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Analysis on the interfacial shear strength of fiber reinforced titanium matrix composites by shear lag method

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ABSTRACT

Based on shear-lag method, two models for a single fiber push-out test are proposed to evaluate the interfacial shear strength of SiC fiber-reinforced titanium-matrix composites (TMCs). In the models, the effects of specimen parameters (such as specimen thickness and fiber volume fraction) on the interfacial shear strength are considered. The interfacial shear strengths of SCS-6/Ti-24-11 and SCS-6/Ti-15-3 composites are predicted using the model. The predicted results indicate that the models can be reliably used to predict interfacial shear stress of TMCs.

1. Introduction

SiC fiber-reinforced titanium-matrix composites (TMCs) can be widely used in aerospace and automobile industries due to their low density, high performance, high specific strength and stiffness at room and elevated temperatures [1,2]. It is well known that the fiber/matrix interfacial behavior critically influences the thermo-mechanical behavior of TMCs. Various examinations have been developed to evaluate mechanical properties of the fiber/matrix interface, such as pull-out test [3,4], fragmentation test [5,6] and push-out test [7,8]. Among these methods, push-out test has been emerging as an important experimental technique for characterizing the interfacial performance of this kind of composites owing to the relative simplicity of preparing specimens and conducting the experiments. However, there exists high bond strength and residual stresses at the interface of TMCs owing to the strong chemical activation of titanium and the mismatch of thermal expansion coefficients between fiber and matrix. So it is necessary to use thin slices of composites to avoid the fracture of indenter or the crush of fiber [9]. Moreover, for almost all thin slice push-out tests of TMCs, it is observed that interface debonding initiates at the bottom face of the specimen due to thermal residual stresses at the interface [9–13]. This is different from interface debonding in ceramic matrix composites, whose debonding initiates at the top face.

Both the interfacial shear strength and the fracture toughness play important roles in performance of TMCs. Although there have been some simulations or computations on the interfacial fracture energy, in most cases the interfacial behavior has been characterized by the interfacial shear strength and/or the interfacial sliding frictional stress. Originally, the interfacial intrinsic shear strength τ^* (namely, the shear stress required for breaking the interfacial bond) was defined as

$$\tau^* = \tau^{P_{max}} = \frac{P_{max}}{\pi d_f L} \tag{1}$$

where $\tau^{P_{max}}$ is the critical applied shear stress, P_{max} is the maximum load, d_f is the diameter of fiber and L is specimen thickness. Here the shear stress only comes from the applied load P. The equation is widely applied to calculate the interfacial shear strength [14–16]. Subsequently, the interfacial sliding frictional stress was included in Chandra's model [11,12,17], as follows:

$$\tau^{P_{max}} = \tau^* + \mu F \quad F = \begin{cases} -\sigma_r & \sigma_r < 0 \\ 0 & \sigma_r \geq 0 \end{cases} \tag{2}$$

where F and σ_r are radial compressive stress at the interface, and μ is interface friction coefficient. Based on Eq. (2), the interfacial thermal residual shear stress was taken into account by Yuan [13], namely

$$\tau^* = \tau^{P_{max}} - \tau_{fr} + \tau_{rz} \tag{3}$$

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where τ_{fr} is the interfacial sliding frictional stress and τ_{rz} is the interfacial thermal shear stress. By Eq. (3), the interfacial shear strength of SCS-6/Timetal 834 was calculated to be 502 MPa, where $\tau^{P_{max}}$ $(=P_{max}/\pi d_f L$, $P_{max}=31$ N), τ_{fr} $(=P_{fr}/\pi d_f L$, P_{fr} is the applied load against τ_{fr}, and $P_{fr}=15$ N) and τ_{rz} are 162 MPa, 78 MPa and 418 MPa, respectively [18]. Obviously, the calculated result in terms of Eq. (3) is much closer to the simulated result 500 MPa by finite element method [18] than the results in terms of Eqs. (1) and (2), since τ_{rz} is considered. Actually, there exists high τ_{rz} at fiber/matrix interface, and τ_{rz} decreases with the increase of temperature. In addition, during push-out test of TMCs, the τ_{rz} is anti-symmetrically distributed about the middle of the specimen thickness and is in the same direction as $\tau^{P_{max}}$ in the lower part of the specimen, but they are in opposite directions in the upper part of the specimen [13]. Hence, τ_{rz} was used to explain the experimental phenomenon that P_{max} of sigma/Ti-6-4 [19] and $\tau^{P_{max}}$ of SiC/Ti-15-3 and SiC/Ti-24-11 [20] increase initially as temperature rises and then decrease, in which the initial increase was attributed to the need of compensating for the reduced τ_{rz} near the bottom face of the specimen.

Unfortunately, $\tau^{P_{max}}$ and the interfacial sliding frictional stress obtained from formulas (1–3) are just average stresses. In fact, the interfacial shear stress is nonuniform along fiber axis during push-out test. Additionally, the models (1,2) supposed that the debonding initiates at the top face of specimens [11,12,14–17]. However, for almost all thin slice TMCs specimens, interface debonding is likely to initiate at the bottom face during push-out test.

In this paper, therefore, two analytical models for interfacial shear strength are presented, which include the terms of the interfacial shear stress $\tau_b^{P_{max}}(z)$ induced by P_{max} in the bonded interface, the interfacial sliding frictional stress $\tau_d(z)$ in the debonded region, and thermal residual shear stress τ_{rz}, where the subscripts b and d represent the bonded (continuous) and debonded regions, respectively, and z is along fiber axis. The interface crack initiates at bottom face of the specimen, and propagates from the bottom face to the top face. The models take into account the parameters, E_f, E_m, υ_f, υ_m, specimen thickness L, fiber volume fraction v, μ, $\tau_d(z)$, τ_{rz}, where E and υ represent Young's modulus and Poisson's ratio, respectively, and the subscripts m and f represent matrix and fiber, respectively. The interfacial sliding frictional stress in the debonded region is considered to be a constant τ_0 or a combination of constant τ_0 and the effect of Poisson contraction of the fiber. Two expressions of τ^* are deduced based on shear-lag method. In addition, τ^* of the composites such as SiC/Ti-24-11 and SiC/Ti-15-3 are predicted by the models.

2. Development of the models

2.1. Synopsis for the models

In the analytical model, it is assumed that the fiber/matrix interface is perfectly bonding before loading, and there is no spontaneous debonding caused by thermal residual stresses. The geometry of the fiber/matrix cylinder model is shown in Fig. 1, in which a fiber with a radius r_f is embedded at the center of a coaxial cylindrical shell of the matrix with a radius r_m and a total length L. A set of cylindrical coordinates (r, θ, z) is employed, where the z-axis corresponds to the fiber axis and r is the perpendicular distance to the z-axis. It is assumed that the deformation is symmetric about the fiber axis (i.e. axisymmetric), thus the stress components $(\sigma^r, \sigma^\theta, \sigma^z, \tau^{rz})$ and the displacement components (u^r, u^z) are independent of the tangential coordinate θ, and the remaining stress and displacement components are all zero. At the

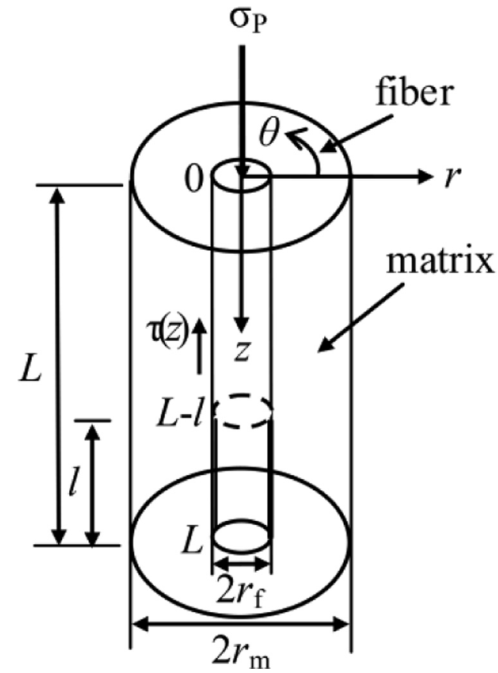

Fig. 1. Idealized fiber push-out model.

top face of the specimen, $z=0$, the fiber is loaded by a force P, and at the bottom face, $z=L$, the matrix is fixed. Therefore, for perfectly elastic and isotropic fiber and matrix, the general relationships between strains and stresses are

$$\varepsilon_f^z(r,z) = \frac{1}{E_f}\{\sigma_f^z(r,z) - \upsilon_f[\sigma_f^r(r,z) + \sigma_f^\theta(r,z)]\} \tag{4}$$

for the fiber (i.e. $0 \le r \le r_f$), and

$$\varepsilon_m^z(r,z) = \frac{1}{E_m}\{\sigma_m^z(r,z) - \upsilon_m[\sigma_m^r(r,z) + \sigma_m^\theta(r,z)]\} \tag{5}$$

$$\varepsilon_m^{rz}(r,z) = \frac{1}{2}\frac{\partial u_m^z(r,z)}{\partial r} = \frac{1+\upsilon_m}{E_m}\tau_m^{rz}(r,z) \tag{6}$$

for the matrix (i.e. $r_f \le r \le r_m$), where ε, σ and τ represent strain, normal stress and shear stress, respectively. In Eq. (6) for the matrix shear strain, compared with the axial displacement gradient with regard to the r-direction, the radial displacement gradient with regard to the z-direction is neglected. In the r-direction, the axial stresses in the fiber and the matrix are assumed to be the average stresses to simplify analysis [21,22], i.e.

$$\sigma_f^z(z) = \frac{2}{r_f^2}\int_0^{rf} \sigma_f^z(r,z)rdr \tag{7}$$

$$\sigma_m^z(z) = \frac{2\gamma}{r_f^2}\int_{rf}^{rm} \sigma_m^z(r,z)rdr \tag{8}$$

where γ $(=r_f^2/(r_m^2 - r_f^2))$ is the volume ratio of the fiber to the matrix. It is assumed that a fiber is embedded in a relatively large matrix as in a practical fiber push-out experiment. As expected, in reality, the variation of the axial (matrix) stress with regard to the radial direction is substantial only near the fiber free end (or near the debonded crack tip for partial debonding) where all stress components are concentrated. This assumption incorporating Lame solutions had been successfully presented to predict the matrix axial deformation by Hutchinson [23] and Freund [24]. The internal stress is transferred from the fiber to the surrounding matrix through the interfacial shear stress. The applied