

Effect of Particle Size on Mechanical Properties and Fracture Behaviors of Age-Hardening SiC/Al–Zn–Mg–Cu Composites

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Abstract

15 vol.% SiC/Al-6.5Zn-2.8 Mg-1.7Cu (wt%) composites with varying particle sizes (3.5, 7.0, 14 and 20 µm), *i.e.*, C-3.5, C-7.0, C-14, and C-20, respectively, were fabricated by powder metallurgy (PM) method and subjected to microstructural examination. The effect of particle size on mechanical properties and fracture behaviors of the T6-treated composites was revealed and analyzed in detail. Element distribution and precipitates variations in the composites with varying particle sizes were emphatically considered. Results indicated that both tensile strength and plasticity of the T6-treated composites increased first and then decreased with particle size decreasing. The C-7.0 composite simultaneously exhibited the highest ultimate tensile strength (UTS) of 686 MPa and best elongation (EL) of 3.1%. The smaller-sized SiC particle would introduce more oxide impurities, which would react with the alloying element in the matrix to cause Mg segregation and depletion. According to strengthening mechanism analysis, the weakening of precipitation strengthening in the T6-treated C-3.5 composite was the main cause of the lower tensile strength. Additionally, the larger SiC particle, the more likely to fracture, especially in the composites with high yield strength. For the T6-treated C-20 composites, more than 75% SiC particles were broken up, resulting in the lowest plasticity. As decreasing particle size, the fracture behaviors of the T6-treated composites would change from particle fracture to matrix alloy fracture gradually.

Keywords Metal matrix composites \cdot Mechanical properties \cdot Particle size \cdot Fracture behaviors \cdot Strengthening mechanisms

1 Introduction

Owing to the superiorities of high modulus, low density, high strength, and low cost-effectiveness, SiC particles were usually introduced into the Al alloys matrix to manufacture the particle-reinforced aluminum matrix composites with excellent mechanical properties [1–6]. In recent years, as a promising candidate material for aerospace structural components and weapons manufacturing, SiC particle-reinforced Al–Zn–Mg–Cu composites attracted much attention [7–9].

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² School of Materials Science and Engineering, University of Science and Technology of China, Shenyang 110016, China However, the poor plasticity of the SiC/Al–Zn–Mg–Cu composites still was difficult to be improved, which limited the application of the composites.

Among multiple factors that affected the mechanical properties of the SiCp/Al composites, SiC particle size played a critical role [10–12]. In previous studies, the composites with smaller-sized particles were more likely to be densified [13], but the small-sized particles were difficult to distribute homogeneously in the aluminum matrix [14-16]. In addition, the SiC particle size could also affect the grain size of the matrix, which determined the grain boundary strengthening effect [17]. According to the Hall–Petch relationship, the smaller grain size, the higher strength of the composites [18]. In recent years, the effect of SiC particle size on the thicknesses of geometrically necessary dislocation (GND) zones surrounding particles also was investigated by transmission electron microscopy (TEM) and nanoindentation [17]. For a composite with a constant SiC particle volume fraction, the total volume of GNDs was increased with decreasing particle size, resulting in higher

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strength of the matrix alloy [17, 19]. In the SiC/Al, SiC/Al-Cu-Mg, and SiC/Al-Mg-Si composites [14, 17, 20], it was generally believed that with the decrease in particle size, the strength of the composites was enhanced, while the plasticity was reduced.

However, in the abovementioned researches, the precipitated phases in the composites with varying particle sizes were rarely discussed. For the Al-Zn-Mg-Cu alloys, precipitation strengthening was the most crucial strengthening mechanism to determine the tensile strength of the alloys. By peak-aging treatment, the yield strength of the Al-7.5Zn-1.8 Mg-1.7Cu alloy could be enhanced by 130% [21]. Thus, the effect of SiC particle size on precipitation strengthening of SiC/Al-Zn-Mg-Cu composites had to be considered. In the T6-treated Al-Zn-Mg-Cu alloys, the nanophase η' (MgZn₂) was the main strengthening phase, and η $(MgZn_2)$, T $(Al_2Mg_3Zn_3)$, S (Al_2CuMg) and θ (Al_2Cu) might also exist [22, 23]. The type and amount of aging-induced precipitates strongly depend on the chemical composition of the alloy (particularly the Zn/Mg ratio [24]). However, in the SiC/Al–Zn–Mg–Cu [21] composites, SiC/Al interface reaction was unavoidable. Mg could react with the oxide impurities on the SiC particle surface to form MgO, which would cause Mg depletion in the matrix and Mg segregation around SiC particles. The capability of precipitation strengthening of T6-treated SiC/Al-Zn-Mg-Cu composites was significantly reduced. For the SiC/Al-Zn-Mg-Cu composites with a constant SiC particle volume fraction, the smaller SiC particle size, the more SiC/Al interfaces. Thus, it was reasonable to speculate that SiC particle size could affect Mg content in the matrix and precipitation strengthening effect.

Particle fracture, interfacial debonding, and matrix fracture were the main fracture form of particle-reinforced aluminum matrix composites [16]. Under the premise of ensuring good interface bonding, both experimental and simulation results showed that the larger the particle size, the more obvious particle fracture [6, 20]. Particle fracture was a typical cleavage fracture. Once the crack was formed, it would rapidly expand to the matrix. Lots of particle fractures would not only reduce the strength but also seriously reduce the plasticity of the SiC/Al-Zn-Mg-Cu composites. The yield strength of most Al-Zn-Mg-Cu matrix alloys was higher than that of pure Al, Al-Cu-Mg, and Al-Mg-Si alloys [2, 25–27], so particle fracture could be more obvious in the SiC/Al-Zn-Mg-Cu composites. In order to better understand the particle size effect in the SiC/Al-Zn-Mg-Cu composites, it was necessary to study the effect of particle size on the fracture behaviors of the composites.

In this study, 15 vol.% SiC/Al-6.5Zn-2.8 Mg-1.7Cu (wt%) composites with varying particle sizes were fabricated by powder metallurgy (PM) technique. Mechanical properties and fracture behaviors of the T6-treated and

annealed composite were compared. The aim is to (a) understand the effect of SiC particle size on the microstructures and mechanical properties of the age-hardening SiC/ Al–Zn–Mg–Cu composites and (b) improve mechanical properties of SiC/Al–Zn–Mg–Cu composites by adjusting SiC particle size.

2 Experimental

2.1 Fabrication of SiC/Al-Zn-Mg-Cu Composites

15 vol.% SiC/Al-Zn-Mg-Cu composites with varying particle sizes (3.5, 7.0, 14, and 20 µm) were fabricated using the PM technique. Depending on the reinforcing particle size, the as-fabricated samples were hereafter denoted as C (composite)-3.5, C-7.0, C-14, and C-20, respectively. The nominal composition of the matrix alloy was Al-6.5Zn-2.8 Mg-1.7Cu (wt%). The mean size of alloy powder (99.9% purity) was 13 µm. The SiC particles were mixed with the alloy powders for 7 h using a bi-axis rotary mixer with a rotation speed of 50 rpm. The mixed powders were degassed in a vacuum furnace under 10⁻² Pa and sintered at 540 °C for 2 h and then hot pressed into billet in a die under a pressure of 50 MPa. The diameter of the billets is 75 mm, and the height is about 70 mm. The billets were extruded at 420 °C with the extrusion ratio at 17:1 to get the extruded bars, which were 18 mm in diameter. The extrusion speed was 1 mm s^{-1} .

A part of the extruded bars was solid solution treated at 470 °C for 2 h and subsequently quenched into the cool water and then subjected to aging treatment at 120 °C for 24 h, *i.e.*, T6 treatment. Another part of the extruded bars was annealed at 450 °C for 2 h and then cooled with the furnace to room temperature.

2.2 Mechanical Property Tests

Tensile specimens were machined from the extruded bar with the tensile axis parallel to the extrusion direction. Tensile specimens 5 mm in diameter and 30 mm in gauge length were tested on an Instron 8801 tester at a strain rate of 1×10^{-3} s⁻¹ and room temperature. At least 3 tensile specimens for each heat treatment condition were measured to calculate the average values.

2.3 Characterization of Microstructure and Microchemistry

According to Archimedes' principle, the densities of the asextruded composites were measured by Analytical Balance (METTLER TOLEDO). The microstructures were observed using optical microscopy (OM, Leica). The metallographic specimens were ground with 2000-grit abrasive paper, mechanically polished, and then eroded by Keller's etchant. Phases in the composites were identified using an X-ray diffractometer (XRD, D/max 2400). Thermodynamic calculations were conducted using Thermo-calc software and its thermodynamic database. The calculations were based on Al-6.5Zn-*x*Mg-1.7Cu alloy (the content of Mg is in the range of 1.0–4.0 wt%). The weight percentages of η (MgZn₂), T (AlZnMgCu), S (Al₂CuMg), and θ (Al₂Cu) phases were calculated by equilibrium model.

Chemical compositions in the matrix were analyzed by a wave dispersive spectrometer (WDS). The element distributions in the composites were examined by electron probe microanalysis (EPMA-8530F). The interface microstructure and precipitated phases were examined by transmission electron microscopy (TEM, TECNAI G2 F20). The thin foils for TEM were mechanically polished and ion-milled. The fracture surface of the tensile samples was observed by scanning electron microscopy (SEM, quanta 600).

3 Results and Discussion

3.1 Phase Analysis

Figure 1a shows XRD patterns of the SiC particles with varying particle sizes. Apart from SiC, the major impurities in the SiC particles, free Si and SiO₂, also were detected clearly. As particle size increased, the diffraction peak intensity of impurity phases gradually decreased. It implied that impurity content could be lower in the larger-sized SiC particle.

In order to verify the phase composition of the bulk composites, XRD patterns of the as-extruded SiC/Al–Zn–Mg–Cu composites with varying particle sizes are shown in Fig. 1b. As the main precipitated phase of SiC/Al–Zn–Mg–Cu composites, MgZn₂ was detected clearly. The diffraction peak intensity of the MgZn₂ phase in the C-20, C-14, and C-7.0 composites was similar and much more obvious than that in the C-3.5 composite. Furthermore, although the diffraction peak intensity of the Mg₂Si phase was very weak, it still could be identified. During sintering, Mg in the alloy matrix could react with Si and SiO₂ to form Mg₂Si [28]. In previous studies, Mg could also react with Al and SiO₂ to form MgO or MgAl₂O₄ [29]. In this study, the content of Mg-containing oxide might be too low to be detected. In any case, those chemical reactions would consume Mg atoms and reduce the content of MgZn₂.

3.2 Microstructures of the Composites with Varying Particle Sizes

Table 1 shows the densities of the as-extruded composites with varying particle sizes. All the composites were densified, and no porosities were observed. Figure 2a, c, e, and g shows microstructures of the T6-treated C-20, C-14, C-7.0, and C-3.5 composites, respectively. The composites all exhibited typical as-extruded microstructures. As demonstrated, the SiC particle distribution in the C-20, C-14, and C-7.0 composites was fairly homogenous. In contrast, the SiC particles in the C-3.5 composite were inclined to agglomerate, as evidenced by their local clustering and the extrusion traces (Fig. 2 g). This observation was consistent with the conventional notion that it would become more

Table 1 Measured density and relative density of the as-extruded composites

Materials	Measured density (g ⁻ cm ⁻³)	Theoretical density (g cm ⁻³)	Relative density (%)
C-3.5	2.864	2.864	100.00
C-7.0	2.862		99.93
C-14	2.862		99.93
C-20	2.861		99.90



Fig. 1 XRD patterns of a as-received SiC particles and b as-extruded SiC/Al-Zn-Mg-Cu composites with varying particle sizes



Fig. 2 OM images and grain size statistic plots of a, b C-20, c, d C-14, e, f C-7.0, g, h C-3.5 composites under T6-treated condition

difficult to homogeneously distribute reinforcing particles in the metal matrix with decreasing particle size [14]. As shown in Fig. 2b, d, f, and h, the matrix mean grain size in the C-20, C-14, C-7.0, and C-3.5 composites was estimated to be 4.7, 4.4, 3.0, and 2.3 µm, respectively, showing obvious dependence on the SiC particle size. In the sintering process, the SiC particle could inhibit grain growth at a high temperature [30]. The spacing of smaller-sized SiC particles was smaller, which could more significantly inhibit grain boundary motion. What is more, during extrusion deformation, due to the inconsistency between particles and matrix deformation, a large number of dislocations that could induce the formation of new grain boundary could be formed around SiC particles [31]. The smaller SiC particle size, the higher dislocation density [17]. Thus, as particle size decreased, the matrix grain size would gradually decrease. The increase in yield strength (YS) due to the grain boundary strengthening $\Delta \sigma_{\rm gb}$ could be given by the Hall–Petch equation [32]:

$$\Delta \sigma_{\rm gb} = k_{\rm v} d^{-1/2},\tag{1}$$

where *d* is the mean grain size and k_y is the Hall–Petch coefficient with the value of ~0.12 MPa m^{-1/2} [21, 32] for Al-Zn-Mg alloys. From Eq. (1), $\Delta \sigma_{\rm gb}$ of the C-20, C-14, C-7.0, and C-3.5 composites could be calculated to be about 55, 57, 69, and 79 MPa, respectively. The results indicated that the smaller-sized SiC particle was more conducive to grain boundary strengthening.

Figure 3 shows elemental distribution maps of the T6-treated composites with varying particle sizes. In all the composites, the distribution of zinc element was quite uniform. Meanwhile, the Cu element also was homogeneously distributed in the C-20 and C-14 composites. With decreasing particle size to 7.0 and $3.5 \,\mu\text{m}$, Cu-rich phases could be observed in the C-7.0 and C-3.5 composites. The Cu-rich phase was generally considered as Al₇Cu₂Fe formed during the sintering process [21]. Fe could come from mixing balls and jars. Due to the small size and low content, the effect of Cu-rich phases on the mechanical properties could be ignored. In addition, Mg and O elements both showed considerable agglomeration in all the composites, and their enrichment sites were almost identical, preferentially close to the SiC/Al interface. As expected, as particle size decreased, Mg and O element segregation was more obvious. It was reasonable to speculate that Mg segregation was caused by the chemical reaction between the matrix and the impurities introduced by SiC particles. According to XRD results (Fig. 1a), the impurities in the smaller-sized SiC particles were more, so element segregation in the composites with small-size particles was more obvious (Fig. 30).

Chemical compositions in the homogeneous region of alloying elements were disclosed by wave dispersive spectrometer (WDS), and test results are listed in Table 2. The results indicated that the actual content of Zn and Cu elements in the matrix region was close to the nominal content. However, for the composites with varying particle sizes, there were obvious changes in Mg content. The measured Mg content in the matrix region of the C-20, C-14, C-7.0, and C-3.5 composites was 0.6 wt%, 0.4 wt%, 0.7 wt%, and 1.4 wt% lower than the nominal content, respectively, which was likely to cause variation of equilibrium phase components in the composites. Thus, the thermodynamic calculation was analyzed, and the results are shown in Fig. 4. The mass fraction of each phase was calculated by Thermo-calc software. The contents of Zn and Cu were fixed, 6.5 and 1.7 wt%, respectively. When the content of Mg was between 1.0 wt% and 3.0 wt%, the equilibrium phase was mostly MgZn₂, with a small amount of θ (Al₂Cu), S (Al₂CuMg) or T (AlZnMgCu). With increasing Mg content from 1.0 wt% to 4.0 wt%, the θ phase disappeared gradually, and the S and T phase started to form. The content of MgZn₂ first increased and then decreased after reaching a maximum value. Although the nominal content of Mg element in all the composites was 2.8 wt%, the actual Mg content in the C-3.5, C-7.0, C-14, and C-20 composites was measured as 1.4 wt%, 2.1 wt%, 2.4 wt%, and 2.2 wt% by WDS (Table 2), respectively, resulting in different phase content. As shown by the dotted line in Fig. 4, the content of MgZn₂ in the C-7.0, C-20, C-14, and C-3.5 composites decreased in turn. Both thermodynamic calculation and XRD analysis results (Fig. 1b) indicated that the content of MgZn₂ in the C-3.5 composite was the lowest. It was necessary to investigate the variation of the morphology and density of the precipitated phase in the composites with varying particle sizes.

In the T6-treaded SiC/Al-Zn-Mg-Cu composites, the disk-liked η ' phase was the main precipitates, which was characterized by TEM (Fig. 5(a, b, c and d)). The mean length of the precipitates in the C-20, C-14, C-7.0, and C-3.5 composites was measured as 4.8, 5.4, 5.8, and 6.2 nm, respectively. During the aging process, after nucleation, the η phases preferentially grow along the diameter direction and almost remain constant width, resulting in the platelet shape of the precipitated phase [33]. Due to different orientation relationships with the aluminum matrix, the projections of η ' phases in [110]_{A1} planes could exhibit plate-like or rod-like (Fig. 5e). As decreasing SiC particle size, the rod-liked phases were easier to be observed, speciously in the C-3.5 composite (Fig. 5d). This finding could be interpreted by the size variation of the η ' phase. With the growth of the η ' phase, the difference in diameter and width of the η ' phase would be increased. In the bright-field TEM image of the C-3.5 composite (Fig. 5d), the rod-liked phases that tended to be black were more visible than the plate-liked phases that tended to be gray.

In this study, according to WDS results (Table 2), by decreasing particle size, Zn content remained constant,



Fig. 3 Elemental maps of a-d C-20, e-h C-14, i-l C-7.0, m-p C-3.5 composites under T6-treated condition

Table 2 Chemical compositions in the matrix region of the SiC/Al–Zn–Mg–Cu composites with varying particle sizes determined by WDS (wt%)

Materials	Al	Zn	Mg	Cu
C-20	89.5 <u>+</u> 0.4	6.5 ± 0.1	2.2 ± 0.1	1.8 ± 0.1
C-14	89.4 ± 0.5	6.5 ± 0.1	2.4 ± 0.2	1.7 ± 0.1
C-7.0	89.7 ± 0.4	6.5 ± 0.1	2.1 ± 0.2	1.7 ± 0.1
C-3.5	90.4 ± 0.5	6.6 ± 0.2	1.4 ± 0.1	1.6 ± 0.1

whereas Mg content would be reduced, i.e., Zn/Mg ratio was increased. A high Zn/Mg ratio was conducive to the growth of the precipitates. Shu [34] investigated the microstructures of high-strength Al–Zn–Mg–Cu alloys with different Zn/Mg ratios in detail and found that the size of precipitates was larger in the alloy with a higher Zn/Mg ratio. The same conclusion was found in the SiC/Al–Zn–Mg–Cu composites [21]. Meanwhile, the density of dislocations in the composite with smaller-sized SiC particles was higher [17]. Dislocations could accelerate the diffusion of solute



Fig. 4 Relationships between equilibrium phase components and Mg content in the composite with constant Zn and Cu contents

atoms and the growth of the precipitates [21]. Thus, the size of precipitates in the C-3.5 composite was larger than that in the C-7.0, C-14, and C-20 composites (Fig. 5). Additionally, according to XRD analysis and thermodynamic calculation results (Fig. 1b and Fig. 4), the content of MgZn₂ in the

C-3.5 composite was lower, so the precipitates in Fig. 5d were slightly sparser.

For the T6-treated SiC/Al–Zn–Mg–Cu composites, the strengthening mechanisms from the precipitates, $\Delta \sigma_{or}$, can be expressed as follows [35, 36]:

$$\Delta\sigma_{\rm or} = \frac{0.4G{\rm Mb}}{\pi\sqrt{1-\nu}} \frac{\ln(1.63r/b)}{\lambda_p},\tag{2}$$

where *M* is the mean orientation factor, 3.06, *G* is Shear modulus, 26.9, *b* is Burgers vector, 0.286 nm, ν is the Poisson's ratio, 0.33 [35], *r* is the mean radius of precipitates, and λ p is the mean precipitate spacing, 16, 16, 17, and 20 nm, corresponding to the C-20, C-14, C-7.0, and C-3.5 composites, respectively. For simplified calculation, the precipitates could be supposed to be spherical, $r=l/\pi$, where *l* is the mean length of the precipitates. From Eq. (2), $\Delta\sigma_{or}$ of the C-20, C-14, C-7.0, and C-3.5 composites could be estimated to be about 495, 464, 507, and 443 MPa, respectively. Obviously, the precipitation strengthening effect in the C-3.5 composite was much lower than that in the C-7.0 composite.

Figure 6a shows the representative TEM image of dislocations adjacent to the SiC/Al interface in the T6-treated C-3.5 composite. As expected, a high density of dislocations was



Fig. 5 Bright-field TEM images of a C-20, b C-14, c C-7.0, d C-3.5 under T6-treated condition taken along [110] Al zone axis, and e spatial distribution schematic diagram of precipitated phase

Fig. 6 Microstructures of the T6-treated C-3.5 composite: a dislocations adjacent to the SiC particle and b,c SiC/Al interface reaction layer; d,e the enlarge views of white rectangles in c; f,g FFT images of d,e, respectively

observed in this region, which was likely to cause strengthening of the matrix. Presently, although geometrically necessary dislocations (GNDs) around the SiC with varying particle sizes were investigated experimentally by TEM analysis and nanoindentation tests [17], applying these results for interpretation and evaluation of the dislocation punched zones still was not rigorous, since not all GNDs could be observable under the specific diffraction condition and grain size also was a crucial factor affecting dislocation extension [37, 38]. Thus, in this work, the effect of particle size on GNDs strengthening was studied mainly by theoretical analysis. The contribution by the GND strengthening $\Delta \sigma_{gnd}$ could be estimated by the following equation [39]:

$$\Delta \sigma_{\rm gnd} = \beta G b \sqrt{\frac{12\sqrt{2}Vp\Delta \text{CTE}\Delta T}{(1-Vp)bdp}},\tag{3}$$

where β is Taylor's coefficient with the value of 1.25, Vp is the volume fraction of SiC particle, Δ CTE is the difference between the coefficient of thermal expansion of matrix alloy and SiC particle, 19.5×10^{-6} K⁻¹, dp is the average size of SiC particle, and ΔT is the difference between processing and test temperature, 450 K. The mismatch dislocation strengthening values in the C-20, C-14, C-7.0, and C-3.5 composites were estimated to be 20, 24, 34, and 49 MPa, respectively. The results indicated that the smaller-sized SiC particle was more conducive to mismatch dislocation strengthening.

Figure 6b shows the microstructure of the SiC/Al interface in the T6-treated C-3.5 composite. The obvious interface reaction was found, and the thickness of the interfacial layer was about 50–120 nm. The interface reaction products were analyzed with high-resolution transmission electron microscopy (HRTEM) (Fig. 6ce). With combined HRTEM and fast Fourier transform (FFT) analysis (Fig. 6e, g), the continuous interface products were identified as face-centered cubic MgO, which could be formed by the interface reaction between Mg atom and SiO₂ impurity on the surface of SiC particle during sintering [21]. MgO had a crystallographic orientation relationship with the 6H SiC with the typical hexagonal structure, (2-1-0)1, (2-1-0)1, (2-1-0)12-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0)1, (2-1-0) $(2-1-0_{SiC}||(011)_{MgO}$ (Fig. 6d, f). On the surface of asreceived SiC particles, the oxide impurities were nonuniform, resulting in nonuniform interface reaction products. Under the same preparation process, for the SiC with varying particle sizes, the amount of Mg consumption strongly depended on the content of impurities on the surface of SiC particles. According to XRD analysis of SiC particles (Fig. 1a) and WDS results (Table 2), the content of impurities in the 3.5 µm SiC particles was the most, and Mg content decrease was the most obvious in the C-3.5 composite.

3.3 Mechanical Properties of the Composites with Varying Particle Sizes

Table 3 Tensile properties of the SiC/Al–Zn–Mg–Cu composites under T6-treated and annealed conditions

Stress-strain curves of the T6-treated SiC/Al–Zn–Mg–Cu composites with varying particle sizes are shown in Fig. 7a, and tensile properties are listed in Table 3. Generally, in the SiC/Al, SiC/Al-Cu-Mg, and SiC/Al–Mg-Si composites [14, 17, 39], as decreasing particle size, the strength of the composite was enhanced, while the plasticity of the composite with larger-sized particle was better. However,

quite surprisingly, the rule of the particle size effect in the SiC/Al-Zn-Mg-Cu composites was completely different. The ultimate tensile strength (UTS: ~652 MPa) and yield strength (YS: ~583 MPa) of the T6-treated C-3.5 composite both were lower than that of the T6-treated C-7.0 composite. Meanwhile, the T6-treated C-20 composite had the lowest elongation (El.: ~ 1.4%). 7.0 µm was the fittest size of reinforcing particles to fabricate the SiC/ Al-Zn-Mg-Cu composites with excellent mechanical properties (UTS: ~ 686 MPa, YS: ~ 618 MPa, El.: ~ 3.1%). In order to weaken the effect of element distribution and precipitated phases on the mechanical properties of the SiC/Al-Zn-Mg-Cu composites, the annealed composites were also studied. The results of tensile tests are shown in Fig. 7b and listed in Table 3. The annealed C-3.5 composite had the highest UTS (299 MPa), and the excellent plasticity could be obtained in the annealed C-14 and C-20 composites. As decreasing particle size, the strength of the composite was enhanced, and the plasticity was decreased. This rule was consistent with previous studies [14, 17, 39].

The strengthening mechanisms in the T6-treated composites included grain boundary strengthening, solid solution strengthening, precipitation strengthening, dislocation strengthening, and load transfer strengthening [21], while precipitation strengthening in the annealed composites would not be considered. After aging and annealing, lots of solute atoms were precipitated, so solid solution strengthening could also be ignored.

Fig. 7 Stress-strain curves of the SiC/Al-Zn-Mg-Cu composites under a T6-treated and b annealed condition

Materials	T6 treated		Annealing treated			
	YS (MPa)	UTS (MPa)	El. (%)	YS (MPa)	UTS (MPa)	El. (%)
C-3.5	583 <u>+</u>	4 652	$\pm 42.3 \pm 0.1$	165 -	± 2 299 :	$\pm 1 6.5 \pm 0.1$
C-7.0	618 <u>+</u>	1 686	$\pm 13.1 \pm 0.2$	135 ±	<u>1 260</u>	$\pm 2 \ 8.1 \pm 0.5$
C-14	616 <u>+</u>	3 657	$\pm 52.9 \pm 0.2$	120 ±	±2 240 :	$\pm 1\ 10.2 \pm 0.3$
C-20	602 <u>+</u>	1 629	$\pm 11.4 \pm 0.2$	110 -	<u>1 229</u>	$\pm 1 \ 9.2 \pm 0.1$

Assuming that all the SiC/Al interfaces were well bonded, the strengthening due to the load transfer effect $\Delta \sigma_{l-t}$ could be estimated by the following equations [40]:

$$\Delta \sigma_{1-t} = S/2V p(\sigma_{\rm m} + \sigma_0), \tag{4}$$

$$\sigma_{\rm m} = \Delta \sigma_{\rm gb} + \Delta \sigma_{\rm or} + \Delta \sigma_{\rm gnd},\tag{5}$$

where *S* is the aspect ratio of particulates with the value of 2.5, σ_m is the increment of YS from the matrix, and σ_0 is the YS of the pure Al in single crystal, $\sigma_0 = 17$ MPa. From Eq. (4), $\Delta\sigma_{l-t}$ of the T6-treated C-20, C-14, C-7.0, and C-3.5 composites could be calculated to be about 110, 105, 117, and 110 MPa, respectively.

In the T6-treated composites, as decreasing particle size from 7.0 µm to 3.5 µm, the values of $\Delta \sigma_{\rm gb}$ and $\Delta \sigma_{\rm gnd}$ were increased by 10 and 15 MPa, respectively, whereas the values of $\Delta \sigma_{\rm or}$ and $\Delta \sigma_{\rm l-t}$ were reduced by 64 and 7 MPa, respectively. Clearly, the weakening of precipitation strengthening was the main reason for the strength reduction of the T6-treated C-3.5 composite. Meanwhile, partial particle clustering could further deteriorate the mechanical properties of the T6-treated C-3.5 composite (Fig. 2g). The values of $\Delta \sigma_{gb}$, $\Delta \sigma_{or}$, $\Delta \sigma_{I-t}$ and $\Delta \sigma_{gnd}$ in the T6-treated C-20 composite were all lower than those in the T6-treated C-7.0 composite. It was easy to understand why the tensile strength of the T6-treated C-20 composite was lower. However, strengthening mechanism analysis could not explain why the plasticity of the composites with larger-sized SiC particles was lower. According to tensile test results (Table 3), the yield strength of the T6-treated composites, which was more conducive to SiC particle fracture during tensile deformation. Therefore, it was necessary to consider the effect of fracture behaviors on mechanical properties.

3.4 Fracture Behaviors of the Annealed and T6-Treated Composites

Three fracture phenomena such as deformation of matrix, SiC particle fracture, and interfacial debonding coexisted in the fractography of the annealed SiC/Al–Zn–Mg–Cu composites (Fig. 8). Due to the excellent plasticity of the annealed C-20 composite, dimples, and tear ridges of the matrix fracture were considerably obvious (Fig. 8a). Meanwhile, cleavage fractures of SiC particles and a few

Fig. 8 Fractographs of a C-20, b C-14, c C-7.0, d C-3.5 composites under annealed condition

interfacial debonding also were found. As decreasing particle size, the depth of dimples decreased, and particle fracture disappeared gradually (Fig. 8b, c, and d).

Fig.9 Statistic bar diagram of SiC particle failure modes in the annealed composites

To better understand fracture behaviors of the composites, SiC particle failure modes of the annealed composites were analyzed semi-quantitatively from a number of SEM images (more than 5 images for each specimen), and the average results are shown in the statistic bar diagram of Fig. 9. According to the observation results of the C-20 composite, more than 50% of the SiC particles tended to fracture, thereby resulting in relatively lower tensile strains below that of the C-14 composite. In the C-14 composite, the depth of dimples was similar to that in the C-20 composite, indicating a similar matrix deformation capacity (Fig. 8a, b). But the fraction of particle fracture in the C-14 composite was much lower, thereby resulting in more excellent plasticity. Due to the low YS of the annealed composites, the smaller-sized SiC particles were difficult to break up. For the annealed C-7.0 and C-3.5 composites, the ductile fracture of the matrix was the main fracture mechanism, and very few particle fractures occurred.

As shown in Fig. 10, the SiC particle fracture phenomenon in all the T6-treated composites was much more obvious than that in the corresponding annealed composites, especially in the T6-treated C-20 composite (Fig. 10a). More than 70% of the 20 μ m SiC particles were broken up

Fig. 10 Fractographs of a C-20, b C-14, c C-7.0, d C-3.5 under T6-treated condition

Fig. 11 Statistic bar diagram of SiC particle failure modes in the T6-treated composites

(Fig. 11). Such a high fraction of particle fracture would seriously deteriorate the plasticity of the composite. Due to the relatively higher YS of the T6-treated C-20 composite, the stress concentrated on the particles was high enough to cause particle fracture during the tensile deformation. Once cracks were formed inside the SiC particles, it would propagate to the matrix rapidly. The failed particles were no longer capable of bearing the load and equivalent to a hole, which was not favorable for enhancing the strength and plasticity of the composites. In this study, a large number of SiC particle fracture was the main reason for the low plasticity of the T6-treated C-20 composite. The plastic deformation capacity of the matrix in the T6-treated C-20 composite was not fully developed, and the dimples in the fractography were very shallow (Fig. 10a). For the T6-treated C-20 composite, cleavage fracture of the SiC particles was the main fracture mechanism. As decreasing particle size, the fraction of SiC particle cracking was decreased gradually, the fracture behavior of the T6-treated composite was transformed from particle fracture to matrix fracture.

4 Conclusions

1. With decreasing particle size, the uniformity of alloying element distribution decreased. Due to the chemical reaction between Mg and the impurities on the surface of SiC particles, Mg atoms were obviously clustered around SiC particles, which would inevitably cause alloying element depletion in the matrix. The measured Mg content in the matrix region of the C-20, C-14, C-7.0, and C-3.5 composites was 0.6 wt%, 0.4 wt%, 0.7 wt%, and 1.4 wt% lower than the nominal content, respectively.

2. For the annealed composites, as decreasing particle size, UTS and YS showed a moderate enhancement, and El. decreased gradually. However, the mechanical properties of the T6-treated composites had no monotonic change rule. The T6-treated C-7.0 composite exhibited optimum strength and plasticity simultaneously.

3. The strengthening mechanism analysis indicated that decreasing particle size could lead to the enhancement of grain-boundary strengthening and dislocation strengthening, and the reduction of precipitation strengthening. The weakened precipitation strengthening was the main reason for strength reduction in the T6-treated C-3.5 composite.

4. In the T6-treated SiC/Al–Zn–Mg–Cu composites, more than 70% of the 20 μ m SiC particles were broken up. Such a high fraction of particle fracture would seriously deteriorate the plasticity of the composite. Thus, the T6-treated C-20 composite exhibited the lowest elongation. As decreasing particle size, the fracture behavior of the T6-treated composite was transformed from particle fracture to matrix fracture.

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