

EFFECT OF TEMPERATURE AND STRAIN RATE ON THE INTERFACIAL SLIDING STRESS OF A GCMC

I. Puente, M.R. Elizalde, J.M. Sánchez, A. Martín-Meizoso, J.M. Martínez-Esnaola*

The interfacial sliding stress is a key parameter of the fracture behaviour of CMCs. In this work, its value has been estimated for a laminated CAS/SiC with architecture $[(0/90)_3]_s$, using the ACK prediction, based on experimental measures of the matrix crack spacing. This evaluation has been done for specimens tested in air, in a wide range of temperatures (from 20 to 900° C) and strain rates (from 10^{-4} s^{-1} to 2 s^{-1}). In the light of the obtained results, the evolution of the mechanical behaviour of the material in these conditions is discussed. It is concluded that the variations in the interfacial sliding stress, due to oxidation and degradation of the interface, can explain the behaviour of the material in all the conditions of temperature and strain rate that have been studied.

INTRODUCTION

Continuous fibre ceramic composites (CFCCs) are mainly designed for high temperature structural applications such as turbo-engine components. However, the use of these materials is limited by their intrinsic variability, price and potential embrittlement. For this reason, a systematic study of the mechanical reliability of the CFCCs is still required.

The objective of the present work is to analyse the effect of temperature and strain rate on the interfacial properties of a Nicalon/CAS CFCC with a $[(0/90)_3]_s$ “cross-ply” architecture. The values of the interfacial sliding stress (τ) have been estimated from the matrix crack spacing, using the ACK model (1). The variation of τ with temperature and strain rate has also been studied and compared with the values obtained from nanoindentation techniques after different heat treatment.

* CEIT, Department of Materials, PO Box 1555, 20080 San Sebastián. Spain, and ESII, Universidad de Navarra. PO Box 1674, 20080 San Sebastián. Spain.

EVALUATION OF τ FROM ACK MODEL

The ACK model is a simple analysis, based on energetic considerations, that predicts the matrix crack spacing for composite material with unidirectional reinforcement. According to this model, the minimum saturation space between cracks, X , is given by:

$$X = \frac{V_m \sigma_{mu} r}{V_f 2\tau} \quad (1)$$

where σ_{mu} is the strength of the matrix; V_m and V_f are the matrix and fibre volume fraction, respectively; r is the mean fibre radius and τ is the interfacial sliding stress.

Kimber and Keer (2), based on the ACK model, proposed the following expression for the mean value of the matrix crack spacing, l_{mcs} :

$$l_{mcs} = 1.34 \cdot \left(\frac{V_m}{V_f} \right) \frac{\sigma_{mu} \cdot r}{2 \cdot \tau} \quad (2)$$

The strength of the matrix can be obtained from the matrix micro cracking stress σ_{mc} (3), taking into account the effect of the differences of thermal expansion coefficients between fibre and matrix. A residual stress q must be added to the external stress, so

$$\sigma_{mu} = \sigma_{mc} \frac{E_m}{E_c} + q \quad (3)$$

where E_m and E_c are the elastic modulus of matrix and composite, respectively, and σ_{mc} is obtained from the loss of linearity of the stress-strain curves (4). The value of the residual stress, q , can be analytically obtained from the thermal misfit strain (3), ϵ , as

$$q = E_m \frac{\lambda_2}{\lambda_1} \left(\frac{E_f}{E_c} \right) \left(\frac{V_f}{1 - v_m} \right) \epsilon \quad (4)$$

where v_m is the Poisson's ratio of the matrix and the coefficients λ_1 and λ_2 are

$$\lambda_1 = \frac{1 - \left(1 - \frac{E_c}{E_f} \right) \frac{1 - v_f}{2} + (1 - v_f) \frac{v_m - v_f}{2} - \frac{E_c}{E_f} \left[v_f + (v_m - v_f) \frac{V_f E_f}{E_c} \right]^2}{(1 - v_m) \left[1 + v_f + (v_m - v_f) \frac{V_f E_f}{E_c} \right]} \quad (5)$$

$$\lambda_2 = \frac{1 - \frac{E_c}{E_f} + (1 + v_f) \frac{v_m - v_f}{2}}{1 + v_f + (v_m - v_f) \frac{V_f E_f}{E_c}}$$

and ϵ can be obtained from the difference of temperature between the actual and T_{gt} as

$$\epsilon = (\alpha_m - \alpha_f) \Delta T \quad (6)$$

where α_m and α_f are the thermal expansion coefficients of matrix and fibres, respectively.

Some of the parameters that should be used in the calculation of τ are of geometric nature (V_m , V_f and r); and therefore, are independent of temperature and strain rate, as are the thermal expansion coefficients of fibres and matrix and the glass transition temperature of the matrix (α_m , α_f and T_{gr}). The elastic properties of fibres (E_f and ν_f) are also invariant up to 1000° C (5). Finally, the Poisson's ratio of the matrix (ν_m) has also been considered as a constant. The numerical values of these constants that have been used in this work are gathered in Table 1.

TABLE 1 - Values of the constants used to estimate the value of τ .

Constant	MATRIX			FIBRES		
	Value	Units	Source	Value	Units	Source
Volume fraction	0.825	-	Microst. Obs.	0.175	-	Microst. Obs.
Radius	-	-	-	6.9	μm	Microst. Obs.
Elastic modulus	-	-	-	200	GPa	(6)
Poisson's ratio	0.3	-	(7)	0.2	-	(7)
Thermal expansion coeff.	$5 \cdot 10^{-6}$	$^{\circ}\text{C}^{-1}$	(6)	$4 \cdot 10^{-6}$	$^{\circ}\text{C}^{-1}$	(6)
Glass transition temp.	750	$^{\circ}\text{C}$	(7)	-	-	-

The value of the elastic modulus of the matrix has been estimated from the elastic modulus measured for the whole composite, using the rule of mixtures

$$E_m = \frac{E_c - E_f V_f}{V_m} \quad (7)$$

Finally, the value of the mean matrix crack spacing, l_{mcs} , has been measured on the surface of the specimen, for each condition of temperature and strain rate, by means of a Leinz optical microscope, without polishing, to avoid the damage induced in this process.

RESULTS

In order to evaluate the variation of the interfacial sliding stress, four temperatures (room temperature, 500, 700 and 900° C) and three strain rates (0.0005, 0.01 and 1 s^{-1}) have been selected. At each condition, a set of specimens has been tested, and the elastic modulus of the composite (E_c), the matrix micro cracking stress (σ_{mc}) and the strain energy density (SED) have been measured.

At each condition of temperature and strain rate, the matrix crack spacing has been measured on the specimen which exhibits the value of the absorbed strain energy density closest to the mean value obtained for this condition.

The results obtained from tensile tests are presented in Table 2, together with the mean values of matrix crack spacing and those obtained for the interfacial sliding stress, using Eq. (2) and the procedure described above.

TABLE 2 - CAS/SiC [(0/90)₃]₈: Results of tensile tests, fractographic analysis and evaluation of τ for ten conditions of temperature and strain rate

T ° C	$\dot{\epsilon}$ s ⁻¹	E_c GPa	σ_{mc} MPa	SED MJ/m ³	l_{mcs}		τ MPa
					Mean μm	Standard dev. μm	
25	0.0005	119	62.2	1.43	214	78	7.75 ± 2.07
25	0.01	119	52.0	1.84	200	70	7.35 ± 1.91
25	2	138	56.2	2.18	176	57	9.29 ± 2.27
500	0.0005	105	41.6	3.45	107	25	8.34 ± 1.58
500	2	110	42.4	1.59	176	57	5.27 ± 1.29
700	0.0005	109	43.6	1.31	250	101	3.26 ± 0.94
700	2	103	48.6	1.38	300	138	2.93 ± 0.92
900	0.0005	110	52.0	0.57	3030	3030	0.31 ± 0.15
900	0.01	108	61.2	0.59	333	167	3.28 ± 1.09
900	2	113	81.0	0.89	429	260	3.44 ± 1.30

Finally, the interfacial mechanical behaviour has been characterised from push-in tests on fibres using nanoindentation techniques and the Hsueh model (8). The specimens were heat treated for 100 hours in air at temperatures between 375°C and 800°C and then tested at room temperature. Figure 1 presents the values of the friction coefficient obtained from this set of tests.

DISCUSSION

The interfacial sliding stress is highly dependent on the degradation of the interface and the magnitude of residual stresses. The interface of the CAS/SiC has a very thin (1 μm) carbon layer (7). This carbon layer has a lubricant effect on the interfacial sliding. When the oxidation removes this interfacial carbon layer, at temperatures as low as 375° C, the lubricant effect disappears, and the friction between fibre and matrix increases. If the temperature is higher (above 700° C) the fibre surface can also be oxidised, then increasing de the roughness, and so, the friction (see Figure 1).

On the other hand, increasing the temperature, the residual stresses are reduced, and so, the sliding stress will also be reduced. This effect contrasts with the increase of friction coefficient, and therefore, the effect of temperature on the interfacial sliding stress is very complex. In fact, up to 500° C, the increase of roughness is more important than the decrease of residual stress, and so, the sliding stress increases slightly. Nevertheless, above 500° C, the residual stress are reduced faster than the increase of roughness, and so, the interfacial sliding stress is dramatically reduced.

The effect of strain rate is also intricate. The general tendency is to increase the sliding stress with increasing strain rates, particularly above 0.01 s⁻¹. Nevertheless, when high temperature is also present, increasing the strain rate reduces the time for interfacial

oxidation (the ambient O₂ is less time in contact with the interface), and so, the roughness is reduced. In fact, at 500 and 700° C, the sliding stress is reduced with the strain rate.

At 900° C and 0.0005 s⁻¹, the behaviour of the material is brittle, due to the formation of silica bridges (9-10). In this case, the value obtained for interfacial sliding stress must be neglected, because the sliding on the interface has not occurred.

Finally, a correlation between SED and interfacial sliding stress has been observed (see Figure 2). It has been previously reported (11) that one of the most important contributions to the energy absorption of the material is the matrix microcracking, and so, one may anticipate that this relation is due to the influence of the interfacial sliding stress on the mean matrix crack spacing (1).

ACKNOWLEDGEMENTS

The authors acknowledge Rolls-Royce plc (UK); the “Departamento de Educación, Universidades e Investigación” of the Basque Government (EC9206 and PGV 92/17); the “Dirección General de Electrónica y Nuevas Tecnologías” and the CICYT (Mat93-0348) of the Spanish Government for the financial support of this work.

REFERENCES

- (1) Aveston, J. et al.; in Proc. of the National Physical Laboratory, 1971, pp. 15-21.
- (2) Kimber, A.C. and Keer, J.G.; J. Mater. Sci. Lett, Vol. 1, 1982, pp. 353-354.
- (3) Cao, H.C. and Thouless, M.D.; J. Amer. Ceram. Soc., Vol 73, 1990, pp. 2091-2094.
- (4) Sánchez, J.M. et al.; in Proc. IVth Spanish Conf. on Mec. Properties of Solids, 1994.
- (5) Bunsell, A.R. and Berger, M.H.; Key Eng. Materials, Vol. 127-131, 1997, pp. 15-26.
- (6) Beyerle, D.S. et al.; J. Am. Ceram. Soc., Vol. 75, 1992, pp. 3321-3330.
- (7) Daniel, A.M.; Ph. D. Thesis, Physics Department, University of Warwick, UK, 1994.
- (8) Hsueh, C.H.; J. Am. Ceram. Soc., Vol. 76, 1993, pp. 3041-3050.
- (9) Allen, R.F. et al.; in Proc. ICCM-9, Vol. 2, 1993, pp. 121-128.
- (10) Puente, I.; Ph. D. Thesis, ESII San Sebastian, University of Navarra, Spain, 1997.
- (11) Puente, I. et al.; Mat. Sci. & Tech., 1998, in press.

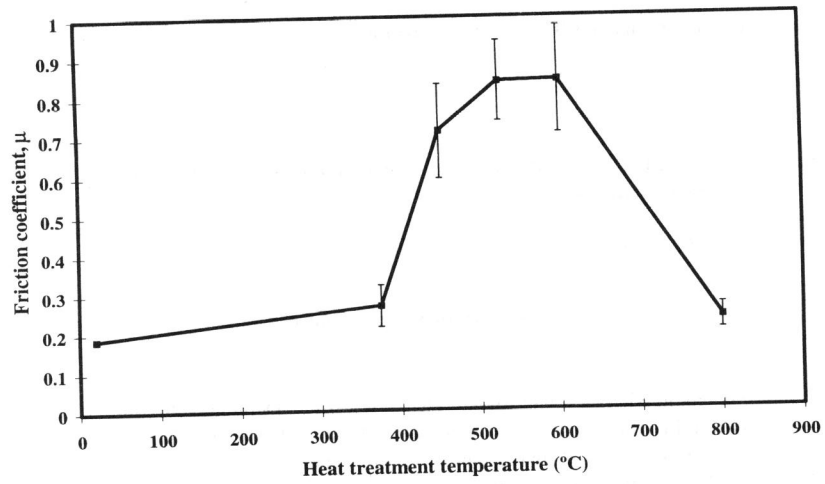


Figure 1. CAS/SiC [(0/90)₃]_s. Interfacial friction coefficient as a function of heat treatment temperature.

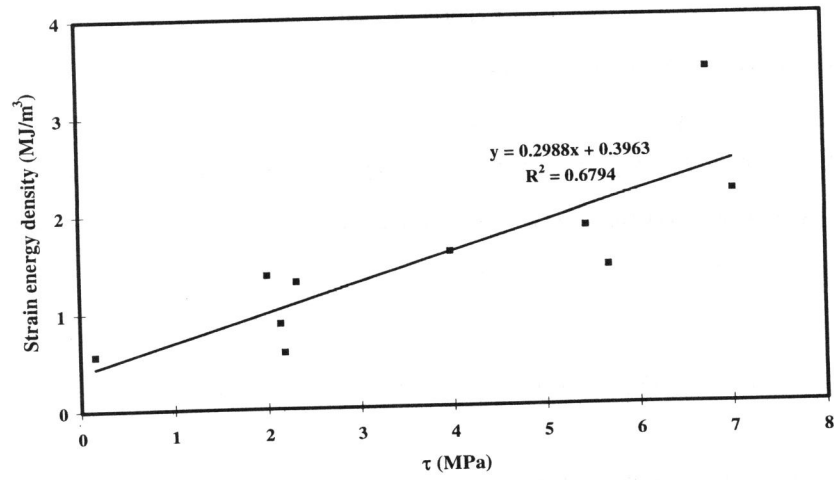


Figure 2. CAS/SiC [(0/90)₃]_s. SED vs. interfacial sliding stress. As it can be shown, a correlation coefficient greater than 0.8 is obtained.