Effect of fiber coating on the fatigue crack initiation and multiplication of unnotched SCS-6/Ti₃Al composites

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Abstract

The effect of fiber coating on the fatigue crack initiation and multiplication of an unnotched SCS-6 fiber-reinforced titanium aluminate (Ti₃Al) matrix composite was investigated. The initiation and growth of each individual matrix crack, the evolution of matrix crack density and stiffness degradation of the composite with Ag/Ta-coated fibers were characterized quantitatively, and compared with those of the composite with uncoated fibers. The results showed that the composite with Ag/Ta-coated fibers exhibited a shorter crack initiation life, higher number of matrix cracks, and higher crack growth rate than the composite with uncoated fibers. Therefore, a higher matrix crack density and a lower residual stiffness were found in the composite with coated fibers during fatigue loading. The high interfacial cracking density and high interfacial bond strength in the Ag/Ta-coated SCS-6/Ti₃Al composite is believed to be responsible for its poor fatigue damage resistance. Finally, an interface-controlled fatigue cracking model is used to predict the evolution of matrix crack density and residual stiffness of the composite under fatigue loading. © 1999 Elsevier Science S.A. All rights reserved.

Keywords: Titanium matrix composites; Interfacial bond strength; Crack initiation; Fatigue

1. Introduction

Continuous SiC fiber-reinforced titanium aluminate (Ti₃Al) matrix composite has been considered as a structural material for high temperature aerospace applications [1]. The composite has excellent tensile strength and creep resistance along the fiber direction. However, the low transverse mechanical properties and poor transverse creep resistance due to the low interfacial bond strength at the fiber–matrix interface, are deleterious to the performance of the composite [2]. Furthermore, the brittle interfacial reaction layer, if thick enough, leads to premature interfacial and matrix cracking, and consequently, degrades the mechanical properties of the composite [3]. Therefore, several new interface systems with different fiber coatings were developed to improve interfacial compatibility [4,5].

Among these coating concepts, the Ag/Ta duplex fiber coating was shown to be an effective diffusion barrier to limit the interfacial reactions and an effective compliant layer to help accommodate the residual stress [6]. The Ag layer, which is expected to act as a diffusion barrier, is designed to prevent the reaction between Ta and SiC fiber. Meanwhile, Ta, a β-Ti stabilizer, is expected to promote the formation of a ductile β-Ti rich zone through interdiffusion into the matrix during consolidation. This β-Ti rich zone serves as a compliant layer to reduce the thermal residual stresses. Also, it may blunt the cracks which propagate into the matrix. It has been demonstrated that the transverse tensile and flexural creep resistance of the Ag/Ta-coated SCS-6/Ti₂₅-10 composite under longitudinal and transverse loading improved significantly due to its stronger interfacial bond strength, compared with the one with uncoated fibers [7]. However, the stronger interfacial bond strength also reduced the fracture toughness and fatigue crack growth resistance of the composite [8].

To date, the effect of the Ag/Ta duplex fiber coating on the fatigue damage evolution of the unnotched SCS-6/Ti₃Al composite has not been examined. The objective of this work was to quantitatively characterize the fatigue crack initiation and multiplication of the unnotched Ag/Ta-coated and uncoated SCS-6/Ti-25-10 composites. The initiation and growth of each individual matrix crack, and the stiffness degradation as a
function of fatigue cycles were carefully monitored. Also, an integrated analytical model based upon interface-controlled fatigue cracking mechanism was used to predict the evolution of matrix crack density and residual stiffness of the composite under fatigue loading.

2. Materials and experimental procedures

The unidirectional Ag/Ta-coated and uncoated SCS-6 silicon carbide fiber-reinforced Ti–25Al–10Nb–3V–1Mo (at.%) Ti3Al composites were used in this study. SCS-6 has a carbon rich surface coating layer ≈ 3 μm in thickness, which is designed as a sacrificial layer to prevent the Ti alloy from reacting directly with the SiC fiber. The Ag/Ta duplex layer were coated onto the SiC fiber by physical vapor deposition to improve the chemical and physical compatibility between SiC fiber and the Ti3Al matrix. The thickness of the Ag and Ta layers were ≈ 2 and 5 μm, respectively. The composites were fabricated by using a typical foil/fiber/foil approach followed by a vacuum diffusion bonding process. The 8-ply uncoated and the 4-ply Ag/Ta coated SCS-6/Ti-25-10 composites contain ≈ 33 vol.% of fiber. Typical microstructural features near the fiber–matrix interface for both composites with coated and uncoated fibers are shown in Fig. 1(a) and (b). In the uncoated SCS-6/Ti-25-10 composite, a reaction layer was formed after consolidation. The interfacial reaction products have been identified as (Ti,Nb)C1-x(Ti,Nb,Al)5Si3 near the fiber, and (Ti,Nb)3AlC(Ti,Nb,Al)5Si3 near the matrix [9]. The thickness of the interfacial reaction layer was measured to be ≈ 1 μm. In the Ag/Ta-coated SCS-6/Ti-25-10 composite, a residual Ag/Ta layer was found between the carbon-rich layer and the β-rich zone. The thickness of this interfacial layer was measured to be ≈ 2 μm. The mechanical properties of the coated and uncoated composites are listed in Table 1, [4,10].

Fatigue specimens with dimensions of 76.2 × 6.35 mm (length × width) were cut from composite panels using a low speed diamond saw. The free edges of the unnotched specimens were carefully polished to 1 μm to eliminate machining damage. Fatigue tests were performed on a computer-controlled servohydraulic Instron system. All tests were conducted at a stress ratio of 0.1 and a frequency of 10 Hz under a load-controlled mode. The maximum applied stress was chosen to be 600 MPa. The residual stiffness of the specimens were monitored by plotting the stress-strain curves at different fatigue cycles during testing. The fatigue strain was measured by attaching a clip-on extensometer with a gauge length of 12.7 mm onto one surface of the specimen. Both face sections of the specimens were polished to 1 μm in order to measure the matrix crack lengths. The initiation and multiplication of matrix cracks were monitored using cellulose acetate replicating foils which were pasted on the specimen by acetone solution. The replicating foils were then examined under an optical microscope with a travelling stage to

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<th>σth (MPa)</th>
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<th>Interfacial shear strength (MPa)</th>
<th>Frictional stress (MPa)</th>
<th>Post-fatigued frictional stress (MPa)</th>
<th>Fracture toughness (MPa*m1/2)</th>
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measure the number of matrix cracks and matrix crack lengths. Fatigue matrix crack initiation life was defined as the corresponding fatigue cycles when a crack length in the range of 200–400 μm was observed [11]. The matrix crack density, which represented the degree of fatigue damage of the composites, was determined by dividing the total length of matrix cracks by the product of gauge length and width of specimen. The post-fatigued specimens were then subjected to a tensile loading until they failed. The failed specimens within the gauge length were metallographically polished. A scanning electron microscope (SEM) was used to observe the fatigue cracking path, the interfacial cracking, and the fracture surfaces of the post-fatigued composites.

3. Results

3.1. Fatigue matrix crack initiation and multiplication

For unidirectional TMCs subjected to a cyclic loading at a low stress level, multiple matrix cracking with fiber bridging is the major damage mode [12]. Potential matrix crack initiation sites include the damaged fibers and ruptured interface in the composites. For the uncoated SCS-6/Ti-25-10 specimen under fatigue loading at $\sigma_{\text{max}} = 600$ MPa, matrix cracks were found to initiate from the damaged fiber near the free edges, as shown in Fig. 2(a). However, cracks initiated from the ruptured interface near the free edges and in the central region of the Ag/Ta-coated SCS-6/Ti-25-10 specimen, as shown in Fig. 2(b). As the number of fatigue cycles increased, both the number and length of matrix cracks increased, resulting in a progressive stiffness degradation of the composites.

The evolution of matrix crack number as a function of fatigue cycles for the uncoated and Ag/Ta-coated SCS-6/Ti-25-10 composites tested at $\sigma_{\text{max}} = 600$ MPa is shown in Fig. 3. The initiation life of the first matrix crack was found to be $4 \times 10^4$ cycles for uncoated SCS-6/Ti-25-10 and $2 \times 10^4$ cycles for Ag/Ta-coated SCS-6/Ti-25-10. The numbers of matrix cracks in the gauge length after $1 \times 10^6$ cycles were 2 and 12 for the uncoated, and Ag/Ta-coated SCS-6/Ti-25-10 composites, respectively. Surprisingly, the Ag/Ta-coated SCS-6/Ti-25-10 composite showed a shorter matrix crack initiation life and a higher number of matrix cracks compared to the composite with uncoated fibers. As we examined the interfaces of the composites after fatigue loading for $1 \times 10^6$ cycles, severe interfacial cracking was found in the Ag/Ta-coated SCS-6/Ti-25-10 composite, while the interfacial reaction layer in the composite with uncoated fibers remained fairly intact even in the vicinity of the matrix crack plane, as shown Fig. 4(a) and (b). Meanwhile, since the interfacial layer of the composite with Ag/Ta-coated fibers is thicker than that of the composite with uncoated fibers, a higher local stress intensity factor is expected to be induced at the tip of the crack in the interfacial layer. The combining effects of high density of cracks in the interfacial region and high local stress intensity factor of the Ag/Ta-coated SCS-6/Ti-25-10 composite are believed to be responsible for its shorter matrix crack initiation life and higher matrix crack number in comparison with the composite with uncoated fibers.

3.2. Distribution of matrix crack growth rate

For the composites exhibiting multiple matrix cracking under fatigue loading, the growth rate of each individual matrix crack might be different due to the complex interaction between matrix cracks resulting in a different stress field at the tip of each matrix crack. As a result, a distribution of matrix crack growth rates was found in the unnotched Ti$_3$Al composites. Fig. 5
shows the distributions of matrix crack growth rates of the Ag/Ta-coated and uncoated SCS-6/Ti-25-10 composites under fatigue loading at $\sigma_{\text{max}} = 600$ MPa. The uncoated SCS-6/Ti-25-10 composite had only two matrix cracks, with the matrix crack growth rates of 3.6, and 5.4 nm/cycle, respectively. For the Ag/Ta-coated SCS-6/Ti-25-10 composite, most matrix cracks exhibited crack growth rates ranging from 10 to 20 nm/cycle. The average matrix crack growth rates for the uncoated and Ag/Ta-coated SCS-6/Ti-25-10 composites were 4.5, and 15 nm/cycle, respectively. Apparently, the fatigue crack growth resistance of the Ag/Ta-coated SCS6/Ti-25-10 composite is much lower than that of the uncoated SCS-6/Ti-25-10 composite. Fig. 6 shows the tensile fracture surfaces of the post-fatigued Ag/Ta-coated and uncoated SCS-6/Ti-25-10 composites. Obviously, a flat surface with minimum fiber pulled-out was found in the Ag/Ta-coated SCS-6/Ti-25-10 composite, while a significant amount of fiber pulled-out was found in the composite with uncoated fibers. Extensive interfacial debonding along the fiber direction in the composite with uncoated fibers, which acted as a toughening mechanism, should account for its lower averaged crack growth rate compared with the composite with Ag/Ta-coated fibers.

3.3. Evolution of matrix crack density and stiffness degradation

The evolution of matrix crack density and stiffness reduction as a function of fatigue cycles for the Ag/Ta-coated and uncoated SCS-6/Ti-25-20 composites tested at $\sigma_{\text{max}} = 600$ MPa is shown in Fig. 7. Typical three-
regime matrix crack density evolution and stiffness reduction curves were found in both composites. In the early stage, where the number of matrix cracks was still low, the matrix crack density increased slowly with fatigue cycles (Regime I). As the number of fatigue cycles increased, the number of matrix cracks increased, which lead to a rapid increase of the matrix crack density (Regime II). After a long period of fatigue loading, since all the existing matrix cracks became through cracks, and no new crack was initiated, the matrix crack density would reach a saturated value (Regime III). Similarly, the stiffness reduction curve was composed of an initial slow stiffness reduction regime, followed by a steep stiffness drop regime, and then reached a plateau regime with minimal stiffness variation. It is evident that Ag/Ta-coated SCS-6/Ti-25-10 exhibited a much faster matrix crack density increase rate and a higher stiffness reduction rate in regime II than uncoated SCS-6/Ti-25-10. The Ag/Ta-coated SCS-6/Ti-25-10 composite suffered more severe fatigue damage with matrix crack density of 0.80 1/mm after \(1 \times 10^6\) cycles and a residual stiffness of 66%. The uncoated SCS-6/Ti-25-10 composite had a matrix crack density of 0.07 1/mm after \(1 \times 10^6\) cycles with 93.8% residual stiffness. It is clear that a higher matrix crack density results in a lower residual stiffness of the composite. Since the matrix crack density was determined by the total length of matrix cracks, the number of matrix cracks as well as the matrix crack growth rate are dominant factors determining the matrix crack density evolution of the composites. The Ag/Ta-coated SCS-6/Ti-25-10 composite with a higher number of matrix cracks and a faster average crack growth rate than the composite with uncoated fibers, therefore, showed a higher matrix crack density and a lower residual stiffness.

4. Discussion

The Ag/Ta duplex fiber coating was designed to prohibit the interfacial reaction between the SiC fiber and Ti3Al matrix, and to produce a ductile \(\beta\)-rich zone around the fiber. The interfacial region in the Ag/Ta-coated SCS-6/Ti-25-10 composite consists of three distinctive layers, a c-rich layer, a residual Ag/Ta layer and \(\beta\)-rich zone Fig. 1. It is expected that these ductile, compliant interphases will suppress crack initiation near the interfacial region during mechanical/thermal cycling. However, the residual Ag/Ta layer appears to have been embrittled during composite processing. This is substantiated by the occurrence of extensive cracking of this layer during fatigue loading as shown in Fig. 4(b). Similar interfacial cracking pattern was found in the uncoated SCS-6/Ti composite with a thick, brittle reaction layer [13]. Elemental analysis reveals the presence of Ag, Ta, Ti, Si, and Nb in this layer. The embrittlement of the residual Ag/Ta layer may be attributed to the interdiffusion between the fiber and matrix. Therefore, it appears that Ag may not be an effective diffusion barrier to prevent the interdiffusion of atomic species across the interface, leading to the formation of complex carbides and silicides. The exact composition and microstructure of the residual Ag/Ta layer needs to be further identified by transmission electron microscopy (TEM).
Due to the presence of multiple cracking in the brittle layer at the interface, a shorter fatigue crack initiation life and a higher number of matrix cracks were found in the Ag/Ta-coated SCS-6/Ti-25-10 composite than those in the composite with uncoated fibers. The results are consistent with an earlier study by Her et al. that a composite with a thicker interfacial reaction layer exhibited a higher crack density at the reaction layer during fatigue loading [13]. Besides the higher number of matrix cracks, the crack growth rate in the unnotched Ag/Ta-coated SCS-6/Ti-25-10 composite was found to be much higher than that of the uncoated SCS-6/Ti-25-10 composite. A similar result has been found in the notched Ag/Ta-coated and uncoated SCS-6/Ti-25-10 composites [8]. The poor fatigue crack growth resistance of the Ag/Ta-coated SCS-6/Ti-25-10 composite may be attributed to its high interfacial bond strength (140 MPa), which prevent interfacial debonding along the longitudinal direction. On the other hand, the uncoated SCS-6/Ti-25-10 composite, having a weaker interfacial bond strength (115 MPa), showed extensive interfacial debonding and exhibited better fatigue crack growth resistance. The combined effects of a higher number of matrix cracks and a higher crack growth rate result in more severe fatigue damage and a lower residual stiffness in the Ag/Ta-coated SCS-6/Ti-25-10 composite.

Since the fatigue matrix crack initiation and multiplication in the Ag/Ta-coated SCS-6/Ti-25-10 composite were enhanced by the cracking of the interface, an interface-controlled fatigue cracking model developed by Her et al. is used to predict the evolution of matrix crack density and residual stiffness of the composite. The model which incorporates the strain-life equation, the modified fiber bridging model by Marshall et al., and a modified shear-lag model has been successfully used to model the interface-controlled fatigue cracking of the Ti matrix composites [14]. The evolution of matrix crack density and residual stiffness of the Ag/Ta-coated SCS-6/Ti-25-10 composite are predicted as shown in Fig. 8. The predicted results agree quite well with the experimental results.

This study along with earlier works [4,6–8], clearly illustrate the complexity of designing an appropriate fiber coating to improve the chemical compatibility and bond strength of the interfacial region in Ti and Ti3Al composites. Thus far, the selections of candidate fiber coating have been based solely upon the criterion of chemical compatibility. This is simply because thermodynamic data and calculation methodologies exist to provide an adequate basis for selection. However, there is currently inadequate fundamental understanding of the chemical and microstructural features which influence interfacial bonding to guide the selection of candidate fiber coating [1]. Generally, a weak interfacial bonding is desirable for allowing crack deflection and fiber bridging to enhance damage tolerance. However, a higher interfacial strength is necessary to achieve the mechanical properties (primarily transverse tensile and creep strength) required for engine applications. Clearly, more analytical and experimental efforts are needed to tailor an optimum interfacial bond strength to achieve a combination of balanced properties in the Ti matrix composites.

5. Conclusions

1. The Ag/Ta-coated SCS-6/Ti-25-10 composite shows a shorter matrix crack initiation life, faster fatigue crack growth rate, higher matrix crack density, and lower residual stiffness compared to the composite with uncoated fibers.
Fig. 7. The evolution of matrix crack density and stiffness reduction as a function of fatigue cycles for Ag/Ta-coated and uncoated SCS-6/Ti-25-10 tested at $\sigma_{\text{max}} = 600$ MPa.

Fig. 8. The experimental and predicted results of the evolution of matrix crack density and stiffness reduction as a function of fatigue cycles for Ag/Ta-coated SCS-6/Ti-25-10 tested at $\sigma_{\text{max}} = 600$ MPa.

2. The higher density of cracks in the interfacial region and stronger interfacial bond strength of the Ag/Ta-coated SCS-6/Ti-25-10 composite are responsible for its more severe fatigue damage in comparison with the composite with uncoated fibers.

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References