Tensile and Fracture Properties of 15 vol% SiC_p/2009Al Composites Fabricated by Hot Isostatic Pressing and Hot Extrusion Processes

Junhui Nie · Jianzhong Fan · Shaoming Zhang · Shaohua Wei · Tao Zuo · Zili Ma · Zhaobing Xiang

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Abstract 15 vol% silicon carbide particle (SiC_p)-reinforced 2009Al matrix (15 vol% SiC_p/2009Al) composites were fabricated by hot isostatic pressing (HIP) and hot extrusion processes. The tensile and fracture properties of 15 vol% SiC_p/2009Al were studied. The results showed that hot extrusion increased the ultimate tensile strength (UTS), yield strength (YS), elongation (EL), reduction in area (RA), and fracture toughness of the composites. The heat treatment resulted in the increase in UTS, YS, and fracture toughness, but a decrease in EL and RA. Both hot extrusion and heat treatment had negligible effects on elastic modulus (*E*). With the increase of SiC_p size, the UTS, YS, and *E* decreased, but the EL and RA increased. The fracture toughness increased first and then decreased with increasing SiC_p size, and when the SiC_p size was about 7 μm, the composites obtained the maximum fracture toughness value of 31.74 MPa m^{1/2}.

KEY WORDS: Aluminum matrix composites; Silicon carbide; Tensile properties; Fracture toughness; Particle size

1 Introduction

In recent years, silicon carbide particle (SiC_p)-reinforced aluminum matrix composites (SiC_p/Al) have been widely used in the sporting goods, automotive, aircraft, and aerospace fields because of their excellent properties like light weight, high elastic modulus, and high strength [1–8]. For example, Al matrix composites reinforced with 15–25 vol% SiC_p have been chosen as the candidate materials for aircraft and aerospace applications, owing to their extra-high ultimate tensile strength (UTS) and yield strength (YS), as well as high elastic modulus (*E*). However, when the high strength and stiffness are

obtained, relative low ductility is always presented in the composites. In other words, the increase in strength and stiffness occurs at the expense of ductility. Therefore, to obtain an Al/SiC_p with high strengths combined with good ductility for aircraft and aerospace applications, the factors affecting the strength and ductility of composites should be systematically studied. However, completely understanding the influence factors on the strength and ductility presents quite a challenge, because the strength and ductility are sensitive to the fabricating method, the microstructure condition, heat treatment condition, and the component matching between SiC_p reinforcement and Al alloy matrix [9–12].

As for the fabricating method, powder metallurgy (PM) is usually used to fabricate the high property SiC_p/Al composites. However, pores generally exist in the matrix when fabricating process is not suitable, and thus deteriorate the mechanical properties of composites. Hot isostatic pressing (HIP) sintering process has been verified to be a very effective process to get composites with very low porosity (sometime the porosity value can reach to zero),

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J. Nie · J. Fan (✉) · S. Zhang · S. Wei · T. Zuo · Z. Ma · Z. Xiang
General Research Institute for Nonferrous Metals,
Beijing 100088, China
e-mail: jzfan@grinm.com

because HIP can provide very high and uniform sintering pressure for composite preparation.

The microstructure factors affecting the mechanical properties include the distribution of SiC_p within the matrix, the microstructure change caused by heat treatment, and the interfacial bonding at particle/matrix interface. In PM-fabricated materials, the SiC_p distribution will depend on the blending and consolidation procedures. If the fabricating parameters are improper, the SiC_p clusters would exist in the composites [13]. These cluster defects usually act as stress concentrators that increase the local stress intensity and promote easy crack nucleation, and thus reduces the ductility of composites [14, 15]. Heat treatment always causes the dislocation generation in the matrix as a result of mismatch in the coefficient of thermal expansion (CTE) between SiC_p and Al matrix on quenching from the solution treatment temperature. The dislocation presence changes the microstructure in the composites, and thus affects the mechanical properties of composites, especially the strength properties like UTS and YS [13, 15]. For interfacial bonding, if the interfacial bonding strength is weak, the interface debonding might happen and micro-cracks would initiate under a rather low external applied stress. Thus, the corresponding mechanical properties will be substantially decreased [16]. In order to improve the SiC_p distribution and interfacial bonding in composites, secondary fabrication processing, such as extrusion or rolling, should be used. For example, hot extrusion with a high extrusion ratio can homogenize the SiC_p distribution to some extent, and enhance the interfacial bonding strength at particle/matrix interface, and thus achieve optimum properties [13].

Another important factor affecting the mechanical properties of composites is the component matching between reinforcement and matrix. As reported in previous literatures [15, 17], different-sized SiC_p -reinforced composites can lead to different mechanical properties for composites, and only the proper SiC_p and Al matrix matching can achieve the highest strength and the optimum ductility for composites.

In the present study, HIP and hot extrusion processes were employed to fabricate 15 vol% SiC_p /2009Al composites with different size SiC_p to investigate the effect of component matching between reinforcement and matrix on the mechanical properties.

2 Experimental

The matrix material used in the present study was gas-atomized 2009Al alloy powder with an average diameter of 45 μm and a nominal composition of Al–4.2Cu–1.1 Mg (wt%). 15 vol% SiC_p with an average particle size of 5 μm was added

in the composites as a reinforcement. The 2009Al powders and SiC_p were mixed using ball-mill for 24 h with ball-to-powder weight ratio of 2:1. Sequentially, the mixed powders were cold pressed and degassed in vacuum, followed by hot isostatic pressing (HIP). The HIP sintering temperature is about 580 °C, and the holding time is 3 h. The HIPed specimens were then hot extruded to rods at 470 °C with an extrusion ratio of 14:1. The as-received samples were solution treated at 515 °C for 2 h, then water quenched to room temperature, and finally naturally aged at room temperature for more than 96 h. Un-reinforced 2009Al alloy has also been fabricated under identical preparing processes which was used for comparison. Moreover, in order to investigate the influence of component matching between SiC_p and matrix on the mechanical properties of composites, 15 vol% SiC_p /2009Al composites containing different-sized SiC_p of 3.5 μm (2009Al/ $\text{SiC}_{p(3.5 \mu\text{m})}$), 7 μm (2009Al/ $\text{SiC}_{p(7 \mu\text{m})}$), 11 μm (2009Al/ $\text{SiC}_{p(11 \mu\text{m})}$), and 13 μm (2009Al/ $\text{SiC}_{p(13 \mu\text{m})}$) were also fabricated.

The bulk densities of the fabricated composite samples were measured by standard Archimedes method. Tensile tests of the composite samples were carried out at room temperature on an AMSLER-100-20 testing machine at a tensile rate of 1 mm/min. Dynamic knock method was used to measure the E for composites using a modulus measuring apparatus (RFDHTVP 1750-C, IMCE, Co., Ltd, Belgium) at ambient temperature, and the specimen dimension is 80.0 mm $\times$ 20.0 mm $\times$ 3.0 mm. The fracture toughness tests were performed under three-point bend loading at a nominal crosshead speed of 0.5 mm/min. Optical microscopy, scanning electron microscopy (SEM), and transmission electron microscopy (TEM) were used to characterize the microstructures of 2009Al alloy matrix and 2009Al/ SiC composites. X-ray diffraction (XRD) characterization of composites was performed using a Phillips D-800 generator with CuK_α radiation.

3 Results and Discussion

3.1 Effects of Microstructure and Fabricating Process on Mechanical Properties

3.1.1 Microstructures of Composites

The SiC_p distributions in the HIPed and as-extruded 2009Al/ SiC_p composites with a reinforcement size of 5 μm (2009Al/ $\text{SiC}_{p(5 \mu\text{m})}$) are shown in Fig. 1. Figure 1a shows the microstructure of HIPed composites. It can be found that the irregularly shaped SiC_p is dispersed randomly through the matrix. Additionally, clustering or agglomerations of the reinforcing SiC_p are observed. It is known that if the particle size is much smaller than the matrix powder

Fig. 1 The optical micrographs showing the SiC_p distribution in HIPed composites **a**, as-extruded composites (transverse direction) **b**

Fig. 2 TEM images of particle/matrix interfaces in HIPed composites **a, b**, as-extruded composites **c, d**

size, the reinforcements can cluster very easily in the matrix and result in an inhomogeneous distribution of reinforcements [13, 16]. In the current experiment, the

2009Al powder has an average size of 45 μm, while the mean particle size of SiC_p is about 5 μm. Therefore, this large size deviation between SiC_p and 2009Al powder

Fig. 3 Microstructure morphologies and energy spectra of 2009Al/SiC_p(5 μm): **a** as-extruded composites showing lots of white phases; **b** energy spectrum of the white phase in the as-extruded composites; **c** extruded and T4-treated composites showing only a few coarse white phases; **d** energy spectrum of the white phases in extruded and T4-treated composites

should be the main reason for the presence of SiC_p clusters in the HIPed composites. Hot extrusion provides a homogeneous distribution of SiC_p within 2009Al matrix, and the presence of particle clusters is not observed, as shown in Fig. 1b. This may be due to that the extrusion process generates severe plastic deformation of 2009Al alloy matrix, leading to the rearrangement of the SiC_p. Moreover, the applied shear during extrusion deformation can effectively break up the SiC_p clusters, resulting in a more uniform particle distribution.

Figure 2 shows the TEM images of the interfacial microstructures between SiC_p and 2009Al matrix in the HIPed and as-extruded composites. Figure 2a gives the full morphology of reinforcement–matrix interface region in the HIPed composite. As shown in Fig. 2a, a good interfacial bonding between SiC_p and 2009Al matrix was obtained, and no pores and voids were found. The good interfacial bonding was also obtained in the as-extruded 2009Al/SiC_p composites (Fig. 2c). The high-magnification images (Fig. 2b, d) further demonstrate the good interfacial bonding in the composites. Good interfacial bonding can

effectively transfer the load from the matrix to the SiC_p reinforcements, which is expected to produce a high strength in the composites.

The microstructures and energy spectra of the as-extruded 2009Al/SiC_p(5 μm) before and after T4 treatment are shown in Fig. 3. It is found that a large number of large-sized white phases were present in the Al alloy matrix for the as-extruded 2009Al/SiC_p(5 μm) before T4 treatment (Fig. 3a). The corresponding energy spectrum (Fig. 3b) reveals the white phase consisted of Al, Cu, and Mg. After the T4 treatment which was operated at the solution temperature of 515 °C for 2 h, most of the white phase particles disappeared in the extruded and T4-treated 2009Al/SiC_p(5 μm), as shown in Fig. 3c. However, a few coarse white particles still existed. These phases were determined by energy spectrum (Fig. 3d) to be Fe-rich intermetallics which cannot dissolve during solution heat treatment.

Figure 4 shows XRD patterns of the as-extruded and extruded and T4-treated 2009Al/SiC_p(5 μm). For comparison, XRD patterns of the 2009Al matrix are also given in

Fig. 4 XRD patterns of 2009Al matrix and 2009Al/SiC_p(5 μm)

Fig. 4. As shown in Fig. 4, the Al₂CuMg is detected in the as-extruded 2009Al matrix, while in the as-extruded 2009Al/SiC_p(5 μm) composites, the Al₂Cu and Mg₂Si are identified. The presence of Mg₂Si in the composites is due to that the SiC_p used for the composite manufacturing contains some Si. The Si element is combined with Mg, precipitating in the form of Mg₂Si in the composites [18]. The reason for the formation of Al₂Cu rather than Al₂CuMg in the 2009Al/SiC_p(5 μm) composites can be explained as follows. The phase formation in the Al–Cu–Mg alloys depends on the Cu content in and the Cu/Mg ratio. Small changes in the Cu/Mg ratio have a great influence on the formation and amount of equilibrium phases [18]. The formation of Mg₂Si in the composites decreases the Mg concentration, which results in the increase of Cu/Mg ratio, and thus inhibits the formation of Al₂CuMg in the 2009Al/SiC_p(5 μm) composites, as shown in Fig. 4c [19]. For the extruded and T4-treated 2009Al matrix, the XRD pattern in Fig. 4 confirms the fully dissolution of Al₂CuMg precipitates in the 2009Al matrix which was solution treated at 515 °C for 2 h. In the case of extruded and T4-treated 2009Al/SiC_p(5 μm) composites, the small peaks of Mg₂Si still appears in the XRD pattern, though the Al₂Cu is fully dissolved, suggesting that the Mg₂Si did not (or incompletely) dissolve during solution treatment.

In a word, the homogeneously distributed SiC_p composites and good interfacial bonding at particle/matrix interface are successfully fabricated using the present HIP and hot extrusion processes. Additionally, through the proper heat treatment process used in the present study, the composite is expected to have good mechanical properties.

Table 1 Density and relative density of 2009Al/SiC_p(5 μm) composites

Materials status	Measured density (g/cm ³)	Theoretical density (g/cm ³)	Relative density (%)	Porosity (%)
HIPed	2.83	2.83	100	0
As-extruded	2.83	2.83	100	0
Extruded and T4-treated	2.83	2.83	100	0

Table 2 Mechanical properties of 2009Al matrix and 2009Al/SiC_p(5 μm) composites

Materials	UTS (MPa)	YS (MPa)	E (GPa)	EL (%)	RA (%)
HIPed 2009Al/SiC _p (5 μm)	312.5	190.0	99.5	4.5	5.5
As-extruded 2009Al/SiC _p (5 μm)	341.0	192.0	100.0	8.5	15.5
Extruded and T4-treated 2009Al/SiC _p (5 μm)	566.0	379.5	100.5	8.0	12.0
HIPed 2009Al matrix	271.0	118.0	69.8	18.0	25.0
As-extruded 2009Al matrix	327.0	185.0	70.1	26.0	46.0
Extruded and T4-treated 2009Al matrix	439.0	307.0	70.0	24.0	32.0

3.1.2 Relative Densities and Mechanical Properties of Composites

The relative density values of 2009Al/SiC_p(5 μm) composites are listed in Table 1. As it is expected, all of the composites get the relative densities of up to 100% (Table 1). These results indicate that the HIP process can effectively prepare the full dense materials, and nearly no pores and cracks were introduced into the composites during the following hot extrusion procedures or T4 treatment.

The tensile properties of 2009Al/SiC_p(5 μm) composites, including UTS, YS, *E*, elongation (EL), and reduction in area (RA) are tested, and the results are given in Table 2. As shown in Table 2, the HIPed 2009Al/SiC_p(5 μm) composites get the UTS, YS, *E*, EL, and RA of 312.5 MPa, 190 MPa, 99.5 GPa, 4.5%, and 5.5%, respectively. The obtaining of the above good tensile properties may be due to the high relative density (100%) and good interfacial bonding (Fig. 2a) in the HIPed composites. The UTS, YS, EL, and RA of the as-extruded composites are higher than those of HIPed ones, although the modulus almost remains the same. The enhancements in UTS and YS of the as-extruded materials are attributed to the extrusion deformation, which effectively improves the distribution of

Fig. 5 TEM image of dislocations within Al matrix near the interface region of extruded and T4-treated composites

reinforcement in matrix (Fig. 1), since the plastic flow associated with extrusion tends to disperse clusters of reinforcing particles. Moreover, extrusion with the high extrusion ratio can disrupt the oxide film on the metal powder particles, increasing the mechanical properties of the composites [13]. The increases of EL and RA of the as-extruded composites which are compared with HIPed composites (Table 2) are due to the increased plastic deformation capacity of the matrix after extrusion deformation [20].

Extruded and T4-treated composites give an increase in UTS and YS, but a decrease in EL and RA, compared with the as-extruded composites (Table 2). Moreover, there is nearly no difference in the E between as-extruded and extruded and T4-treated composites. The increases of UTS and YS of the extruded and T4-treated composites can be due to the high-density dislocations generated by T4 treatment in. As we known, there is a large difference in CTE between SiC_p and aluminum alloy (CTE of Al is about $23.5 \times 10^{-6}/\text{K}$ and CTE of SiC_p is about $4.2 \times 10^{-6}/\text{K}$). The mismatches in CTE between the matrix and reinforcement result in the generation of thermal tensile residual stresses in the matrix, when the composite is cooled from solution temperature to room temperature. These residual stresses can be partially relaxed by the generation of a large number of dislocations in the matrix, which can cover most of the matrix domain. Furthermore, the smaller particle size (smaller inter-particle spacing) or higher particle volume fraction can result in a higher dislocation density [15, 21, 22].

Figure 5 shows the TEM image of extruded and T4-treated composites, revealing high-density dislocations

Fig. 6 K_{IC} values of 2009Al matrix and 2009Al/ $\text{SiC}_p(5\mu\text{m})$ at three different conditions

located in the Al matrix (near the particle/matrix interface) region. The observed dislocations in extruded and T4-treated composites (Fig. 5) are mostly generated through solution treatment, but this still needs further experimental verification because extrusion deformation can also cause the generation of large number of dislocations in matrix. The degradations in EL and RA in extruded and T4-treated composites (Table 2) can also be attributed to the increase of dislocation density caused by T4 treatment.

For comparison, the tensile properties of the 2009Al matrix are also summarized in Table 2. As shown in Table 2, under the same preparing condition, the composites get the higher UTS, YS, and E , while the lower EL and RA than those of 2009Al matrix. For example, the UTS, YS, and E of the extruded and T4-treated composites are 566 MPa, 379.5 MPa, and 100.5 GPa, respectively, which are 28.9, 23.6, and 43.6% higher than those of extruded and T4-treated 2009Al matrix. Nevertheless, the EL and RA of extruded and T4-treated composites are 8.0 and 12.0%, which are far lower than those of extruded and T4-treated matrix (EL = 24.0%, RA = 32.0%).

Figure 6 shows the fracture toughness, K_{IC} , of the composites at three different conditions. For the purpose of comparison, the K_{IC} values of 2009Al matrix are also given in Fig. 6. In case of composites, the HIPed composite obtains the minimum K_{IC} , and the as-extruded composite gets higher fracture toughness than the HIPed composite. Moreover, the extruded and T4-treated composite has the maximum fracture toughness value. Similar fracture toughness variation trends are observed in the 2009Al matrix. Additionally, the composite has lower K_{IC} value than that of 2009Al matrix at the same fabricating condition, i.e., the fracture toughness of the extruded and T4-

Table 3 Comparison of tensile properties among this work and other reported values

Materials	UTS (MPa)	YS (MPa)	E (GPa)	EL (%)	RA (%)
2009Al/SiC _p	566.0	379.5	100.5	8.0	12.0
2009Al/SiC _p [20]	520.0	365.0	–	8.4	–
2009Al/SiC _p [23]	537.0	367.0	100.0	9.3	9.2

treated composites is $30.42 \text{ MPa m}^{1/2}$, which is 24.7% lower than that of 2009Al alloy matrix ($40.40 \text{ MPa m}^{1/2}$). This indicates that the addition of SiC_p reinforcement into matrix significantly degrades the toughness of matrix.

In a word, the PM HIP process, hot extrusion, as well as heat treatment can significantly affect the tensile properties and fracture toughness of composites through changing the microstructures in the composites. This means that a good microstructure without defects in material is very important for obtaining high mechanical properties.

Previously, the preparation and mechanical properties of 15 vol% SiC_p-reinforced 2009Al composites have also been studied by other processes, i.e., hot pressing and extrusion [20, 23]. The comparison of tensile properties among the current work and other reported values is shown in Table 3. As shown in Table 3, the strength, modulus, and RA of this work are higher than those reported in Ref. [20, 23]. The lower EL (Table 3) suggests that the increase in strength and stiffness always occurs at the expense of ductility.

3.2 Effects of Matching between SiC_p and Matrix on Mechanical Properties

As mentioned above, the proper preparing processes can reduce and eliminate the microstructure defects (pores, clusters of particle) in the composites, and thus improve the mechanical properties of composites such as UTS and YS (Table 2). However, when the composite is used for aviation and aerospace materials, not only the high UTS and YS, but also the high stiffness and good plastic properties are required. In order to get the composites with high mechanical properties, the proper choice of SiC_p size matching with matrix is very important [13, 15]. Therefore, five composites reinforced with 3.5, 5, 7, 11, and 13 μm SiC_p are fabricated to investigate the effects of component matching on mechanical properties. The five engineering stress versus strain curves of the 2009Al/SiC_p composites fabricated by HIP and hot extrusion processes are shown in Fig. 7. It is apparent that the five 2009Al/SiC_p composites obtain different UTS and YS, as shown in Fig. 7.

The UTS, YS, and E variations with SiC_p sizes are shown in Fig. 8. For the constant particle fraction

Fig. 7 The engineering stress versus engineering strain curves of the five 2009/SiC_p composites reinforced with different SiC_p sizes **Fig. 8** The UTS, YS, and E variations with SiC_p sizes in extruded and T4-treated composites

(15 vol%), a decrease in composite strength with increasing SiC_p size can be seen in Fig. 8. According to the reported studies [13, 15], these results can be rationally explained by the following reasons: (1) In larger particle-reinforced composites, the inter-particle distance is greater and the total particle/matrix interface area is fewer. Thus, the total amount of dislocations generated by the thermal expansion mismatch mechanism is fewer, and consequently, the strengthening mechanisms are less effective. (2) The larger the particle, the more it will be loaded by fiber and end-loading mechanisms, and hence the particle fracture is more likely to occur in large particle composites. (3) The larger particles will have a greater probability of containing a critical size defect, and therefore, the intrinsic strength of the particle is lower and the particle fracture is

Fig. 9 SEM morphologies of the tensile fracture surfaces: **a** 2009Al/SiC_p(3.5 μm); **b** 2009Al/SiC_p(7 μm); **c** 2009Al/SiC_p(13 μm); **d** high-magnification observation of Fig. 9c showing particle cracking

more likely to occur. If a particle breaks, it loses the ability to bear a tensile load, and the stress in the matrix increases, reducing the strength of the composite.

Figure 9 shows the tensile fracture surfaces of the 2009Al/SiC_p composites reinforced with 3.5, 7, and 13 μm SiC_p. The fracture surfaces of 3.5 and 7 μm SiC_p-reinforced composites show no SiC_p cracking because of the high intrinsic strengths of 3.5 and 7 μm SiC_p. However, when SiC_p size increases up to 13 μm, the break of SiC_p can be apparently observed on the fracture surfaces of 2009Al/SiC_p(13 μm) due to the weaker intrinsic strength of 13 μm SiC_p, compared with the finer ones (3.5 or 7 μm). Therefore, these different fracture features observed in Fig. 9 could, at least partially, confirm the above-mentioned strengthening reasons. Furthermore, no interface debonding is observed in the present work (Fig. 9d), which further suggests that the composite prepared by HIP and hot extrusion techniques has a strong particle/matrix interfacial strength.

As for E , the theories used to explain the variation of the E of metal matrix composites with volume fraction of reinforcement do not normally take into account the effect of reinforcement particle size. However, it can be seen from Fig. 8 that there is a clear decrease in the E of Al matrix composite with increasing SiC_p size for the given fraction of SiC_p (15 vol%). Other researchers [17, 24] also studied the effect of reinforcement size on E , and they got the same trend of E change. Such experimental observations may be explained as follows: (1) In small particle-reinforced composites, the inter-particle distance is smaller and the total particle/matrix interface area is larger. The interface area increase provides a more efficient load transfer in the composites, and thus making an increase in E in the small particle-reinforced composites, compared with larger particle-reinforced composites [17, 24]. (2) Another reason to explain this phenomenon is to consider the trend of density of composite reinforced with different size SiC_p [17, 24]. Table 4 shows that density decreases

Table 4 Densities of composites reinforced with different-sized SiC_p (in g/cm³)

2009Al/ SiC _p (3.5 μm)	2009Al/ SiC _p (5 μm)	2009Al/ SiC _p (7 μm)	2009Al/ SiC _p (11 μm)	2009Al/ SiC _p (13 μm)
2.83	2.83	2.828	2.827	2.826

Fig. 10 The EL and RA variations with SiC_p sizes in extruded and T4-treated composites

with increasing SiC_p size. The low density may cause the low E value. However, in order to fully understand the relationship between SiC_p size and E , other influencing factors such as particle shape, orientation, and distribution should be taken into account, and this is our future work.

Figure 10 shows the EL and RA of the five different composites. Both the EL and RA increase with increasing SiC_p sizes, indicating that the large particle composite has higher plasticity than the small particle composite. This is attributed to the fact that the particle size increase is associated with interfacial matrix/particle area decrease, leading to dislocation density decrease. Hence, the plastic straining around the particles occurs more easily in the large particle composites, resulting in a higher plastic deformation capacity in the large particle composites, compared with the small particle composites.

The fracture toughness of the composites is shown in Fig. 11. It is observed from Fig. 11 that the fracture toughness of composites was found in the range of 28.74–31.74 MPa m^{1/2} for SiC_p sizes in the range of 3.5–13 μm. Moreover, the fracture toughness increases first and then decreases with increasing SiC_p size, and when the SiC_p size is 7 μm, the corresponding composite exhibits the maximum fracture toughness value of 31.74 MPa m^{1/2}. In order to explain the above observed variation behaviors of fracture toughness, the matrix and particle intrinsic strengths, interfacial bonding strength, and SiC_p distribution condition in different composites should be taken into

Fig. 11 The relationship between fracture toughness and SiC_p size

account. However, the effects of particle size on fracture toughness are still less understood according to previous work [15, 25, 26]. Further, theoretical analysis should be carried out to explain the relationship between fracture toughness and SiC_p size.

Through the above investigation, we can see that the component matching between SiC_p and matrix plays a very important role in the composite properties. If the high strength and stiffness are asked for composite application, the smaller particle (3.5 or 5 μm) may be a good choice to match with Al matrix to fabricate composites (Fig. 8). However, when high EL and RA are required, the larger particle (11 or 13 μm)-reinforced composites can meet the property demand (Fig. 10). If the composites should have high fracture toughness and good strength and stiffness, the medium size (7 μm) SiC_p matching with 2009Al matrix seems to be the best choice to fabricate composites (Fig. 11).

4 Conclusions

- (1) 15 vol% SiC_p/2009Al composites with full dense microstructure, homogeneous SiC_p distribution, and good interfacial bonding strength at particle/matrix interface were successfully fabricated by HIP and hot extrusion processes.
- (2) Preparing processes affected the microstructures and mechanical properties of composites. Hot extrusion improved the distribution of SiC_p reinforcement, and effectively enhanced the UTS, YS, RA, EL, and fracture toughness of the composites. Heat treatment improved the UTS, YS, and fracture toughness, but did not contribute to the enhancements of RA and EL.

Both hot extrusion and heat treatment almost had no effect on E .

- (3) Composite component matching between reinforcement and matrix had significant influence on the mechanical properties. The smaller size SiC_p gave higher strength and E , but lower RA and EL for the 2009Al/ SiC_p composites. However, the larger size SiC_p -reinforced composites exhibited higher EL and RA, but lower strength and EL. The fracture toughness increased first and then decreased with increasing SiC_p size, and when the SiC_p was about 7 μm , the composites obtained the maximum fracture toughness value of 31.74 $\text{MPa m}^{1/2}$.

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References

- [1] Q. Wang, L.C. Bian, H.J. Li, Compos. Struct. **94**, 2218 (2012)
- [2] M.N. Yuan, Y.Q. Yang, C. Li, P.Y. Heng, L.Z. Li, Mater. Des. **38**, 1 (2012)
- [3] D. Veazie, J.L. Jordan, J.E. Spowart, B.W. White, N.N. Thadhani, Exp. Mech. **53**, 1213 (2013)
- [4] Y.S. Su, Q.B. Ouyang, W.L. Zhang, Z.Q. Li, Q. Guo, G.L. Fan, D. Zhang, Mater. Sci. Eng. A **597**, 359 (2014)
- [5] Q. Liu, F. Ye, Y. Gao, S.C. Liu, H.X. Yang, Z.Q. Zhou, J. Alloys Compd. **585**, 146 (2014)
- [6] C. Xue, J.K. Yu, Mater. Des. **53**, 74 (2014)
- [7] O. El-Kady, A. Fathy, Mater. Des. **54**, 384 (2014)
- [8] S. Bathula, R.C. Anandani, A. Dhar, A.K. Srivastava, Mater. Sci. Eng. A **30**, 97 (2012)
- [9] E.A. Diler, R. Ipek, Mater. Sci. Eng. A **548**, 43 (2012)
- [10] D. Mandal, S. Viswanathan, Mater. Charact. **85**, 73 (2013)
- [11] D. Mandal, S. Viswanathan, Mater. Charact. **86**, 21 (2013)
- [12] R. Bhushan, S. Kumar, S. Das, Int. J. Adv. Manuf. Technol. **65**, 611 (2013)
- [13] D.J. Lloyd, Int. Mater. Rev. **39**, 1 (1994)
- [14] N. Chawla, U. Habel, Y.L. Shen, Metall. Mater. Trans. A **31**, 531 (2000)
- [15] M.T. Milan, P. Bowen, J. Mater. Eng. Perform. **13**, 775 (2004)
- [16] M. Song, Y.H. He, Mater. Des. **31**, 9859 (2010)
- [17] H.K. Jung, Y.M. Cheong, H.J. Ryu, S.H. Hong, Scr. Mater. **41**, 1261 (1999)
- [18] P. Rodrigo, P. Poza, V. Utrilla, A. Ur̃ena, J. Alloys Compd. **479**, 451 (2009)
- [19] J. Zhou, J. Duszczyk, J. Mater. Sci. **34**, 545 (1999)
- [20] P. Jin, B.L. Xiao, Q.Z. Wang, Z.Y. Ma, Y. Liu, S. Li, J. Mater. Sci. Technol. **27**, 518 (2011)
- [21] Z. Zhao, Z. Song, Y. Xu, Mater. Sci. Eng. A **132**, 83 (1991)
- [22] N. Han, G. Pollard, R. Stevens, Mater. Sci. Technol. **8**, 184 (1992)
- [23] T.S. Srivatsan, M. Al-Hajri, C. Smith, M. Petraroli, Mater. Sci. Eng. A **346**, 91 (2003)
- [24] J.C. Lee, K.N. Subramanian, J. Mater. Sci. **29**, 49015 (1994)
- [25] Y. Flom, R.J. Arsenault, Acta Mater. **37**, 2413 (1989)
- [26] M. Manoharan, J.J. Lewandowski, Acta Mater. **38**, 489 (1990)