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Predicting Interfacial Strengthening Behaviour of Particulate-Reinforced MMC — A Micro-mechanistic Approach

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Abstract

The fracture properties of particulate-reinforced metal matrix composites (MMCs) are influenced by several factors, such as particle size, inter-particle spacing and volume fraction of the reinforcement. In addition, complex microstructural mechanisms, such as precipitation hardening induced by heat treatment processing, affect the fracture toughness of MMCs. Precipitates that are formed at the particle/matrix interface region, lead to improvement of the interfacial strength, and hence enhancement of the macroscopic strength properties of the composite material. In this paper, a micro-mechanics model, based on thermodynamics principles, is proposed to determine the fracture strength of the interface at a segregated state in MMCs. This model uses energy considerations to express the fracture toughness of the interface in terms of interfacial critical strain energy release rate and elastic modulus. The interfacial fracture toughness is further expressed as a function of the macroscopic fracture toughness and mechanical properties of the composite, using a toughening mechanism model based on crack deflection and interface cracking. Mechanical testing is also performed to obtain macroscopic data, such as the fracture strength, elastic modulus and fracture toughness of the composite, which are used as input to the model. Based on the experimental data and the analysis, the interfacial strength is determined for SiC particle-reinforced aluminium matrix composites subjected to different heat treatment processing conditions.

Keywords

Interface strength, metal matrix composites, aluminium alloys, mechanical behaviour, fracture toughness

1. Introduction

When interfacial fracture occurs, a polycrystal exhibits brittle fracture behaviour [1, 2], which is considered to be a major weakness of many advanced, high performance structural materials, such as metal matrix composites used in high-

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temperature applications. In contrast, crack deflection at the interface has been associated with improved mechanical properties of the material at the interface. Crack deflection is associated either with crack attraction or repulsion by second phase particles due to residual strains. An important factor regulating crack growth behaviour in metal matrix composites is the matrix–reinforcement interface property, which relates to precipitation hardening mechanisms [3].

Furthermore, it is known that molten aluminium does not wet silicon carbide readily, which is one of the major concerns that needs to be overcome to prevent silicon carbide particles being displaced from molten aluminium and to ensure SiC/Al bonding (Fig. 1). In addition, as mentioned, heating above a critical temperature can lead to the undesirable formation of Al_4C_3 flakes. MC-21, Inc. patented melt stirring, a method of satisfying these requirements and producing high quality composites. SiC particulates are added to Al–Si casting alloys where Si in the alloy slows down the formation of Al_4C_3. The process yields material with a uniform distribution of particles in a 95–98% dense aluminium matrix. The rapid solidification inherent in the process ensures minimal reaction between reinforcing material and the matrix [3–5].

In composite materials with ductile matrix and hard-brittle reinforcement, interfaces can be assumed to behave in a similar manner as in the case of grain boundaries. In such cases, the crack would propagate through the matrix and the crack-tip would meet the interfacial region, where plasticity and/or energy changes. Then, the crack may, (a) continue to propagate through the reinforcement or (b) be deflected by the matrix–reinforcement interface. It is, therefore, essential to be able

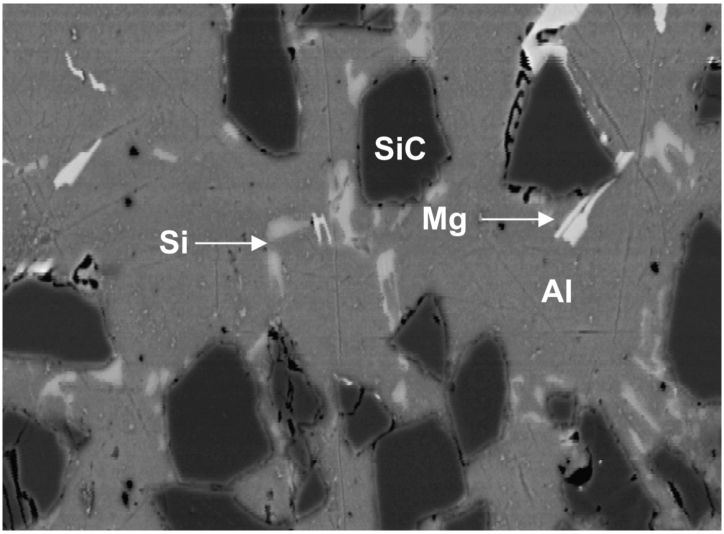

Figure 1. Microstructure of rolled 31% SiC in the as received condition showing four distinct phases: aluminium matrix, SiC particles, eutectic region of aluminium and silicon and Mg phase.

to predict whether the interfacial region has enough fracture strength in order to resist propagation of the crack through the interface.

Many factors can play a significant role in affecting the fracture properties of MMCs, such as particle size, inter-particle spacing, and volume fraction of the reinforcement [6–8]. Furthermore, more complex microstructural mechanisms such as precipitation hardening achieved by heat treatment processing, influence the fracture toughness values of MMCs. Using appropriate heat treatment conditions, precipitates are formed in the matrix material in the form of separate phases, leading to an improvement of interfacial strength of the composite, thereby enhancing the overall strength of the material [9, 10].

The thermodynamics of vacancy and impurity absorption at interfaces and grain boundaries in solids has been studied in the recent years with theoretical models proposed [11] in order to predict the behaviour of vacancies at interfaces in a stress gradient, as well as the interface strength at fracture. It has been reported in the literature that the tendency for intergranular fracture is closely related to the type and structure of grain boundaries. Low-energy boundaries are resistant to fracture while high-energy or the so-called random boundaries are favoured locations for crack nucleation and propagation. Faulkner and Shvindlerman [11] and Lim and Watanabe [12] have recognized the important role interface structure plays in determining the amount of predicted segregation and hence the change of interfacial energy caused by segregation.

Certain amounts of plastic deformation are involved with crack propagation along an interface. The parameters to be considered are the stain rate sensitivity to stress and the dislocation pile-up behaviour at the advancing crack-tip. Using this approach, the effective work parameter can be shown to be thousand times larger than the surface energy [11]. This implies that minute changes in surface energy caused by segregation would result in large changes in interfacial fracture stress.

In ductile materials such as metals, plastic deformation occurs at the crack-tip. Much work is required in producing a new plastic zone at the tip of the advancing crack. Since the plastic zone has to be produced upon crack growth, the energy for its formation can be considered as energy required for crack propagation. This means that for metals R (crack propagation resistance), $\mathrm{d}W/\mathrm{d}A$ is mainly plastic energy; the surface energy is so small that it can be neglected [13].

2. Model

A model proposed by McMahon and Vitek [14] predicts the fracture resistance of a ductile material that fails by an intergranular mechanism. Based on this model, an effective work parameter can be developed to predict fracture strength of an interface at a segregated state using Griffith crack-type arguments. The Griffith equation, which was derived for an elastic body, is applied here because it is assumed that the yielding zone size ahead of the crack is small enough and the fracture is governed by the elastic stress field. The model further assumes that small changes in interfacial energy caused by segregation of impurities at the interface will result in a much larger change in the work of fracture. This is due to the fact that the work of fracture must be provided by a dislocation pile-up mechanism around the advancing crack-tip on the interface. This implies that additional work must be provided to deform the material at the crack-tip in addition to the work needed to overcome the interface energy and to replace it with two surfaces. The definition of interfacial fracture strength, σ_{int}, is then given by:

$$\sigma_{\text{int}} = \sqrt{\frac{100\varepsilon_{\text{p}} E_{\text{int}}}{\pi d}}, \tag{1}$$

where E is Young's modulus and d is the particle thickness, since it is assumed that cracks of the order of the particle size are present when considering crack propagation through the interface and the particulate; ε_{p} is the energy required to create two fracture surfaces, with

$$\varepsilon_{\text{p}} = 2\varepsilon_{\text{s}} - \varepsilon_{\text{gb}} (= \varepsilon_0),$$

where ε_{s} is the surface energy, and ε_{gb} is the grain boundary energy.

The $100\varepsilon_{\text{p}}$ component allows for dislocation interaction and movement ahead of the crack-tip in ductile materials. This refers to the work required for a total separation of the lattice planes, which is equal to the area under the force–extension curve.

From equation (2) ε_{p} can be estimated if K_{int} (interface fracture toughness) and E_{int} (interface Young's modulus) are known [15].

$$\frac{K_{\text{int}}^2}{100E_{\text{int}}} = \varepsilon_{\text{p}} \left(1 - \frac{ZRT \ln(1 - c + Bc)}{\varepsilon_{\text{p}}}\right)^n, \tag{2}$$

where Z describes the density of interface sites which are disordered enough to act as segregation sites ($= D\rho_{\text{s}}$), with D the thickness of the interface region and ρ_{s} the density of the interface region ($D = 300$ nm) ($\rho = 2.6889$ g/cm^3 for aluminium and 3.22 g/cm^3 for SiC). R is the gas constant ($= 8.314472(15)$ J $\cdot$ K^{-1} $\cdot$ mol^{-1}); T is the absolute temperature ($T = 803.15$ K for T6, $T = 723.15$ K for HT1); c is the segregate concentration needed to cause embrittlement ($= 0.1$ for pure aluminium); B describes the modification of the boundary energy by impurities using the Zuchovitsky equations; and n is the work hardening exponent ($n = 10$ for FCC aluminium).

3. Mechanical Properties

The mechanical properties of the 31% SiC aluminium matrix composite have been obtained from previous work [16]. The fracture toughness $K_{1\text{c}}$ value has been measured for three different heat treatment conditions. Also, the Young's modulus has been calculated, shown in Table 1.

Table 1.
Mechanical properties of Al/SiCp composite

Material	Heat treatment	E (GPa)	$Rp_{0.2}$ (MPa)	K_{IC} (MPa $\sqrt{m}$)
A359 Al	–	71	75	35
A359/SiC/31p	AR	108	158	19.28
A359/SiC/31p	T6	116	290	22.05
A359/SiC/31p	HT1	110	155	20.75

3.1. Interface Fracture Toughness K_{int}

In hard particle-reinforced metal matrix composites, the stress transfer from the matrix to the particles is mainly controlled by the misfit of the elastic constants between the two phases [17]. To measure the stress transfer to the particle, in a homogeneous material subjected to tensile loading, the stress carrying capability, L, of the particle is defined as the ratio of the normal stress σ_N to the particle in the loading direction to the macroscopic tensile stress, σ_T, i.e., the ratio $L = \sigma_N/\sigma_T$. By using Eshelby's theory, the stress carrying capability of a spherical inhomogeneity can be written as [18]:

$$L = \frac{9x(2+3x)}{(1+2x)(8+7x)},$$

where $x = E_i/E_m$, and E_i and E_m are Young's moduli for inhomogeneity and matrix, respectively.

Furthermore, the shear lag model, originally developed by Cox [19] modified by Llorca [20], can be used to estimate the stress carrying capability of a particulate, assuming that the volume fraction of reinforcement is small and the average stress in the matrix is approximately equal to the applied stress:

$$L = 1 + \frac{a}{\sqrt{3}},$$

where $a = \bar{h}/(2\bar{r})$ is the aspect ratio of the reinforcement, with $\bar{h}$ and $\bar{r}$ the average length and the average diameter of the particle.

A model has been proposed to estimate the effects of particle volume fraction on fracture toughness in SiC particle-reinforced aluminium alloy matrix composites. This model assumes that SiC particles are uniformly distributed in the matrix and that the pattern of particle distribution is similar to FCC structure in metals. The fracture toughness of the composite can then be written as [15]:

$$K_{IC} = \frac{K_p}{L_p} V'_m + \frac{2K_{int}}{L_p + L_m}(V_m - V'_m) + \frac{K_m}{L_m} 2V_m + K_m(1 - 3V_m), \qquad (3)$$

where K_{IC}, $K_p = 3$ MPa m$^{-1/2}$, $K_m = 35$ MPa m$^{-1/2}$, and K_{int} is the fracture toughness of the composite, SiC particulates, A359 aluminium alloy matrix and interface, respectively. L_p and L_m are the stress carrying capabilities of a particulate and the matrix, respectively. On average, for SiC particles and aluminium alloy matrix, $L_p \sim L_m \sim 2$. The value of $L_m = 1$ is applicable for clean surfaces. However, due to processing conditions and the physical interaction at the matrix/reinforcement interface, the interface contains partially contaminated surfaces; therefore, $L_m = 2$ since it cannot be considered as a 'clean surface'. V_m and $(V_m - V'_m)$ are the area fractions for particle cracking and interface failure, respectively. These area fractions though are not accurately known. However, Wang and Zhang [21] found that the ratio of particle cracking over interface failure $V_m/(V_m - V'_m)$ was about $0.13(= 1.4\%/10.7\%)$ in a SiC particle-reinforced aluminium alloy composite.

3.2. Young's Modulus of the Interface Region

Young's modulus of the matrix has been obtained for A359 aluminium matrix. The particles' E_p, matrix E_m and interface E_i are shown in the equation

$$E_C = E_p v_f^{2/3} + E_m(1 - V_f'^{2/3}) + E_i(V_f'^{2/3} - V_f^{2/3}). \tag{4}$$

Due to the fact that the difference $(V'_f - V_f)$ is very small, a good approximation is to consider that the Young's modulus of the interface region is close to that of the matrix; $E_i \cong E_m$ [15].

3.3. Constants Calculations

The parameter B describes the modification of the boundary energy by impurities using the Zuchovitsky equations [22, 23], given by:

$$B = e^{(\frac{\varepsilon_1 - \varepsilon_2}{RT})} \cong e^{(\frac{0.75\varepsilon_f}{RT})}, \tag{5}$$

where $\varepsilon_2 - \varepsilon_1$ is the difference between the formation energy in the impurity in the bulk and the interface region. It is assumed that the values of the surface energy and the impurity formation energy in the bulk are close in value; therefore, the numerator in the exponential term depends on the impurity formation energy in the interface region, which is assumed to be $0.75\varepsilon_f$, where ε_f is the formation energy of the impurity in the bulk.

Using Faulkner's approach [24], to the derivation of impurity formation energy,

$$\varepsilon_f = \varepsilon_s + \varepsilon_e, \tag{6}$$

where ε_s is the surface energy required forming the impurity atom and ε_e is the elastic energy involved with inserting an impurity atom into a matrix lattice site. This is given by:

$$\varepsilon_f = \frac{0.5\varepsilon_s}{1.94} + \frac{8\pi G}{3e} a_m (a_i - a_m)^2 eV, \tag{7}$$

where ε_s is the surface energy (1.02 J/m^2), e is the electronic charge ($1.60217646 \times 10^{19}$ Coulomb), a_i is the impurity atomic radius (0.118 nm for Si), a_m is the matrix atomic radius (0.143 nm for aluminium) and G is the shear modulus (26 GPa for aluminium).

By performing the calculations the impurity formation energy, ε_f, for A359 aluminium alloy (Al–Si–Mg) can be determined and then substituted in equation (5) to calculate B (Zuchovitsky).

4. Results and Discussion

The micro-mechanics model, based on thermodynamics principles, is used to determine the fracture strength of the interface at a segregated state in MMCs. This model uses energy considerations to express the fracture toughness of the interface in terms of interfacial critical strain energy release rate and elastic modulus. The interfacial fracture toughness is further expressed as a function of the macroscopic fracture toughness and mechanical properties of the composite, using a toughening mechanism model based on stress transfer mechanism. Mechanical testing is also performed to obtain macroscopic data, such as the fracture strength, elastic modulus and fracture toughness of the composite, which are used as input to the model. Based on the experimental data and the analysis, the interfacial strength is determined for SiC particle-reinforced aluminium matrix composites subjected to different heat treatment processing conditions and the results are shown in Table 2. It is observed that K_{int} values are close to the K_{1c} values of the composites. Furthermore, σ_{int} values found to be dependent on the heat treatment processing with T6 heat treatment composite obtain the highest interfacial fracture strength.

5. Conclusions

A method of calculation has been applied to predict the interfacial fracture strength of aluminium, in the presence of silicon segregation. This model considers the interfacial energy caused by segregation of impurities at the interface and uses Griffith crack-type arguments to forecast the energy change in terms of the coincidence site stress describing the interface and the formation energies of impurities at the interface. Based on Griffith's approach, the fracture toughness of the interface was expressed in terms of interfacial critical strain energy release rate and elastic modulus. The interface fracture toughness was determined as a function of the

Table 2.
Interfacial fracture strength of Al/SiC$_p$ composite

Condition	K_{int} (MPa $\sqrt{m}$)	$E_i \cong E_m$ (N/m^2)	T (K)	c	D (μm)	B	ε_f	N	ε_p (J/m^2)	σ_{int} (MPa)
T1	22.4	7.1×10^{10}	300	0.1	17	1.5	0.303	10	1.42	94
T6	29.5	7.1×10^{10}	803.15	0.1	17	1.5	0.303	10	3.91	260
HT1	26.3	7.1×10^{10}	723.15	0.1	17	1.5	0.303	10	3.55	236

macroscopic fracture toughness and mechanical properties of the composite using two different approaches, a toughening mechanism model based on crack deflection and interface cracking and a stress transfer model. The model shows success in making prediction possible of trends in relation to segregation and interfacial fracture strength behaviour in SiC particle-reinforced aluminium matrix composites. The model developed here can be used to predict possible trends in relation to segregation and the interfacial fracture strength behaviour in metal matrix composites. The results obtained from this work conclude that the role of precipitation and segregation on the mechanical properties of Al/SiC_p composites is crucial, affecting overall mechanical behaviour.

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