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Microstructure and wear behavior of *in-situ* hypereutectic Al–high Si alloys produced by selective laser melting

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In this study, almost dense hypereutectic Al-50 Si alloys were fabricated *in situ* by selective laser melting (SLM) from the mixture of Al and Si powders under argon atmosphere. The effects of laser power on the microstructure and mechanical properties were determined regarding the size and morphology of primary Si phase (proeutectic Si). The phase analysis shows that the main phase of SLM-processed Al-Si alloys consists of Al and Si for all samples. Additionally, due to the vaporization behavior of Al, the Si content increases after melting process. The primary Si phase presents a smaller size (about 5 μm) than that in the conventional produced hypereutectic Al-Si alloys. Moreover, while the laser power increases, the spherical morphology of primary silicon turns into irregular shape. Besides, the maximum value of microhardness (188 Hv) was detected for the samples obtained at laser power of 320 W. The sample obtained at laser power of 350 W shows the lowest wear rate (5.5×10^{-4} mm^3 N^{-1} m^{-1}).

1. Introduction

Hypereutectic Al-Si alloys have shown extensive commercial applications in aircraft, automotive and electronic packaging industries as thermal management materials, due to their good castability, high thermal conductivity, low thermal expansion coefficient, low density and high mechanical properties [1]. However, the conventional casted alloys always show a coarse microstructure of primary silicon phase (proeutectic silicon) as a consequence of low cooling rate, leading to a poor mechanical properties and workability. Liu et al. [2] reported that the fine, spherical and uniform distribution of primary silicon phase contributes to improve the mechanical properties of Al-Si alloys. Hanna et al. [3] and Li et al. [4] indicated that adding a modification agent, such as Ce, Sr and P, would be an effective way to improve the mechanical properties by refining the primary silicon phase. However, the study of Hogg and Atkinson [5] showed that the addition of a modification agent is not sufficient to refine the primary silicon phase when the Si content is over 25 wt.%. Another way to refine the microstructure of hypereutectic Al-Si alloys is rapid solidification [6], which exhibits a reduction of cost and no limits of composition. Among rapid solidification methods, the spray forming process is used to fabricate commercial hypereutectic Al-high Si alloys with very fine primary Si phase [7], such as, CE7 (Si-30Al), CE11 (Si-50Al), CE17 (Al-30Si).

Regarding the technologies mentioned above, the unavoidable extra machining process to obtain a desired part leads to an increasing of cost and time consumption. As a new additive manufacturing technology, selective laser melting (SLM) presents great potential applications in the fabrication of complex parts with fine microstructure [8]. In this process, a three dimensional part is designed with the computer aided design (CAD) software, then it is built layer-by-layer in which the computer controlled laser beam selectively melts the powder using the suitable parameters. Until now, common metallic materials, such as Fe [9], Al [10] and Ti [11] are already manufactured by SLM process. Especially, lightweight metals such as Al, Ti and their alloys, which are widely used in aeronautics and astronautics industry, are expected to be produced by SLM process. E.O. Olakanmi et al. [12] summarized the producing possibility of pure Al, Al-Mg alloy and Al-Si alloy by SLM and indicated that the Al-Si alloy was hereby proved as a suitable candidate material for SLM, due to a low thermal expansion and uniform distribution of its surface oxide film. Additionally, compared with the conventional pre-alloyed feedstock powders, the powder mixture presents high economic properties and high composition flexibility. However, so far few works focus on synthesizing hypereutectic Al-Si alloys by selective laser melting from powder mixture. Current studies show that Al-Si alloy with high Si content shows a similar coefficient of expansion compared to semiconductor materials, but has poor mechanical properties. Regarding alloys with high content of Al, on the contrary, satisfactory mechanical properties and poor thermal properties appear. Thus, according to previous study results [13,14], Al-50Si (wt.%) alloy was chosen as feedstock powder, which was a compromise in terms of thermal and mechanical properties.

In this work, the Al-0Si (wt.%) alloy was manufactured by SLM using a mixture of Al and Si powders under argon protective atmosphere. Afterwards, the microstructure, microhardness and wear behavior of the

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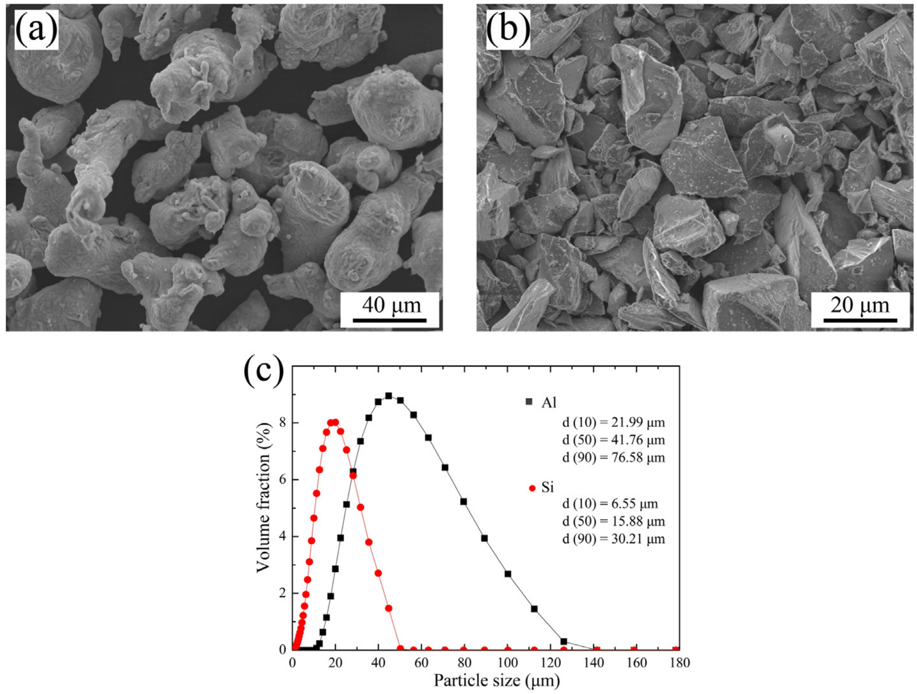

Fig. 1. SEM images of the morphology of powders (a) aluminum, (b) silicon and (c) their size distributions.

SLM processed samples were characterized with an emphasis on the size and morphology of primary silicon phase.

2. Materials and methods

The pure Al (XINGRONGYUAN, China) that was produced by gas atomization in argon atmosphere (converging - diverging flow duct process) and irregular pure Si powder (KAIMAN, China), were blended with weight proportion of Al: Si = 50: 50 in a tumbling mixer for 60 min and dried at 80 °C for 4 h before use. The morphology overviews of Al and Si powder are presented in Fig. 1 (a) and (b). The size distributions of Al and Si particle are shown in Fig. 1 (c), which illustrates that the average particle sizes of Al and Si are 41.76 μm and 15.88 μm respectively.

A commercial apparatus MCP-realizer SLM 250 (MCP-HEK Tooling GmbH, Germany) was used in this study. The laser source is YLR-100-SM single-mode CW ytterbium fiber laser (1064–1100 nm). The spot size and maximum laser power are 40 μm and 400 W. The layer thickness, laser scanning speed and hatch distance were fixed at 50 μm, 500 mm/s and 45 μm respectively. The laser power ranged from 260 W to 360 W under a high-purity argon atmosphere containing oxygen <0.2%. The zigzag laser trajectory is shown in Fig. 2 (a). The pure Al substrate was sandblasted before manufacturing and heated to 400 K during the SLM process. The small SLM fabricated cube (8 × 8 × 8 mm) samples are shown in Fig. 2 (b).

The image analysis method (NIH Image J, Software, USA) was used to determine the porosity, whose average values corresponded to measurements of 10 images for 3 times, and a size distribution of primary silicon phase by considering more than 300 particles. The microstructure of the specimens was observed using optical microscopy (OM) (Nikon, Japan) and scanning electron microscopy (SEM) (JEOL JSM 7800F, Japan) equipped with X-ray energy dispersive spectroscopy EDS. X-ray Diffraction (XRD) was performed with a Cobalt anticathode (λ = 1.78897 Å) and operated at 35 kV and 40 mA. A Leitz-Wetzlar (Germany) instrument was used to measure Vicker microhardness with load of 2.94 N and dwelling time of 25 s on polished samples. Microhardness was performed at the positions which are indicated in Fig. 2 (a). Unlubricated ball-on-disk wear test was performed with CSM-TRIBOMETER instruments (Switzerland) under ambient atmosphere on polished surface. The counterpart was a 3 mm diameter Al_2O_3 ball with a mirror finished surface. The normal load, rotation diameter and sliding velocity were 5 N, 4 mm and 20 mm/s respectively. Wear rate

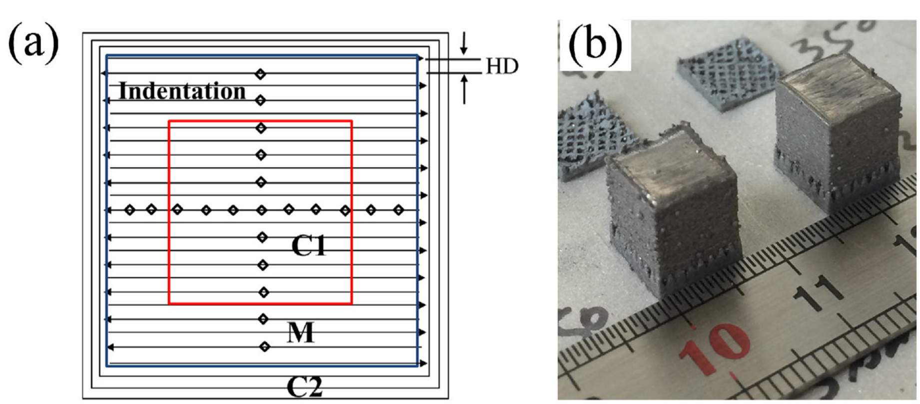

Fig. 2. the schematic illustration of (a) laser trace mode with indenter (C1 = Center, M = Middle, C2 = contour) and (b) the SLM-processed samples.