particles and along the boundaries between SiC and Al grains, owing to the combined effects of thermal stresses and solidification shrinkage. In this study, samples A and D with high energy density were almost free of defects such as microcracks and pores (Figure 4a,b,g,h). The other samples (B, C, E, and F), with lower linear energy density, exhibited microcracks and pores in different extensions. It indicates that both samples with the highest linear energy density of 0.6–0.8 J/mm and the lowest scan speed of 500 mm/s (A and D) possess the best processability for laser-based additive manufacturing processes such as the LPBF process.

Under the intense energy density of the laser beam, the melt pools reach a high temperature where a chemical in-situ reaction occurs between the liquid aluminum and SiC particles [2].



Under this reaction, needle-shaped Al_4SiC_4 and granular Si particles formed in the melt pool. According to [39], raising the laser energy density enhanced the level of in-situ reaction within the melt pool and intensified the chemical reactions occurring at the interfaces. Figure 5 shows a detailed view of the microstructure in the melt pools under different process conditions. In the samples with high energy density (Figure 5a,d), the microstructures contained a significant amount of needle-shaped Al_4SiC_4 and numerous sub-micron granular Si particles. On the other hand, the medium energy-processed samples (Figure 5b,e) showed a reduced number of Al_4SiC_4 and granular Si particles, whereas the low energy-processed samples (Figure 5c,f) exhibited minimal Al_4SiC_4 phase. The laser energy density played a crucial role in determining the area fraction of the reaction products in the LSR process (Table 2). Increasing the energy density directly resulted in a higher area fraction of the Al_4SiC_4 phase. Its area fraction was 4.1% when the energy density was 0.33 J/mm in the sample C, while the maximum area fraction of Al_4SiC_4 reached 21.3% with 0.8 J/mm energy density in the sample D.

Table 2. The area fraction of Al₄SiC₄ with different laser energy densities.

Sample IDs	Linear Energy Density (J/mm)	Area Fraction of Al ₄ SiC ₄ (%)
A	0.60	14.55
B	0.42	7.24
C	0.33	4.10
D	0.80	21.29
E	0.57	12.49
F	0.44	8.85

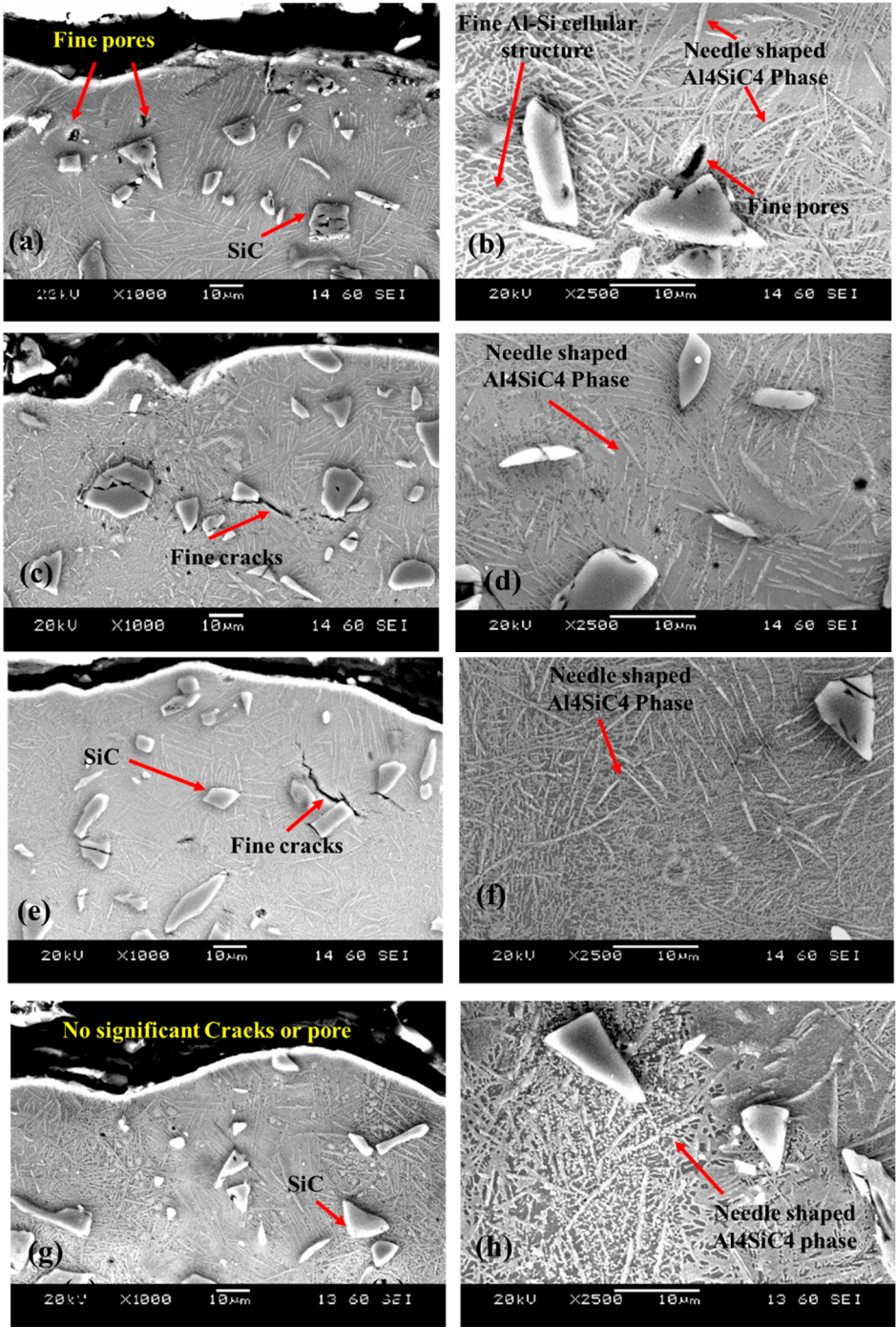


Figure 4. Cont.

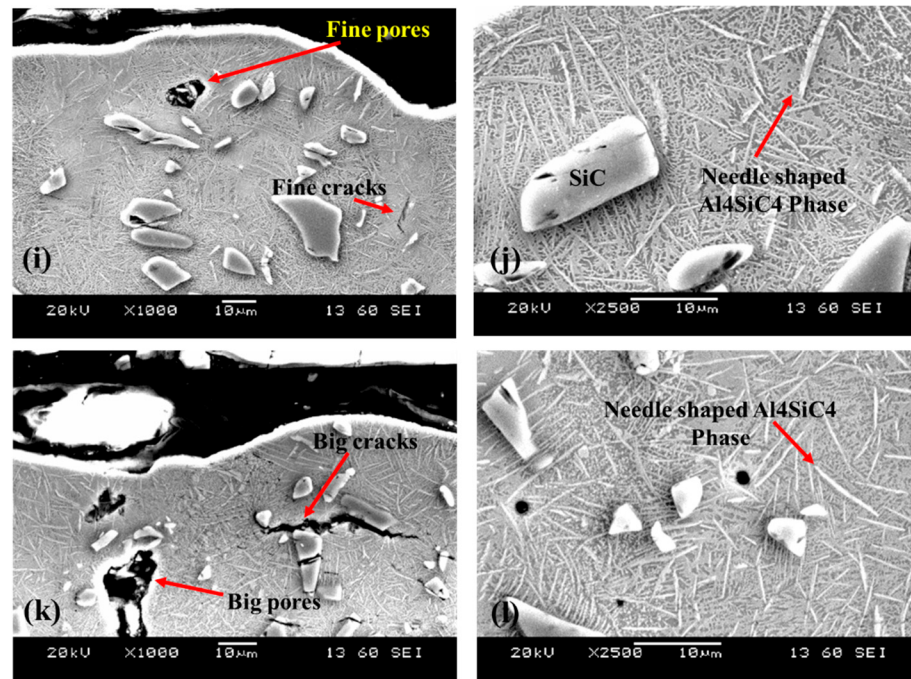


Figure 4. SEM images of the cross-sectional microstructures after LSR for the samples with different processing parameters: (a,b) sample A, (c,d) sample B, (e,f) sample C, (g,h) sample D, (i,j) sample E, (k,l) sample F for the AlSi9Mg-20%SiC composite.

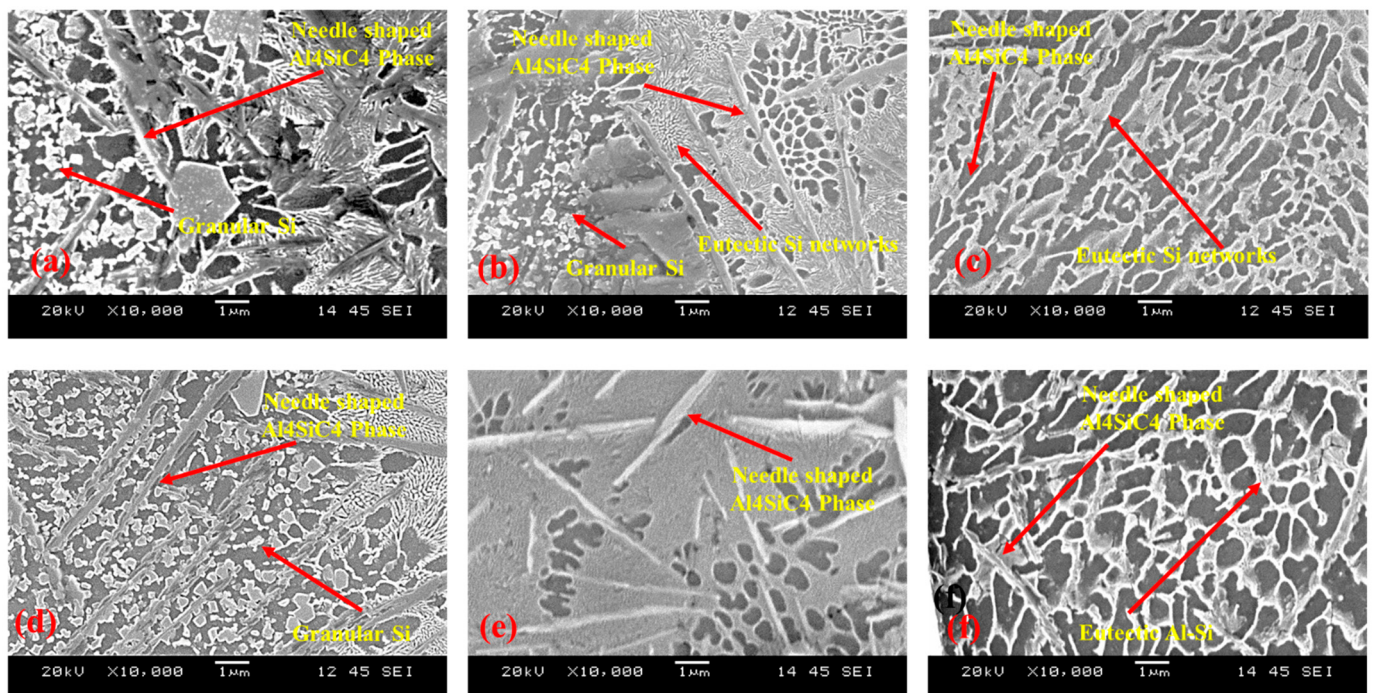


Figure 5. Detailed view of microstructures in melt pools under different process conditions with high-magnification SEM images: (a) condition A, (b) condition B, (c) condition C, (d) condition D, (e) condition E, and (f) condition F.

3.2. Microstructure Evolution During T5 and T6 Post-Heat Treatments

The direct aging of the T5 treatment effectively maintained the integrity of the Al-Si cellular structure for all laser-processed samples. It was observed that no changes in the morphology and size of the SiC and Al₄SiC₄ phases occurred during the T5 treatment owing to the low ageing temperature. Because of the rapid solidification conditions of

the LSR process, the aluminum cell matrix became highly supersaturated with Si and Mg solutes after LSR, which promoted the precipitation of nanoscale Si and MgSi strengthening phases that are discussed later based on TEM observations.

Figure 6 shows SEM images of T6 peak-aged samples in different process conditions. After the T6 solid solution treatment, the microstructures of all LSR samples were coarsened owing to the high solution temperature applied (530 °C for 2 h). The SiC particles remained intact, but eutectic Si networks were completely broken up, followed by the coarsening and spheroidization of Si particles. This resulted in a bi-modal distribution of Si particles in the T6 microstructure as depicted in Figure 6, i.e., both coarse and irregular Si particles originating from disintegrated Si networks (approx. 4 μ m), and small spherical Si particles originating from the in-situ reaction product of granular Si particles (approx. 1–2 μ m). Meanwhile, the fraction of the needle-shaped Al_4SiC_4 was slightly reduced for all samples as a result of the high-temperature solution treatment.

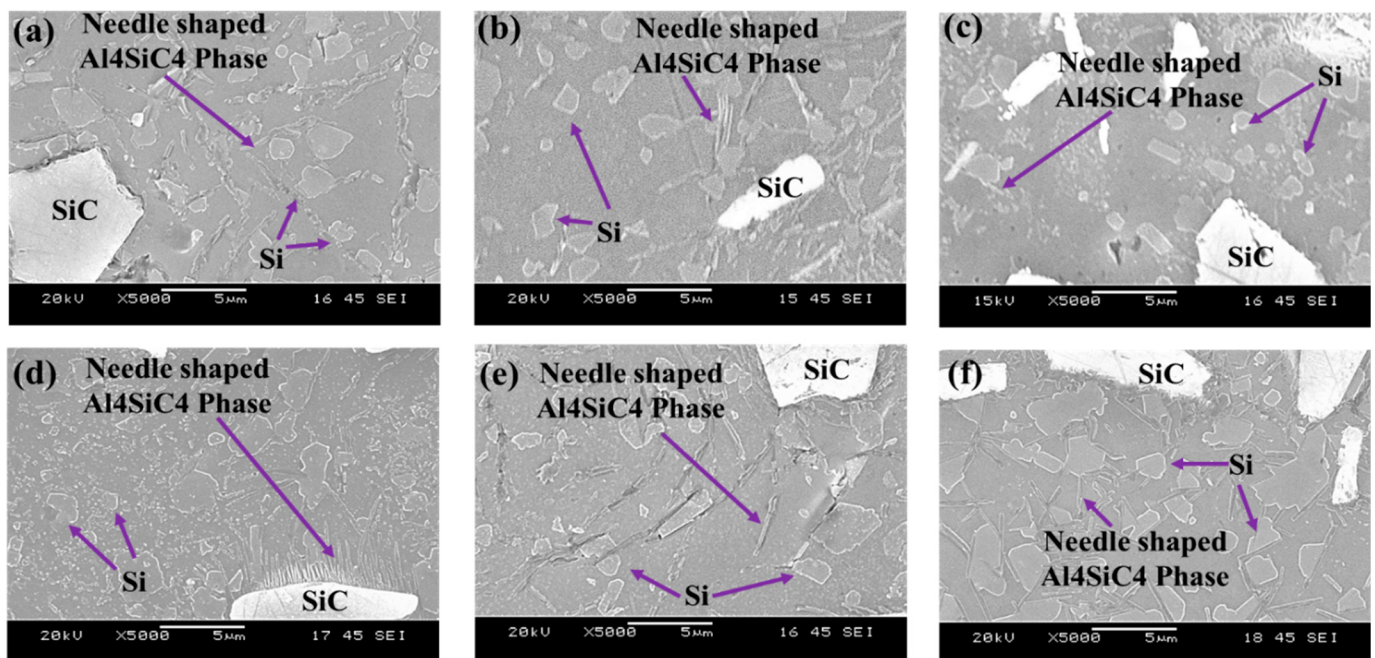


Figure 6. SEM images showing the microstructures after T6 treatment: (a) A, (b) B, (c) C, (d) D, (e) E, and (f) F conditions.

The TEM results in Figure 7 reveal the precipitation microstructures for sample D in the as-laser-scanned, T5, and T6 conditions. In the as-laser-scanned condition (Figure 7a), the eutectic Si networks and α -Al formed fine cellular structures. The needle-shaped Al_4SiC_4 was randomly disturbed across the cellular structures. It was reported [40] that in the as-laser-scanned condition, Si particles exhibited numerous stacking faults and dislocations, effectively compensating for the lattice misfit between the Al (200) and Si (220) planes. Consequently, the interface between the Si network and Al matrix exhibited a semi-coherent nature. After the T5 treatment (Figure 7b), the Al-Si cellular structure and Al_4SiC_4 phase remained intact. A large number of nano-sized Si particles in the form of fine granules and rod-like β'/β'' -MgSi precipitates with certain orientations were formed in the Al cell matrix (Figure 7e). The average diameter of the Si nanoparticles was 21 nm, while the average length of the MgSi precipitates was 75 nm. After the T6 treatment (Figure 7c), the eutectic Si networks were completely broken down into irregular coarse Si particles. The needle-shaped Al_4SiC_4 was less affected. Rod-like β'/β'' -MgSi precipitates were the main precipitates in the Al cell matrix. The average size of the MgSi precipitates in the T6 condition was 66 nm (Figure 7f).

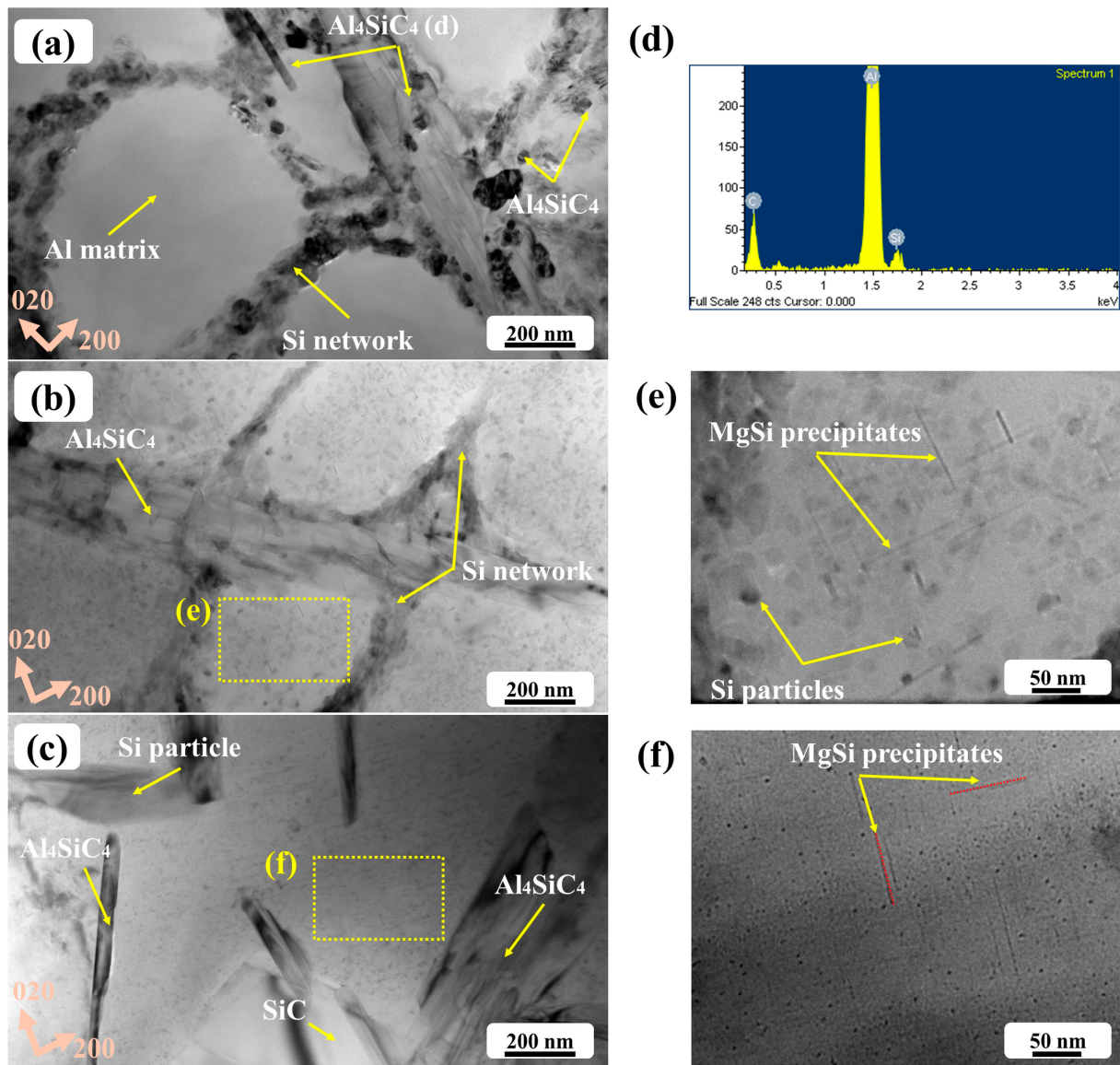


Figure 7. Bright-field TEM images of sample D showing the cellular structure and precipitation microstructure for (a) as-laser-scanned, (b) T5, and (c) T6 conditions; (d) TEM-EDS analysis of the Al₄SiC₄ phase in (a), (e) enlarged view of fine Si particles and MgSi precipitates in (b) (T5), and (f) enlarged view of MgSi precipitates in (c) (T6).

3.3. Microhardness Evolution and Aging Responses During T5 and T6 Treatments

Figure 8 displays the microhardness of as-cast and laser-remelted samples under different processing conditions. The as-cast sample had a low hardness value (68 HV). All laser-remelted samples exhibited a much higher microhardness compared to the as-cast sample. The enhancement in hardness can be attributed to two primary factors. Firstly, the LSR process resulted in the formation of a very fine Al-Si cellular structure (Figure 5) owing to rapid solidification conditions. The eutectic Si networks served as the primary load-bearing structure and hindered dislocation movement. Secondly, the LSR induced the high temperature in the melt pools and promoted the in-situ reaction, leading to the generation of a large number of the reinforcing needle-shaped Al₄SiC₄ particles (Figures 4 and 5). When the laser energy density was low in the range of 0.33–0.44 J/mm, the samples B, C, and F showed a relatively lower microhardness, which is attributed to the fact that the lower energy density resulted in a less intensive in-situ reaction and hence a lower area fraction of the Al₄SiC₄ phase. As the laser energy density increased to 0.6–0.80 J/mm, samples A and

D exhibited the highest microhardness, owing to the intensive in-situ reaction producing a higher area fraction of the Al_4SiC_4 phase (Table 2).

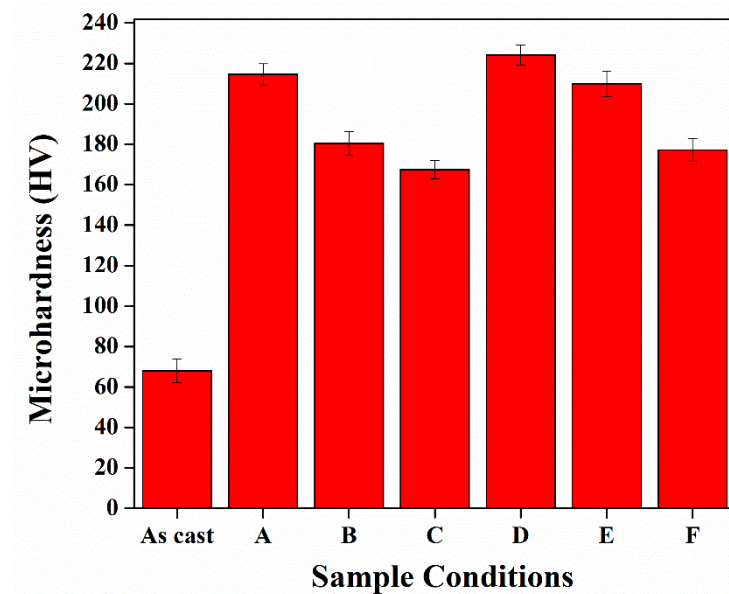


Figure 8. Microhardness of as-cast and as-laser-remelted samples in different processing conditions.

Figure 9 shows the microhardness values as a function of the aging time for both as-cast and laser-remelted samples. In the direct aging T5 condition (Figure 9a), the HV values of the as-cast sample were relatively low, increasing from 68 HV in the initial state to 74 HV after 1 h of aging. It slowly increased with increasing aging time and reached a peak hardness of 87.2 HV after 12 h and remained relatively stable up to 24 h. In general, all laser-remelted samples demonstrated a significant enhancement in hardness (close to 200%) as compared to the cast sample. Their HV values gradually increased with increasing aging time and reached the peak hardness after 12 h. Sample D exhibited the highest peak hardness of 261 HV, followed by samples A and B. The T5 treatment maintained the original cellular structure and needle-shaped Al_4SiC_4 phase in the laser-remelted samples. The increasing strengthening effect during aging is due to the precipitation of a large number of nano-scale Si and MgSi precipitates from the supersaturated solid solution (Figure 7b,e), which acted as supplementary barriers against dislocation movement [41].

(a)

(b)

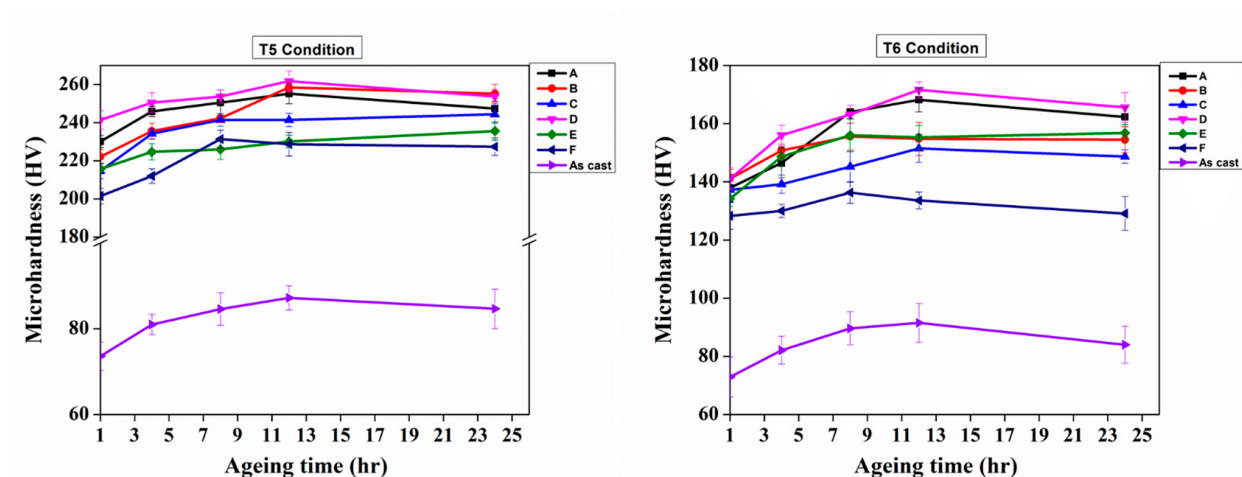


Figure 9. Aging responses of laser-remelted AlSi9Mg-20%SiC composite in (a) T5 and (b) T6 conditions.

In the T6 condition (Figure 9b), the Vickers hardness of the as-cast sample increased slowly with aging time and reached the peak of 91 HV after 12 h. Then it slightly decreased with increasing aging time. All T6 laser-remelted samples exhibited a higher increment in HV with increasing aging time relative to the as-cast sample and reached peak hardness after 12 h of aging. Subsequently, the hardness of all laser-remelted samples remained relatively stable for up to 24 h. Again, samples D and A exhibited the highest peak hardness of ~170 HV in the T6 condition. The aging response and strengthening effect of the T6 treatment are attributed to the precipitation of a large number of nano-scale MgSi precipitates from the supersaturated solid solution induced by the solution treatment (Figure 7c,f). Notably, the peak hardness values of all T6 laser-remelted samples were considerably lower than those in the T5 condition (~65% lower in the case of the sample D). This was attributed to the fact that the T6 condition displayed a coarse microstructure where the eutectic Si networks were fragmented into coarse and irregular Si particles (Figure 6), resulting in a partial loss of strength.

4. Discussion

4.1. Microstructure Evolution of SiC-Reinforced AlSi9Mg Composite During Laser-Based Manufacturing Process

The AlSi9Mg-20%SiC composite generally exhibited good processability and favorable wettability between the SiC ceramic particles and the Al matrix during the LSR process. The distribution of SiC particles within the molten pool was found to be random (Figure 4). The LSR process leads to rapid melting and solidification, which in turn induces a high cooling rate and substantial temperature gradient within the molten pools [42]. As reported by [4], the maximum temperature of such molten pools is around 1150 K (depending on the processing conditions), which is significantly lower than the 3100 K melting point of SiC ceramic. This suggests that the SiC particles were not melted during the laser process. To achieve good processability during LSR, it is important to have adequate laser energy and, hence, a sufficiently high temperature, ensuring the complete melting of the metallic material. Additionally, the appropriate laser energy density plays a crucial role in establishing excellent bonding between SiC particles and the Al melt, resulting in higher density and superior properties. The interaction of the laser with the mixed powders results in a temperature gradient within the composites owing to the Gaussian distribution of the laser beam, which leads to the establishment of Marangoni convection [3]. The presence of Marangoni convection [39], which is influenced by the laser energy, effectively eliminates defects such as pores and non-fusion. It also enhances the distribution of SiC particles in the melt pool and strengthens their bonding with the Al melt [39]. The results in Figure 4 revealed that the samples with the higher laser energy density between 0.6–0.8 J/mm and a lower scan speed of 500 mm/s possess the best processability for laser-based additive manufacturing processes such as LPBF.

However, the needle-like Al_4SiC_4 reaction product is non-uniformly distributed in the Al matrix (Figure 4). This distribution is attributed to the in-situ reaction of SiC particles within the Al molten pools, which are also influenced by the laser energy density. When the laser linear energy density is below 0.57 J/mm, the in-situ reaction of Al with SiC is less intensive, resulting in a lower area fraction of Al_4SiC_4 . Increasing the energy density directly induces a higher in-situ reaction and hence a higher area fraction of the Al_4SiC_4 phase (Figure 6 and Table 2). The sample with the highest energy density of 0.8 J/mm reached the maximum Al_4SiC_4 area fraction of 21.3%, compared to the area fraction of 4.1% in the sample with the lowest energy density of 0.33 J/mm.

Figure 10 provides a schematic comparison of the microstructure evolution during the laser remelting and post-heat treatment processes for both an AlSi10Mg alloy and the

current AlSi9Mg-20%SiC composite. The AlSi10Mg microstructure after laser processing comprises a fine cellular structure in which a supersaturated aluminum matrix is decorated with eutectic Si networks in the surrounding intercellular regions [39]. By contrast, owing to the in-situ reaction between the liquid aluminum and SiC particles, needle-shaped Al_4SiC_4 and granular Si particles are formed and randomly distributed in the microstructure of the AlSi9Mg-20%SiC composite (Figure 10a). During T5 treatment, fine Si nanoparticles and MgSi precipitates are precipitated out from the supersaturated aluminum matrix in both materials (Figure 10b). Because of the low-temperature aging treatment, the original fine cellular structure remains intact, preserving the structural integrity of the Si networks. Upon T6 treatment, the microstructures of both materials undergo a rapid transformation. The Si networks are broken down into coarse Si particles, and a number of MgSi precipitates are formed in the aluminum matrix (Figure 10c). The reinforcing Al_4SiC_4 phase in the AlSi9Mg-20%SiC composite remained unchanged during both the T5 and T6 treatments.

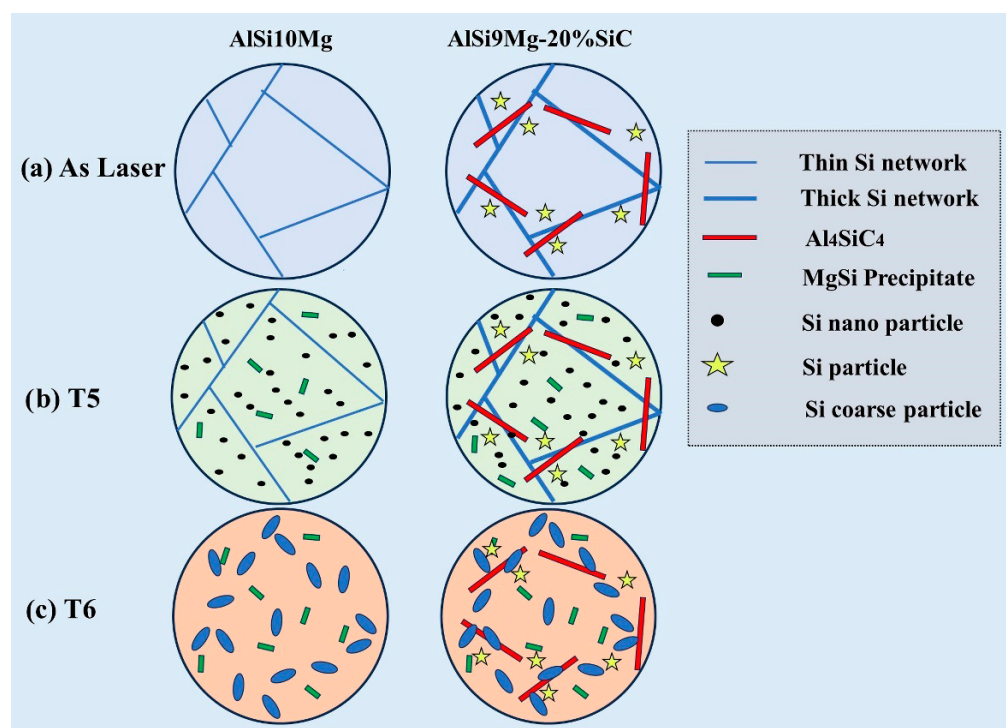


Figure 10. Schematic comparison of the microstructural evolution of a laser-processed AlSi10Mg alloy and AlSi9Mg-20%SiC composite, (a) as-laser melted, (b) T5 condition, and (c) T6 condition.

4.2. Comparison of Strength Between AlSi9Mg-20%SiC and Conventional AlSi10Mg

Due to the nature of the LSR process, it was not possible to produce tensile test samples of the AlSi9Mg-20%SiC composite. To have a first estimation of the yield strength of the AlSi9Mg-20%SiC composite and compare it to conventional AlSi10Mg alloys fabricated by LPBF, an attempt has been made to correlate the microhardness with the yield strength for the LPBF materials. The relationship between the yield strength (σ_{yield} in MPa) and the Vickers hardness can be established using the following equation [43,44]:

$$\sigma_{yield} = 9.81 \left(\frac{HV}{3} \right) 0.1^n \quad (3)$$

where HV is the Vickers hardness value, influenced by the strain hardening coefficient n . Several recent studies conducted on LPBF AlSi10Mg alloys have indicated that the value of n falls within the range of 0.18–0.25 [45,46]. To simplify calculations, a constant value of $n = 0.2$ is used to estimate the yield strength in the as-built condition, while the

utilization of $n = 0.1$ is implemented for materials in the T5 and T6 conditions owing to the reduction in the n value caused by age hardening. The estimated yield strengths for the AlSi9Mg-20%SiC and AlSi10Mg materials are listed in Table 3, along with the yield strength values obtained from experimental tensile results conducted on LPBF AlSi10Mg alloys in the literature.

Table 3. Comparison of Vickers hardness and yield strength between laser-processed AlSi9Mg-20%SiC and conventional AlSi10Mg materials.

Conditions	Materials	Hardness (HV)	Yield Strength (MPa)	References
As-built	AlSi9Mg-20%SiC	224 *	461 **	Present study
	AlSi10Mg	120	276	[47]
	AlSi10Mg	135	293	[48]
	AlSi10Mg	125	268	[49]
	AlSi10Mg	135	279	[48]
	AlSi10Mg	128	263	[50]
T5	AlSi9Mg-20%SiC	261 *	678 **	Present study
	AlSi10Mg	130	296	[47]
	AlSi10Mg	145	313	[42]
	AlSi10Mg	139	310	[50]
T6	AlSi9Mg-20%SiC	172 *	446 **	Present study
	AlSi10Mg	123	248	[42]
	AlSi10Mg	103	239	[49]
	AlSi10Mg	75	153	[48]
Permanent mold casting	AlSi9Mg-20%SiC (T6)	117	338	[8]

* Maximum measured Vickers hardness values in the current study. ** Estimates based on conversion from HV using Equation (3).

Compared to the AlSi10Mg alloys, the AlSi9Mg-20%SiC composite exhibited a significant increase in HV and yield strength, with both the HV and yield strength values being almost doubled. For instance, in the as-built condition, the yield strength of AlSi10Mg alloys ranges from 263 to 293 MPa, but the yield strength of the AlSi9Mg-20%SiC reaches 461 MPa. Similarly, in the T5 condition, the yield strength of AlSi10Mg alloys varies from 296 to 310 MPa, while the AlSi9Mg-20%SiC achieves an estimated yield strength of up to 676 MPa.

The significant enhancements of HV and yield strength in the AlSi9Mg-20%SiC composite in the as-built, T5, and T6 conditions are primarily attributed to the presence of a large number of ex-situ SiC particles and in-situ Al_4SiC_4 particles in the matrix, resulting in the increased resistance against deformation and the introduction of residual stresses. The SiC and Al_4SiC_4 particles can effectively impede dislocation and grain boundary motion and contribute to work hardening, thereby increasing the strength of the AlSi9Mg-20%SiC composite. In addition, the AlSi9Mg-20%SiC composite fabricated by laser processing exhibits a significant residual stress owing to the disparity in the coefficient of thermal expansion between SiC particles ($4.63 \times 10^{-6}/\text{K}$) and AlSi10Mg alloy ($21.7 \times 10^{-6}/\text{K}$) [39]. Furthermore, in the T5 microstructure, the precipitation of a large amount of ultrafine Si and MgSi nanoparticles in the aluminum matrix also plays a crucial role in impeding dislocation movement. Consequently, the T5 condition attains the highest level of HV and yield strength when compared to other as-built and T6 conditions of the AlSi9Mg-20%SiC samples.

On the other hand, the achievable mechanical strengths of the AlSi9Mg-20%SiC AMC fabricated by LPBF are significantly higher compared to traditional permanent mold casting, owing to a much more rapid solidification, giving a more refined microstructure. For instance, the AlSi9Mg-20%SiC AMC fabricated by laser processing has an estimated

yield strength of 443 MPa in the T6 condition, whereas the T6 yield strength of a permanent mold casting typically reaches 338 MPa.

5. Conclusions

In this study, the microstructure, processability, and mechanical properties of an AlSi9Mg-20%SiC composite were systematically investigated after laser surface remelting (LSR) and post-heat treatments (T5 and T6). The following conclusions can be derived from the main results obtained.

1. The AlSi9Mg-20%SiC composite generally exhibited good laser processability owing to the lack of macro solidification cracks with the current processing parameters studied. The samples with the highest laser energy density of 0.6–0.8 J/mm and the lowest scan speed of 500 mm/s possessed the best processability for the laser-based processes owing to the elimination of microcracks and pores. The width and cross-section of the melt pools increased with increasing laser energy density.
2. During the LSR process, all the samples displayed a fine Al-Si cellular structure, in which a supersaturated aluminum matrix was surrounded by eutectic Si networks in the intercellular regions. Owing to the in-situ reaction between the liquid aluminum and SiC particles, a needle-shaped Al_4SiC_4 phase was formed and randomly distributed in the melt pools of the composite. With increasing laser energy density, the area fraction of these fine Al_4SiC_4 needles considerably increased, resulting in significantly higher hardness and strength compared to the as-cast samples.
3. The T5-treated samples preserved the fine Al-Si cellular structure and the strengthening effect of the fine eutectic network. Additionally, a large number of Si and MgSi nanoparticles were precipitated out upon aging. During T6 solid solution treatment, the Si network was broken down into coarse Si particles, disintegrating the Al-Si cellular structure. Although a number of MgSi precipitates were formed during the subsequent aging, the microhardness was significantly lower than in the as-laser-scanned and T5 conditions. The T5 condition attained the highest level of microhardness and strength when compared to the as-laser-scanned and T6 conditions.
4. Compared to conventionally laser-processed AlSi10Mg alloys, the AlSi9Mg-20%SiC composite exhibited a significant increase in microhardness and yield strength, with both values being almost doubled. This significant strengthening effect was attributed to the presence of a large number of ex-situ SiC particles and in-situ Al_4SiC_4 phase in the matrix, which are known to increase the resistance against deformation and induce internal residual stresses. In future work, the mechanical properties and optimal process parameters of AlSi9Mg-20%SiC composite will be verified with LPBF-manufactured parts to substantiate the findings obtained from LSR in real additive manufacturing contexts.

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