FRACTURE OF Ti/SiC COMPOSITES AT ROOM AND HIGH TEMPERATURE: EXPERIMENTS AND SIMULATIONS

C.González, J.LLorca and A.Weck
Department of Materials Science. Polytechnic University of Madrid. Spain
Department of Materials Science and Engineering. Mc Master University. Ontario, Canada

ABSTRACT

The fracture behavior of Ti-6Al-4V uniaxially reinforced with 35 vol. % SiC Sigma 1140+ fibers was studied between 20ºC and 550ºC by three-point bend tests of notched beams. It was found that the fracture energy remained essentially constant in the whole temperature range while the initial toughness decreased linearly with temperature from 78 MPa√m at 20ºC to 44 MPa√m at 550ºC. Fracture occurred by the development of a thin fracture process zone propagated from the notch root where the matrix plastic deformation was localized. The fracture of the composite panels was simulated by representing the fracture process zone by a cohesive crack. The numerical simulations were in good agreement with the experimental results and showed that the reduction of the initial fracture toughness with temperature was associated with the critical condition to reach the maximum load. Below 200ºC the critical condition was attained when the crack opening displacement at the notch tip reached the matrix critical crack opening displacement. Above 200ºC, the maximum load was dictated by the crack opening displacement at the notch tip which led to the fracture of all the fibers ahead of the notch tip.

1. INTRODUCTION

The fracture behavior of Ti-6Al-4V uniaxially reinforced with 35 vol. % SiC Sigma 1140+ fibers was studied between 20ºC and 550ºC by three-point bend tests of notched beams. Fracture occurred by the development of a thin fracture process zone propagated from the notch root where the matrix plastic deformation was localized. The fracture of the composite panels was simulated by representing the fracture process zone by a cohesive crack. The corresponding cohesive law was described by a new micromechanical model and the fracture behavior was simulated numerically using the finite element method. The simulations were in good agreement with the experimental results and showed that the reduction of the initial fracture toughness with temperature was associated with the critical condition to reach the maximum load. Below 200ºC the critical condition was attained when the crack opening displacement at the notch tip reached the matrix critical crack opening displacement. Above 200ºC, the maximum load was dictated by the crack opening displacement at the notch tip which led to the fracture of all the fibers ahead of the notch tip.

2. MATERIALS AND EXPERIMENTAL TECHNIQUES

Composite panels of a Ti-6Al-4V alloy uniaxially reinforced with 35 vol.% Sigma 1140+ SiC fibers were used in this investigation. The panels were consolidated in vacuum at 940

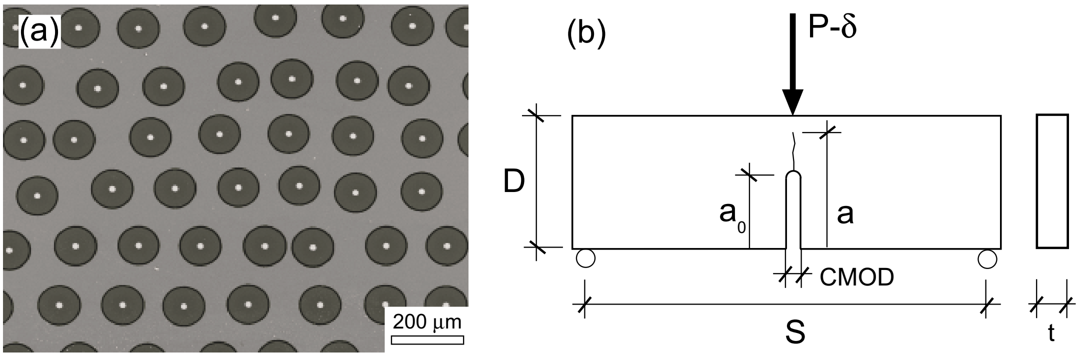

Figure 1. (a) SEM image of the cross section of the composite panel. (b) Geometry of the TPB specimens used in the fracture tests.

Finally, the fractured specimens were examined by scanning electron microscopy to ascertain the failure micromechanisms in the composite due to deformation. The observation of the damage at the notch root with the telescope showed that the damage was initially concentrated in a narrow strip ahead of the notch. The plastic deformation of the matrix in the strip was intense, and specimen fracture occurred by the propagation of a crack within the strip. The average width of this region at ambient temperature was ≈1.4 mm, which was very similar to the panel thickness. Evidence of fiber bridging and pull-out in the crack wake was also observed, Figure 2(c). These tests provided two important parameters to characterize the fracture resistance of these composites as a function of temperature. The first one was the composite fracture energy, G_F, computed as the area under the P-δ curve divided by the area of the initial ligament, $(D-a_0)t$, t being the specimen thickness. The second was the apparent fracture toughness, K_Q, computed from the maximum load in the fracture test, P_{max} and the initial notch length as (assuming linear elastic fracture mechanics).

3. MODELING

The experimental observations described above showed that the crack propagation in the composite panels followed the development of a thin strip with intense plasticity ahead of the crack tip, where the failure processes (fiber fracture, matrix/fiber sliding, and ductile matrix fracture) took place. This scenario is very similar to that found in the fracture of thin metal sheets, where the fracture resistance was modeled successfully by assuming that the fracture process zone behaved as a cohesive crack capable of transferring stresses between the crack surfaces. The critical factor in this approach is the bridging law which determines the relationship between the cohesive stresses and the crack opening displacement of the cohesive crack. The bridging law is determined from the analysis of the fracture micromechanisms within the fracture process zone, and the corresponding parameters are obtained from the properties of the matrix, fibers, and interface as a function of temperature [2],

$$\sigma(w) = f\sigma_f(w) + (1-f)\sigma_m(w) \tag{1}$$

The matrix contribution to the bridging stress can be considered constant and equal to the matrix flow stress, σ_{my}, and will vanish when the cohesive crack opening reaches w_c the critical crack opening displacement which leads to the matrix rupture.

Once the matrix stresses acting on the crack surface have been included, the fiber contribution can be accounted for by using models developed for brittle-matrix composites, where the crack propagates through the matrix leaving the intact fibers behind the crack tip. The fiber-matrix interaction takes place by the relative sliding between them, which is due to elastic matrix unloading in brittle-matrix

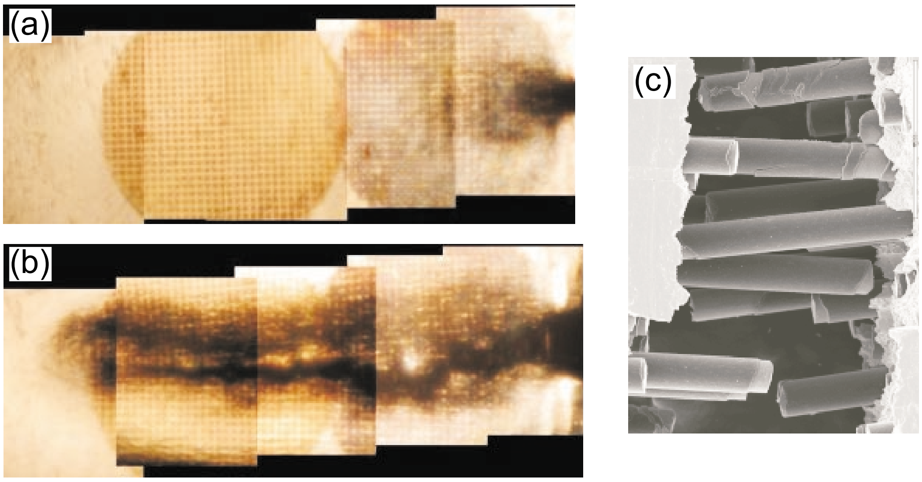

Figure 2.(a) and (b) Optical images of the fracture process zone. (c) SEM image of the crack showing fiber bridging and pull-out.

composites and to matrix plastic yielding in ductile-matrix composites. The bridging stress provided by the intact fibers, σ_b, as a function of the crack opening displacement in the matrix, w, on the assumptions that the fiber-matrix bonding energy was negligible and that the interaction between matrix and fibers was via constant frictional sliding at the interface, τ [3]. Imposing force equilibrium on the fiber and the compatibility of deformations between the matrix and the fiber at the end of the slipping zone yields

$$\sigma_b = \left[\frac{2\tau E_f (1+\eta)}{R} w\right]^{1/2} \text{ with } \eta = \frac{fE_f}{(1-f)E_m} \tag{2}$$

According to equation (2), the stress in the fibers increases with the square root of the crack opening displacement leading to fiber failure. Fractured fibers still contributed to the bridging stresses as they have to be pulled out from the matrix. The relative contribution of intact fibers, which act as elastic ligaments between the crack faces, and broken ones within the matrix, which are eventually pulled out, was analyzed assuming that the fiber strength followed Weibull statistics [4].

The bridging law developed above from micromechanical considerations depends on the matrix, fiber and interface properties which vary with temperature. The properties for the Ti-6Al-V matrix, the SiC Sigma fibers and the interface and their evolution with temperature were determined experimentally or taken from well stablished values of literature [2]. The crack bridging laws as a function of temperature are plotted in Figure 3(b). The stress transmitted through the crack was initially controlled by the elastic deformation of the fibers, which bridged the crack, and rose sharply to a maximum. Fiber failure was immediately triggered and the cohesive stresses decreased rapidly as the fiber pull-out contribution to the bridging stresses was significantly lower. Once all the fibers are broken, the bridging stress remained constant until the critical matrix crack opening displacement, w_c, was reached. This parameter was estimated by comparing the average fracture energy measured in the fracture tests at different temperatures (94.5 kJ/m^2) with that predicted by the micromechanical model, which is given

$$G_F = \int_0^{W_C} \sigma(w)dw \tag{3}$$

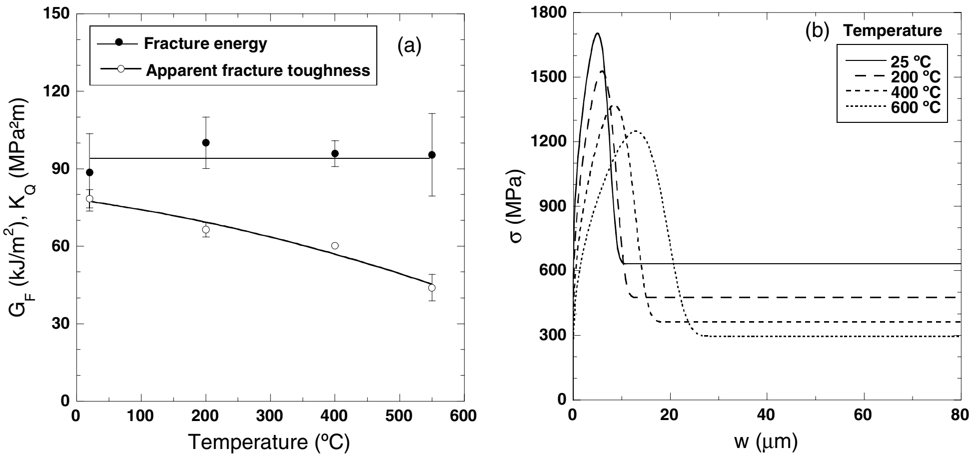

Figure 3 (a). Fracture energy and apparent fracture toughness of the TiSiC composite. (b) Cohesive curves as a function of temperature.

The fracture tests were simulated by finite element analysis of the notched beams. Only one-half of the beams was discretized with isoparametric four-node quadrilaterals. The composite panel was modeled as a linear elastic, transversally-isotropic solid. The cohesive crack was introduced with uniaxial spring elements inserted at the symmetry plane whose force-elongation behavior followed the bridging curves in Figure 3(b). The computed load vs CMOD curve at ambient temperature is plotted together with the experimental ones in Figure 4(a) for the specimen with a_0/D=0.35. The initial slopes of both curves were in excellent agreement but some noticeable differences arose around the maximum load. These experimental results also showed that the presence of a notch did not modify the tearing modulus, i. e., once a real crack began to propagate, the fracture resistance was independent of the notch radius. In agreement with this conclusion, the experimental curves after the maximum load (which are controlled by the tearing modulus) were close to the numerical predictions of the cohesive crack model.

The numerical predictions for the apparent fracture toughness, K_Q, which are compared in Figure 4(b) with the experimental data. While the experimental data showed an approximately linear reduction of K_Q with temperature, the numerical simulations predicted two different regimes. K_Q decreased from room temperature up to 200ºC and then remained essentially constant up to 550ºC. In the low temperature regime, the maximum load was attained when the crack opening displacement at the notch tip reached the critical one and the full potential of the ductile matrix to dissipate energy was used. On the contrary, P_{max} above 200ºC was reached at much lower values of the crack opening displacement at the notch tip, which coincided with the magnitude of w necessary to break all the fibers ahead of the notch tip, and the toughening potential of the ductile matrix was not achieved. K_Q was mainly independent of the temperature in this second regime.

4. CONCLUSIONS

Fracture of the composite panels was simulated through a cohesive crack model approach, which assumed that the deformation associated with the narrow fracture process zone was localized into a fictitious crack which was able to transfer stresses. The corresponding bridging law was determined by considering the contributions of the matrix and the fibers separately. The former was assumed to provide a constant bridging stress equal to the matrix flow stress while the stress transferred by the fibers was the sum of the elastic stresses induced by the intact fibers which bridge the crack and those broken which are pulled out as the crack opens. They were computed

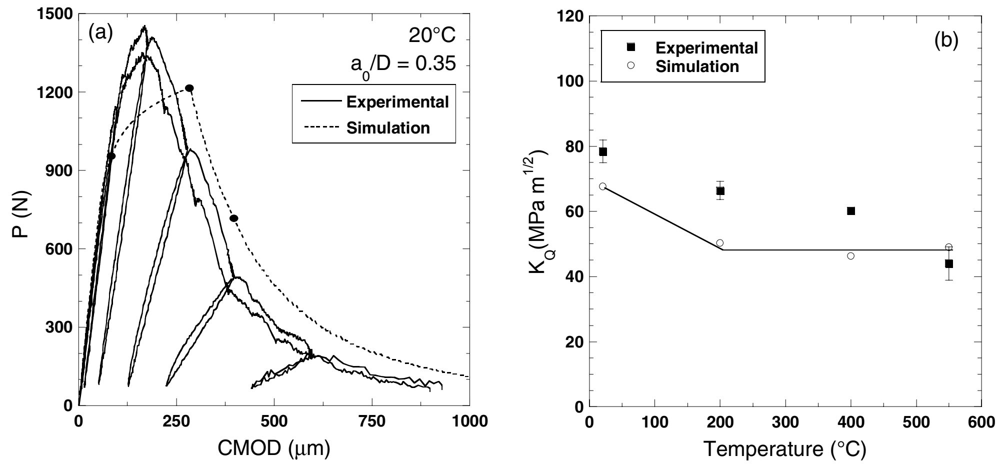

Figure 4 (a). Numerical and experimental results of the P-CMOD curves at room temperature for a beam with $a_0/D=0.35$. (b) Numerical and experimental apparent fracture toughness, K_Q, as a function of temperature.

with a micromechanical model, based on previous studies of brittle-matrix composites, which assumes that the interfacial fracture energy is negligible and that the interaction between matrix and fibers occurred via constant frictional sliding at the interface. The critical parameters which determine the bridging stresses as a function of temperature were measured independently or taken from well-established values in the literature, and the fracture behavior was simulated numerically by the finite element method.

The numerical P-CMOD curves were in reasonable agreement with the experimental ones in the whole temperature range and predicted experimentally observed reduction of the initial fracture toughness with temperature. This phenomenon was due to the progressive drop in the bridging stresses with temperature because of the lower interfacial sliding resistance associated with the release of thermal residual stresses at the interface and to the reduction of the matrix flow stress. It should be noted that all the model parameters were obtained independently, with the exception of the critical value of the matrix crack opening displacement, w_c, which was computed from the average value of the fracture energy measured from that spent in breaking the samples. This parameter is related to the thickness of the fracture process zone under plane stress conditions and depends very probably on the panel thickness and could not be obtained from the micromechanical considerations which led to the bridging law for the cohesive crack.

5. REFERENCES

[1] C.González, J. LLorca, Micromechanical modeling of deformation and failure in Ti-6Al-4V/SiC composites, Acta Materialia 49 (2001) 3505-3519.

[2] C.González, J. LLorca, A.Weck, Toughness of fiber-reinforced titanium as a function of temperature: experimental results and micromechanical modeling, Acta Materialia 52 (2004).

[3] D.B. Marshall, B.N. Cox, A.G. Evans, The mechanics of matrix cracking in brittle-matrix fiber composites, Acta Metallurgica 33 (1985) 2013-2021.

[4] M.D. Thouless, A.G. Evans, Effects of pull-out on the mechanical properties of ceramic matrix composites, Acta Metallurgica 36 (1988) 517-522.