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Modeling the fracture toughness and tensile ductility of SiC_p/Al metal matrix composites

Min Song [a,b,*], Daihong Xiao [b]

[a] *Key Laboratory of Materials Physics, Ministry of Education, Zhengzhou University, Zhengzhou 450052, China*
[b] *State Key Laboratory of Powder Metallurgy, Central South University, Changsha 410083, China*

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Abstract

This paper develops a mechanical model to describe the fracture toughness and tensile ductility of SiC_p/Al metal matrix composites based on Eshelby method and Weibull distribution. It has been shown that the model predictions are in reasonable agreement with the experimental counterparts. The fracture toughness and tensile ductility of the composite are lower than the comparative alloys, and decrease as the volume fraction of the SiC particles increases. The size of the SiC particles also has effects on the tensile ductility and fracture toughness. The SiC particles will initiate microcracks during deformation, and thus decrease the fracture toughness and tensile ductility of the composites.

Keywords: Metal matrix composite; Fracture toughness; Tensile ductility; Modeling

1. Introduction

SiC_p/Al metal matrix composites are widely used as structural materials since they have relatively high strength and elastic modulus compared to the comparative aluminum alloys [1,2]. However, the low tensile ductility and fracture toughness limit their further applications. Thus, lots of work [3–5] has been performed to study the fracture mechanism of the SiC_p/Al metal matrix composites. It has been shown that the SiC particles are brittle and will initiate microcracks during deformation, thus decrease the tensile ductility and fracture toughness of the composites.

A previous work [6] developed a fracture model for hydride-induced embrittlement in Zr alloy. In that work, all the hydride precipitates along the grain boundaries have been treated as the pre-existed microcracks. Our previous work [7] developed a model for the fracture toughness of an aluminum alloy. In that work, all the constituents in the alloy have been treated as pre-existed microcracks. However, it should be noted that in SiC reinforced Al alloy composites, not all of the SiC particles will crack and initiate microcracks since SiC particles have high fracture strength. A previous study [8] indicated that the cracking fraction of the SiC particles during deformation can be described by Weibull distribution. In this paper, a model has been developed to calculate the tensile ductility and fracture toughness of the SiC_p/Al metal matrix composites based on Weibull distribution, Eshelby method [9] and previous studies [6,7].

2. Model development

Fig. 1(a) is a typical micrograph showing the microcracks caused by SiC particles after straining to fracture in a SiC/6013Al composites (20 μm in size and 30% in volume fraction). Since the SiC particles will fracture readily under loading, they are the initiations of the microcracks. However, it should be noted that not all the SiC particles will crack during deformation, this will be discussed lately. For simplicity, we assume that the microcracks caused by the SiC particles are distributed uniformly, as shown in Fig. 1(b).

First, we assume that A and B are two neighboring microcracks caused by the SiC particles cracking during deformation. For simplicity, assume that the microcracks distribute uniformly with dimension of 2a and interspacing of L. Thus, at a distance, r (r is the distance between microcracks A and B), ahead of microcrack A, the strain tensor caused by microcracks A and B

* Corresponding author at: State Key Laboratory of Powder Metallurgy, Central South University, Changsha 410083, China. Tel.: +86 7318877880; fax: +86 7318710855.

E-mail address: Min.Song.Th05@Alum.Dartmouth.ORG (M. Song).

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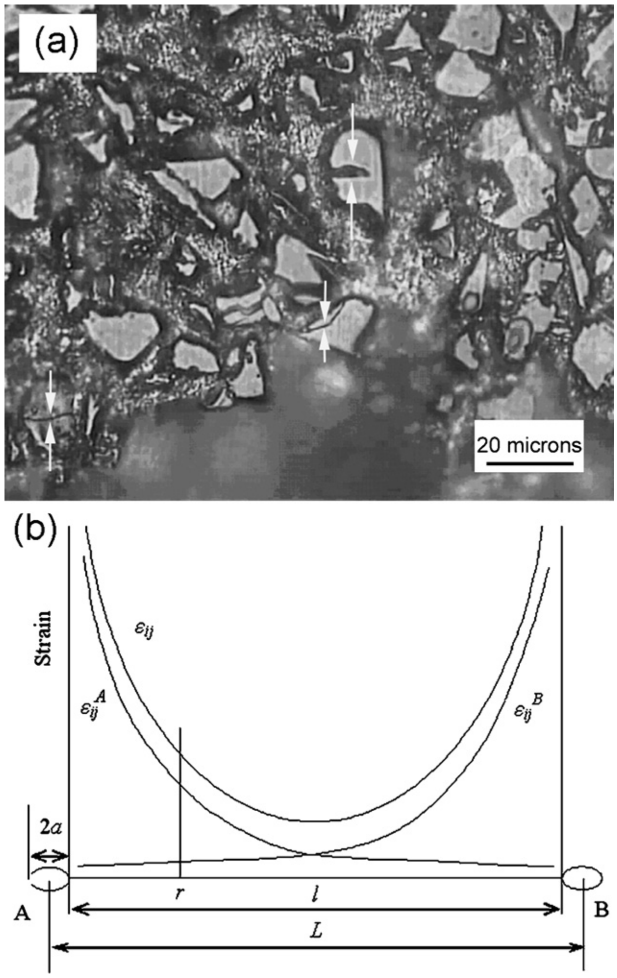

Fig. 1. (a) An optical image of the SiC/Al composite after tensile deformation to fracture and (b) geometric model for fracture toughness and tensile ductility.

is given by [10,11]:

$$\varepsilon_{ij}^{\mathrm{A}} = \alpha\varepsilon_{\mathrm{y}}\left[\frac{J_{\mathrm{A}}}{\alpha\varepsilon_{\mathrm{y}}\sigma_{\mathrm{y}}I_n r}\right]^{n/(1+n)}\tilde{\varepsilon}_{ij}(\theta) \tag{1}$$

$$\varepsilon_{ij}^{\mathrm{B}} = \alpha\varepsilon_{\mathrm{y}}\left[\frac{J_{\mathrm{B}}}{\alpha\varepsilon_{\mathrm{y}}\sigma_{\mathrm{y}}I_n(L-2a-r)}\right]^{n/(1+n)}\tilde{\varepsilon}_{ij}(\theta) \tag{2}$$

where J is the J-integral, ε_{y} the yield strain, σ_{y} the yield stress, n the inverse of the strain hardening exponent, α the material constant in the Ramberg–Osgood constitutive relation [12], and I_n and $\tilde{\varepsilon}_{ij}(\theta)$ are the normalized parameter in the HRR-field [10,11].

The J-integral can be separated into elastic (J_{e}) and plastic (J_{p}) components. For an aluminum alloy-based composite, J_{e} is much smaller than J_{p}, and can be ignored. Thus the J-integral is given by [13]:

$$J_{\mathrm{A}} = J_{\mathrm{B}} \approx \frac{0.405\pi h\sigma_{\mathrm{y}}a[\varepsilon_{\mathrm{p}}]^{(1+n)/n}}{[\alpha\varepsilon_{\mathrm{y}}]^{1/n}} \tag{3}$$

where ε_{p} is the plastic strain, $h = 3/2\sqrt{1+3/n}$. Substituting Eq. (3) into (1) and (2), respectively, leads to

$$\varepsilon_{ij}^{\mathrm{A}} = Q\cdot\varepsilon_{\mathrm{p}}\cdot\left[\frac{a}{r}\right]^{n/(1+n)} \tag{4}$$

$$\varepsilon_{ij}^{\mathrm{B}} = Q\cdot\varepsilon_{\mathrm{p}}\cdot\left[\frac{a}{L-2a-r}\right]^{n/(1+n)} \tag{5}$$

where $Q = [0.405\pi h/I_n]^{n/(1+n)}\tilde{\varepsilon}_{ij}(\theta)$. Adding Eqs. (4) and (5) leads to

$$\varepsilon_{\mathrm{p}} = \frac{\varepsilon_{ij}}{Q}\cdot\left[\frac{r}{a}\right]^{n/(1+n)}\cdot\frac{1}{1+[r/(L-2a-r)]^{n/(1+n)}} \tag{6}$$

When the local effective strain, ε_{ij}, at location r in the matrix ligament reaches the matrix fracture strain, ε_{mf}, the plastic strain, ε_{p}, reaches the nominal fracture strain, ε_{f}, or tensile ductility at the onset of fracture. Thus, Eq. (6) changes to

$$\varepsilon_{\mathrm{f}} = \frac{\varepsilon_{\mathrm{mf}}}{Q}\cdot\left[\frac{r}{a}\right]^{n/(1+n)}\cdot\frac{1}{1+[r/(L-2a-r)]^{n/(1+n)}} \tag{7}$$

For a cubic array, the volume fraction (f) of the microcracks is a function of mean radius (a) and the interspacing λ:

$$f_j = 2\pi K\left[\frac{a}{L}\right]^3 \tag{8}$$

where K is the aspect ratio. For microcracks caused by spherical SiC particles, K is 1.

A previous study [8] indicated that the cracking fraction of the SiC particles in composites during deformation satisfies Weibull distribution, thus the cracking fraction of the SiC particles can be expressed as

$$P_{\mathrm{f}} = 1-\exp\left[-\frac{V_{\mathrm{c}}}{V_0}\left(\frac{\sigma_{\mathrm{c}}}{\sigma_{\mathrm{f}}}\right)^m\right] \tag{9}$$

where σ_{c} is the stress in the SiC particles, V_{c} the volume of the SiC particle, σ_{f} the fracture strength of the SiC particles, V_0 the reference volume of SiC particles when the externally applied stress is σ_{f} and m is the Weibull constant.

According to a previous study [9], the stress in the SiC particles, σ_{c}, can be calculated by externally applied stress through Eshelby method as follows:

$$\sigma_{\mathrm{c}} = \sigma_{\mathrm{I}}+\sigma_{\mathrm{A}}+\sigma_{\mathrm{im}} = C_{\mathrm{M}}(\varepsilon^{\mathrm{c}}+\varepsilon^{\mathrm{A}}+\varepsilon^{\mathrm{M}}-\varepsilon^{\mathrm{T}}) \tag{10}$$

where σ_{I}, σ_{A} and σ_{im} are the stress due to the free shape change, externally applied stress and the image stress in Eshelby method, respectively; C_{M} the elastic modulus of the matrix; ε^{A} the plastic strain of the matrix without SiC inclusions under a stress of σ_{A}; ε^{c}, ε^{M} and ε^{T} are the constrained strain, mean matrix strain and free transform strain, respectively (further details of the Eshelby method can be found in reference [9]). ε^{c}, ε^{M} and ε^{T} can be calculated as follows:

$$\varepsilon^{\mathrm{T}} = -[(C_{\mathrm{M}}-C_{\mathrm{I}})[S-f(S-I)]-C_{\mathrm{M}}]^{-1}\left[-\frac{(C_{\mathrm{I}}-C_{\mathrm{M}})\sigma^{\mathrm{A}}}{C_{\mathrm{M}}}\right] \tag{11}$$

$$\varepsilon^{\mathrm{c}} = S\varepsilon^{\mathrm{T}} \tag{12}$$

$$\varepsilon^{\mathrm{M}} = -f(S-I)\varepsilon^{\mathrm{T}} \tag{13}$$

where C_{I} is elastic modulus of the SiC particles, f the volume fraction of the SiC particles, I the identity tensor and S is the