Preparation of Silicon Carbide Reinforced Aluminium Matrix Composites (SiC/Al) by Selective Laser Melting

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Abstract: Silicon carbide reinforced aluminium matrix composites (AMCs) attracted a lot of attention owing to their low density, high specific strength and specific stiffness, and good wear resistance. Selective laser melting (SLM) technique has been widely used in the preparation of complex structural product with high precision, short production cycle and low cost. In this paper, silicon carbide (SiC) powders and aluminium (Al) powders were mixed together firstly with a weight ratio of 15:85 and then used to prepare AMCs by SLM technology. The effects of laser power on the microstructures, microhardness and flexural strength of formed AMCs were studied. The results show that the relative density, microhardness and flexural strength of the as-prepared samples can be improved with increasing the laser power. Meanwhile, the samples prepared at higher laser power exhibit lower friction coefficient.

1. Introduction

Owing to their high specific strength and stiffness, high wear resistance, particulate reinforced aluminum matrix composites (PRAMCs) are enormously applied in the automotive and aerospace industries[1]. Ceramic particles with high strength and hardness are widely used to prepare AMCs and various investigations have been reported[2-9]. Among different types of ceramic materials, SiC has been recognized as the most promising candidate for preparing high performance AMCs owing to its low thermal expansion, high hardness and excellent corrosion resistance[10].

The powder metallurgy method is the most widely used way for preparing SiC/Al composites. In this process, Al-based alloy powders are mixed together with SiC powders by grinding or mechanical alloying firstly, and then the mixed powders are fired at high temperatures by means of hot isostatic pressing sintering[11], spark plasma sintering[12] or microwave sintering[13]. Unfortunately, these methods always suffer some disadvantages, such as long processing time and high cost, agglomeration of reinforcing phases in the matrix and interface microcracks. Most particularly, it’s difficult to prepare final products with complex shape. It is well known that additive manufacturing (AM) can overcome the above mentioned drawbacks. Moreover, as one of the most important branches of AM, selective laser melting (SLM) was widely used to prepare some complex configurations based on a computer CAD model in a layer-by-layer manner at first, and then the thin layer of loose powders were selectively melted and consolidated by a moving laser beam[14,15]. Meanwhile, it’s easy to obtain non-equilibrium phases with fine-grained microstructures and novel properties under a fast melting/consolidation rate of $10^3$–$10^8$ K/s during the SLM process[16].
A number of studies have been performed to use AM for fabricating aluminum alloys. For example, Noriko[17] et al reported that the laser power and scan speed have a major influence on the porosity development in the AlSi10Mg alloy. Zou[18] et al studied the effect of energy density on the density of AlSi10Mg alloy, they pointed out that the density of the samples increases gradually with increasing the laser energy density to 5.09 J/mm² and then started to decrease with keeping on increasing laser energy density. However, the use of SLM for preparing SiC/Al composites and the effects of SLM process parameters on microstructure evolution and mechanical properties of SiC/Al composites have not been studied in detail and systematically. In this paper, SiC/Al composites were prepared by SLM technique and the effects of laser power on the densification behavior, microstructure evolution and mechanical properties of prepared samples were investigated.

2. Experimental procedures

2.1. Powder materials
The raw materials include 99.3% purity aluminum powder with spherical shape of 33 μm in average diameter and 99% purity SiC powder with a polygonous structure of 30 μm in average particle size. The Al and SiC powders were mixed in the weight ratio of 85:15 before SLM process.

2.2. SLM processing
The SLM process was carried out by a WJ SLM225 SLM machine equipped with YLR-500-WC laser beam at 1.06 μm wavelength (Wuhan Huake 3DTechnology Co. Ltd, China). The laser beam had a spot size of 100 μm and a maximum power of 500 W. High purity argon atmosphere was used as the shielding gas for laser irradiation. The thickness of the powder layer was about 0.05 mm. A strip hatch style with an angle of 17° was applied between the layers to improve the quality of final samples.

2.3. Characterization
The relative density of the as-prepared samples was measured by the Archimedes drainage method. Phases in the final product samples were identified by means of powder X-ray diffraction (XRD) analysis using Mini Flex600 diffractometer (PANalytical, Japan). Patterns ranged 20–90° (2θ) were recorded using Cu Kα radiation(λ=0.1542 nm) at 15 mA and 40 kV. Field emission scanning electron microscopy (FE-SEM, Nova400NanoSEM, PHILIPS, NETHERLANDS, 15 kV), equipped with an energy spectrometer (EDS, Le 350 Penta FEI), was employed to observe the morphology and examine the elemental composition of the samples. The microhardness of the cross section of the polished samples was measured by SH-318-III digital display Vickers hardness tester under a test load of 100 g with 10 s. The friction and wear properties were measured by UMT-2 friction and wear tester through ball-on-flat reciprocating test at room temperature. HRC62 carbon steel balls with a diameter of 2mm were used for linear reciprocating motion. The friction and wear tests were carried out under dry friction, normal load of 20 N, friction rate of 5 mm/s, friction stroke of 6 mm and test duration of 20 min. The tribological test for each sample was repeated three times and then the friction coefficient and wear rate were calculated by averaging the obtained data. The flexural strength was measured by SGW-II digital bent strength testing instrument of engineering porcelain and ceramic.

3. Results and discussion
Figure 1 shows XRD patterns of samples prepared by SLM technique with different laser power. Both strong diffraction peaks of Al phase and SiC phase were observed for all samples, confirming the formation of SiC-reinforced Al composites. In addition, Al₄C₃ and Si were also detected in all samples, which meant that an in situ reaction occurred between aluminum and SiC during the SLM process. Iseki[19] et al reported that aluminum carbide (Al₄C₃) phase can be formed at the interface in the SiC-Al system when they were heated to 973–1473 K during pressureless sintering. Li[20] found that needle-like Al₄C₃ and equiaxed Si were located at the interface between the Al matrix and SiC particles in Al12Si/SiC composites prepared by SLM method because the high laser absorptivity of SiC resulted in the high temperature and accelerated the reaction between SiC and Al.
Figure 1. XRD patterns of SiC/Al powders fabricated under different laser power. Al: JCPDS 00-004-0787; SiC: JCPDS 01-089-1975

Figure 2. OM micrographs of the polished top surface of SLM-processed SiC/Al samples using different laser power: (a) 300 W; (b) 350 W; (c) 400 W; (d) the densification behavior of SLM-processed SiC/Al samples using different laser power.
The optical micrograph (OM) of the polished top surfaces of the SLM-processed composites using various laser power are shown in Figure 2(a)–(c), and the corresponding relative densities are shown in Figure 2(d). As can be seen from the Figure 2(d), the density of the samples increased from 82.1% to 92.2% with increasing the laser power from 300 W to 400 W. At a relatively low laser power (300 W), irregular large pores of about 50 μm in diameter can be observed in the sample, suggesting that the input of energy is too little to supply enough energy to melt the powder. Increasing the laser power to 400 W, no obvious large pores were detected and the densification of sample was improved. Previous works revealed that these pores are formed because of insufficient wetting between metal matrix and ceramic particle[21]. With an increase in the laser power, both the surface tension and Marangoni flow gradient from the center to the periphery of the molten pool decrease. Therefore, the temperature of the molten pool increased with increasing the laser power and then more liquid phase can be produced to promote the effective spreading of the melt. As a result, the density of the sample was increased.

Microhardness of the SLM-processed samples was measured. Figure 3 shows the microhardness of the SLM-processed samples prepared at different laser power. It can be seen that all the SiC/Al samples exhibit much higher microhardness than the unreinforced Al sample, demonstrating that the addition of SiC particles can significantly improve the microhardness of Al bulk. Figure 3 also reveals that the samples show higher microhardness when they were prepared under stronger laser power. Zhao[22] reported the similar phenomenon and proved the improved microhardness may be caused by the less defects, i.e., smaller pore size, less pores and high interface bonding strength between SiC particle and Al matrix.

![Figure 3. The relationship between microhardness and laser power of the SLM-processed sample.](image)

Figure 4 shows the flexural strength of the SLM-processed material under different laser power, indicating that the flexural strength increased from 253 MPa to 273 MPa with increasing the laser power from 300 W to 400 W. This enhancement may be dependent on the microstructure of the samples[23]. Under a laser power of 300 W, a large amount of unmelted Al powder as well as various pores existed in the specimen. With increasing the laser power to 350 W, no unmelted Al particles were observed. At the same time, numerous pits and dimples were found in the SLM-processed sample, indicating that the fracture mode of SLM-processed samples is interfacial debonding and matrix fracture[24]. The dimples should be caused by the void nucleation, growth and subsequent
coalescence during the strong shear deformation and fracture process on the shear plane[25]. Further increased the laser power to 400W, no significant change in microstructure was observed.

Figure 4. The flexural strength of the fabricated samples.

Figure 5 shows the effect of the laser power on the friction coefficient of the SLM-processed samples. Figure 5(a) shows that a large amount of wear debris appears on the worn surface of the SLM-processed SiC/Al sample under the laser power of 300 W, indicating that the main wear mechanism is stripping and abrasion. With increasing the laser power to 350 W, a large number of furrows parallel to the sliding direction and less wear debris were found on the worn surface of the SLM-processed SiC/Al sample, demonstrating that the main wear mechanism of the SLM-processed SiC/Al samples changed to plough wear. Keep on increasing the laser power to 400 W, no significantly difference can be observed on the worn surface of the SLM-processed SiC/Al sample. On the other hand, the friction coefficient of the SiC/Al samples decreased from 0.52 to 0.41 with increasing the laser power from 300 W to 400W. This enhancement may be caused by the improved hardness and densification of the as-prepared samples (Figure 2(d), Figure 3).

4. Conclusions
SiC/Al composites were fabricated by SLM technique under different laser power. The dependence of microstructures, microhardness and flexural strength of formed SiC/Al composites on laser power was studied. The results revealed that:

(1) The densification behavior can be improved by increasing laser power. The densification of the as-prepared SiC/Al composite increases from 82.1% to 92.2% with increasing the laser power from 300 W to 400 W.

(2) The microhardness of the as-prepared SiC/Al composite is enhanced with increasing laser power and the maximum microhardness reaches 152.6 HV0.1.

(3) With increasing the laser power to 400 W, the fracture mode changes from the matrix ductile fracture to the reinforcement cleavage fracture and the flexural strength increases from 253 MPa to 276 MPa.

(4) The friction coefficient is closely related to the laser power. As the laser power increases, the friction coefficient decreases from 0.52 to 0.41, and the wear mechanism changes from delamination wear to plough wear.
Figure 5. The morphologies of the worn surfaces of SLM-processed composites and correspondent friction coefficient with laser power. (a-b) 300 W; (c-d) 350 W; (e-f) 400 W.

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