JOURNAL OF MATERIALS SCIENCE 11 (1976) 356–362

Fracture mechanics of metal matrix-metal fibre composites

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A model of the metal matrix–metal fibre composite has been constructed. The critical stress intensity coefficient has been estimated taking into account the increase of the energy absorbing capacity of a fibre surrounded by a plastic matrix due to the increase of the elongation at rupture of the fibre under these conditions. The results of experiments carried out on aluminium matrix–steel fibres composites support the validity of the model. The effect of conditions at the fibre–matrix interface on the fracture toughness of the composite has also been studied.

1. Introduction

Early studies of the fibrous composites revealed an increase of the apparent plastisity of strong metal wires incorporated in a ductile metal matrix [1–3]. This effect was quantitatively explained in terms of transverse fibre–matrix interaction leading to a common point of necking in both components [4].

Gerberich [5, 6] and also McGuire and Harris [7] have measured the critical values of the stress intensity coefficient in metal–metal composites. These authors, however, did not examine a clear possibility of calculating fracture toughness of the composites on the basis of the model considered in [4].

Gerberich assumed a matrix contribution to the work of fracture of a composite according to Cooper and Kelly [8]

$$G_m = \frac{v_m^2}{v_f} \sigma_m^* e_m^* d, \quad (1)$$

where σ_m^* and e_m^* are the ultimate stress and strain of the matrix material, d the fibre diameter, and v_f and v_m the volume fractions of the fibres and matrix. This expression, which is valid strictly speaking only for a composite with non-plastic fibres, was applied to a composite with ductile fibres. At the same time, Gerberich took into account the contribution of the plastic work of the fibres

$$G_f = 2v_f \sigma_f^* e_{f,c}^* d. \quad (2)$$

Gerberich noted that the value of $e_{f,c}^*$, which is the fibre strain at its rupture in the composite, was dependent on the volume fraction of fibres (see Table 1 in [5]), and he measured $e_{f,c}^*$ by direct metallographic observations of specimens which had been tested previously. This means that Gerberich's model cannot be used for the purpose of designing a composite with a particular value of fracture toughness. The importance of such a design stems from experimental data including those obtained by Gerberich, which show that it is possible to obtain low density, very tough materials by combination of two metal components.

McGuire and Harris also investigated the fracture of a metal matrix–metal fibre composite [7]. They assumed the value of the crack extension force for a composite to be

$$G = h_f v_f \sigma_f^* e_{f,c}^* + h_m v_m \sigma_m^* e_{m,c}^* \quad (3)$$

where h_f and h_m are the size of plastically deformed zones in the fibres and the matrix, respectively. The values of h_f and h_m depend on the volume fraction of fibres, and together with the values of $e_{f,c}^*$ and $e_{m,c}^*$ should be measured during testing of the specimens.

The main purpose of the present work is to produce a model composite from which it would

be possible to estimate the fracture toughness of the material without the need for performing any experiments on composite specimens. We have based our work on the composite model considered in a previous paper [4].

2. A model of ductile fracture of a composite

Let us consider a composite with a metal matrix and metal fibres analysed in [4], where a power approximation was taken for the stress–strain curves of the components, and the bond at the matrix–fibre interface was assumed to be ideal. The strain at rupture of the composite and its components $e^* = e^*_{m,c} = e^*_{f,c}$ and the volume fraction of fibres are then related by the following expression:

$$v_f = \left[1 + \beta \frac{e^* - e^*_f}{e^*_m - e^*} e^{*e^*_f - e^*_m}\right]^{-1} \quad (4)$$

where

$$\beta = \frac{\sigma^*_f}{\sigma^*_m} \frac{e^{*e^*_m}_m}{e^{*e^*_f}_f} \frac{\exp e^*_f}{\exp e^*_m}. \quad (5)$$

The ultimate stress (i.e. the maximum on the stress–strain curve of the composite) will be

$$\sigma^* = v_f \lambda_f \sigma^*_f + (1 - v_f) \lambda_m \sigma^*_m \quad (6)$$

where σ^*_f, σ^*_m, e^*_f, and e^*_m are obtained by testing the matrix and fibres separately. The values of λ_f and λ_m are determined in [4] and are generally close to 1.

Let K^*, K^*_f, and K^*_m be the critical values of the stress intensity coefficient for a crack of mode I in a composite, fibre material, and matrix, respectively. These values are assumed to be the ultimate characteristics of the materials, and imply that they have been obtained under the most severe conditions of plastic restraining. The corresponding values of the crack extension force G are such that $K = (CG)^{1/2}$. For a material with relatively small anisotropy (Al–Fe composite, for example) this relation may be replaced by $K = (EG)^{1/2}$, because direct calculation [6] shows the difference between values of E and G to be small (in this case within 17%).

If we now assume that in a particular material the value of rupture strain e^* has changed as a result of the changing of some internal and/or external conditions, and at the same time physical or mechanical properties remaining unchanged, then the value of G will change in proportion to e^*.

This assumption leads to the following expression for the work of fracture of the composite:

$$G = \frac{(K^*_f)^2}{E_f} \frac{e^*}{e^*_f} v_f + \frac{(K^*_m)^2}{E_m} \frac{e^*}{e^*_m} v_m. \quad (7)$$

This expression may be rewritten in terms of the critical stress intensity factor

$$K^* = (E_f v_f + E_m v_m)^{1/2} \left[\frac{(K^*_m)^2}{E_f} \frac{e^*}{e^*_f} v_f + \frac{(K^*_f)^2}{E_f} \frac{e^*}{e^*_m} v_m\right]^{1/2} \quad (8)$$

where the value of e^* is determined by Equation 4.

TABLE I

	$E \times 10^{-3}$ (kgf mm^{-2})	σ^* (kgf mm^{-2})	e^*
Matrix m_1	7.2	30	0.20
Matrix m_2	7.2	50	0.20
Fibre f_1	20.0	250	0.02
Fibre f_2	20.0	350	0.02
Fibre f_3	20.0	450	0.02

To illustrate this result we may consider the particular example of the behaviour of the composite with an aluminium alloy matrix and high strength steel wires as the reinforcement. The characteristics of the components are given in Table I. These characteristics have been combined with two sets of critical stress intensity coefficients, namely $K^*_f = 300\,\text{kgf mm}^{-3/2}$, $K^*_m = 60\,\text{kgf mm}^{-3/2}$, and $K^*_f = 420\,\text{kgf mm}^{-3/2}$, $K^*_m = 60\,\text{kgf mm}^{-3/2}$. These values are typical for tough, high strength steels and aluminium alloys of the Durale type [9].

The dependencies of the critical strain e^* on volume fraction of fibres according to Equation 4 for the composites under consideration are given in Fig. 1. The curves $K^*(v_f)$ are plotted in Fig. 2 (bold lines). It can be seen that the behaviour of the composites is determined by the value of β. At small β there is a maximum on a curve $K^*(v_f)$. It can also be seen that the value of K^* for the composite may be significantly higher than those for both the matrix and the fibre. Obviously still more information can be obtained with regard to the critical stress intensity coefficient divided by the material density.

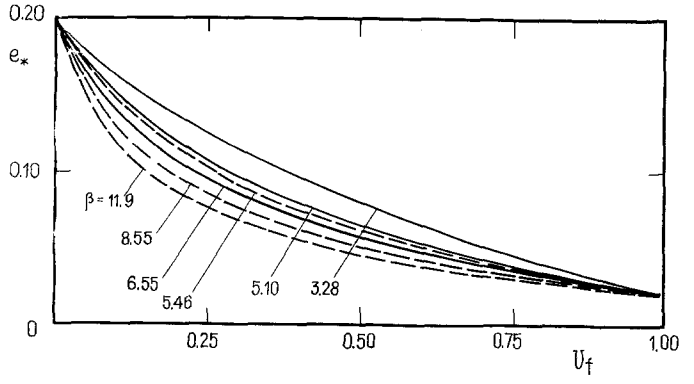

Figure 1 Strain at rupture of composites versus the fibre volume fraction. Dotted and bold lines correspond to composites with matrix m_1 and m_2, respectively (Table I).

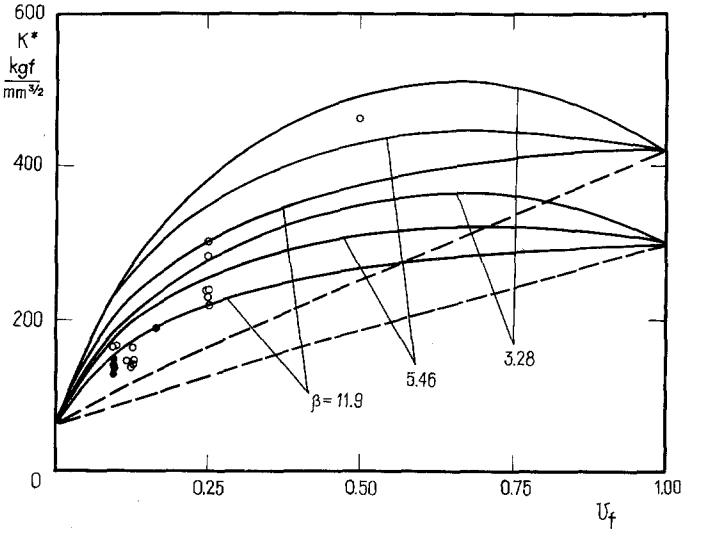

Figure 2 Critical stress intensity coefficients for composites as a function of the fibre volume fraction. The lines are plotted after calculation in accordance with Equation 8, open and solid points correspond to composite specimens of the types of Fig. 3b and c, respectively. The specimen thickness was 1.5 to 3.0 mm.

3. Experimental

In our experiments, Al–Fe composites were prepared and tested. The mechanical properties of the individual components were as follows: $\sigma_m^* = 47\,\text{kgf mm}^{-2}$, $e_m^* = 0.2$, $\sigma_f^* = 320\,\text{kgf mm}^{-2}$, $e_f^* = 0.008$ to 0.015. The diameter of the wire in various composite plates was 0.2 to 0.3 mm. The volume fraction of fibres was 0.1 to 0.5. The composite plates were prepared by a diffusion bonding process at 485° C and pressure of $2.5\,\text{kgf mm}^{-2}$ over 1.5 h.

The specimens used in the tests are shown in Fig. 3. The original cracks were produced by a spark machine and the radius of the tip of the notch was about 0.15 mm.

There were no technical difficulties in obtaining load–COD curves testing the specimens shown in Fig. 3b and c. The original experimental information on testing of the specimens shown in Fig. 3a was obtained in the same manner as that obtained by Gerberich [5, 6]. The position of a crack tip was followed by monitoring fibre breaks

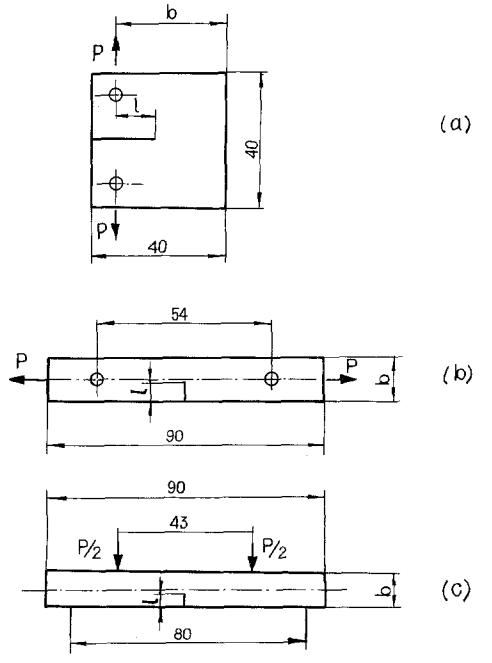

Figure 3 Specimens for fracture toughness measurement.

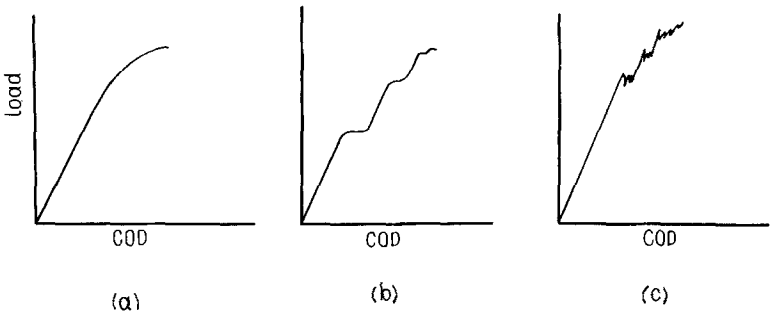

Figure 4 Typical curves obtained in fracture toughness tests.

occurring on the crack path. A signal from a piezoelectric accelerometer related to those breaks was amplified and then recorded on an oscillograph where at the same time, a signal from a dynamometer was recorded.

In Fig. 4 three typical shapes of the load–COD curve are shown. Curves a and b are obtained when a crack cuts the reinforcing fibres. The steps on the curve may be a result of local debonding or breaks of individual fibres without advancing the crack tip. Having no clear interpretation of such steps we shall take a maximum load as a critical value of the load corresponding to the critical stress intensity factor. A curve of the type shown in Fig. 4c is obtained with specimens which show macrodelamination.

All the results of the tests are given in Table II.

Some difficulties arise both in the fracture toughness tests of composites and the interpretation of their results. First, it is not clear what the size of a specimen must be. The usual criterion, based on the estimation of the size of the plastic zone $(K/\sigma_Y)^2$, may not be appropriate in this case, because both the shape and the size of the plastic zone depend on the volume fraction of fibres in the composite. Also the extension of the plastic zone in the matrix differs from that in the fibre [7]. Experimental data obtained on boron–aluminium specimens [10, 11] suggest the use of thin specimens. Further experimental data and the stress analysis of the crack tip in anisotropic non-homogeneous material are necessary in order to establish a minimum thickness of a specimen. This situation is illustrated in Fig. 5 by some results obtained by testing Al–Fe composites.

It is not clear whether it is necessary to sharpen the notch tip by fatigue pre-loading. As one expects high values of K for a metal matrix–metal fibre

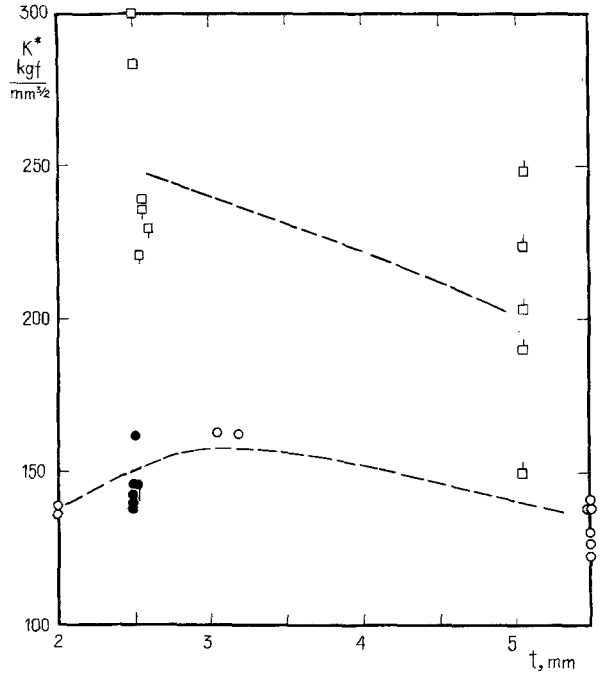

Figure 5 The dependence of critical stress intensity coefficient on the thickness of composite specimens with fibre volume fractions $v_f = 0.10$ (○), 0.13 (●), 0.25 (□). Specimens with fatigue pre-loading, ⚲, ⚲; ⚲ to specimens delaminated at testing. The specimens with $t \geqslant 5$ mm were loaded in bending (Fig. 3c), the specimens with $t < 5$ mm were loaded in tension (Fig. 3b).

TABLE II

Specimen number	v_f	Specimen type (Fig. 3)	t (mm)	l/b	K^* (kgf mm$^{-3/2}$)
515	0.1	a	3.01	0.34	227
516	0.1	a	3.01	0.37	263
519*	0.1	a	3.30	0.38	242
520*	0.1	a	3.33	0.34	260
523†	0.1	a	3.10	0.37	260
524†	0.1	a	3.01	0.38	274
400‡	0.1	a	3.29	0.32	210
401‡	0.1	a	3.30	0.28	214
948	0.1	b	3.05	0.40	163
953	0.1	b	3.10	0.40	163
466	0.1	c	5.60	0.49	139
470	0.1	c	5.42	0.50	142
471	0.1	c	5.44	0.53	123
472	0.1	c	5.60	0.49	127
473	0.1	c	5.65	0.50	139
474	0.1	c	5.64	0.50	131
866	0.1	c	1.95	0.55	139
871	0.1	b	1.93	0.54	136
032	0.13	b	2.50	0.50	138
033	0.13	b	2.50	0.40	163
034	0.13	b	2.50	0.52	140
035	0.13	b	2.70	0.43	143
036	0.13	b	2.70	0.43	146
037	0.13	b	2.60	0.54	146
872	0.17	c	2.43	0.48	187
038	0.25	b	2.60	0.50	230
039	0.25	b	2.55	0.50	240
040	0.25	b	2.55	0.50	221
041	0.25	b	2.50	0.52	283
042	0.25	b	2.50	0.52	302
043	0.25	b	2.65	0.52	239
478	0.25	c	5.04	0.45	> 192
479	0.25	c	5.08	0.52	> 250
480	0.25	c	5.04	0.50	> 225
481	0.25	c	5.05	0.50	> 205
482	0.25	c	5.00	0.44	> 150
860	0.35	b	1.93	0.49	> 289
861	0.35	b	1.95	0.46	> 246
865	0.35	c	1.94	0.48	> 190
955	0.40	b	2.90	0.50	> 304
965	0.40	b	2.90	0.50	> 309
930§	0.50	b	1.39	0.48	> 460

* The specimens were annealed in vacuum at 480° C for 5 h before the standard heat-treatment of the matrix.
† The specimens were annealed at 480° C for 20 h.
‡ The wire was annealed in air at 480° C for 5 h before embedding it in the composite.
§ The specimen contains transversel reinforcement (wire diameter, 0.06 mm, $v_f = 0.05$).

composite one may expect fatigue pre-loading in this case to be of little importance. The data given in Fig. 5 support this suggestion.

At the present time there has been relatively little work carried out on fracture toughness tests of the composites and, as a consequence, the correct choice of a specimen configuration presents difficulties. Since Zobnin and Lomakin

[12] have shown by direct calculation that the relation between load and crack length for a strongly anisotropic material is almost the same as that for an isotropic specimen of the same configuration, the difference between the data obtained on the different specimens remains to be explained. In particular, it is not clear why the values of K obtained on the specimens shown in Fig. 3a are higher than those obtained on the specimens shown in Fig. 3b and c.

4. Discussion

4.1. Dependence of fracture toughness of the composites on fibre volume fraction

Bearing in mind the comments made in the previous section let us plot on the net of curves $K^*(v_f)$ in Fig. 2 the experimental points obtained on specimens of the types (b) and (c) (Fig. 3) with a thickness 2.5 to 3.0 mm and which have not shown delamination. If the experimental data are to be related to the calculated curve taking $\beta =$ 4.3 (which corresponds to the elastic–plastic characteristics of the components of the composites) then K_f^* should be assumed to be about 300 kgf mm$^{-3/2}$. Such a comparison cannot, of course, be a direct confirmation of the model, but it suggests the interesting possibility of determining the fracture toughness of a wire in order to use it in the design of a composite. At the same time this comparison can be considered to be indirect evidence that the model allows a good general description of the behaviour of a metal matrix–metal fibre composite.

4.2. Delamination

In evaluating Equation 3 the interfacial bond was assumed to be not less than the matrix strength, and mode I of the original crack was expected to remain. The validity of the second assumption with respect to the microlevel should be discussed, but on the macrolevel it is not valid at high enough values of v_f. It is well known [5–7] that after some value of v_f is reached, macrodelamination occurs because the direction of crack extension changes so that the crack advances along the fibres. In fact, it is the result of an increase in fracture toughness in the direction normal to the fibres with increase in v_f. The value of v_f when the delamination begins is an optimum in the sense that a further increase in fibre volume fraction does not permit an increase in strength of the reinforcement if a composite structure contains a stress concentrator. It should be noted that all specimens tested delaminate when the load reaches a value corresponding to $K = 200$ to 300 kgf mm$^{-3/2}$.

It is clear that some increase in the structural strength of a composite can be achieved by strengthening the fibre–matrix interface (until the interfacial strength becomes equal to the matrix strength). Further increase in the fracture toughness of a composite can be obtained by an increase in the matrix shear strength and the matrix transverse tensile strength. An effective way of doing this is an additional reinforcement of the matrix in the transverse direction. Indeed the specimen no. 930 (Table II) with a small volume fraction of the transverse reinforcement, delaminated at $K = 460$ kgf mm$^{-3/2}$ only, (this specimen was a little thinner than those used in Fig. 2).

4.3. Influence of the fibre–matrix interface on fracture toughness of a composite

It is known that the fracture toughness of composites with two brittle components is determined by the interface. The role played by the interface between a ductile matrix and a ductile fibre is less clear. A possible macrodelamination was considered above. However, weakening of the interface at low fibre volume fraction can cause a localized delamination and a premature necking of fibres. On the other hand, a weak interface can increase the energy dissipation due to either pull-out or growth of an area of the plastically deformed zone in the matrix. The possible effect of the intefacial strength on the fracture toughness of the composite was examined using the specimens shown in Fig. 3a.

Four composite plates were prepared for this experiment (the corresponding specimens are given in the first eight lines of Table II). The first two specimens were prepared as normal. Four specimens were subjected to diffusion bonding and were then heated in vacuum at a temperature of 480° C for some time (see Table) to form a thicker intermetallic layer at the interface (tensile tests of unnotched specimens cut out of these plates have shown that the fibre properties remained unchanged). Two specimens contained wires having an oxidized surface layer.

The results of tests on these specimens show that a change in the interfacial conditions by the formation of a thin intermetallic layer caused no essential change in the fracture toughness of the

composite. The weakening of the interface in the specimens with fibres covered by an oxidized layer resulted in some decrease in the fracture toughness, although the degree of interfacial weakening is difficult to estimate.

The contribution of energy dissipation on the interface to the work of fracture of a composite may be estimated quantitively by postulating an ideal composite with no bonds along the interface as two parallel bars rupturing independently. The work of fracture of such a system will, therefore, be

$$G = \frac{K_f^{*2}}{E_f} v_f + \frac{K_m^{*2}}{E_m} v_m. \quad (9)$$

The ratio of the critical stress intensity factors of a composite with ideal bonding and that with no bonds and no energy dissipation on the interface will be

$$\frac{K^*}{\bar{K}^*} = \left[\frac{e^*}{e_f^*} \frac{1 + \frac{v_m}{v_f} \left(\frac{K_m^*}{K_f^*} \right)^2 \frac{E_f}{E_m} \frac{e_f^*}{e_m^*}}{1 + \frac{v_m}{v_f} \left(\frac{K_m^*}{K_f} \right)^2 \frac{E_f}{E_m}} \right]^{1/2}. \quad (10)$$

In the case under consideration ($\beta = 4.3$) $K^*/\bar{K}^* = 2.1$. In the experiment, the ratio of values of K^* for composites with high and low values of interfacial strength is about 1.2. This suggests that either the strength of the interface in the specimens with non-ideal bonding is sufficiently high to delay fibre necking, or that the energy dissipation on the weak interface is of the order of the increase in the work of fracture of the wire in the composite, due to the delay of fibre necking. It is possible that a combination of the two reasons may be the correct explanation.

5. Conclusions

(1) A model of a metal matrix–metal fibre composite with a crack has been constructed. This model is based on the increase in the stability of a fibre in a ductile matrix. Experiments carried out on Al–Fe composites supports the model concept.

(2) The possibility of an experiment giving the critical stress intensity factor of a high strength wire has been suggested. The basic test is that of a composite specimen with a notch.

(3) It has been shown that the increase in the transverse tensile and longitudinal shear strength by an additional reinforcement in the transverse direction can lead to the realization of a high resistance to crack extension in a metal–metal composite in the direction normal to the fibres.

(4) It has been shown experimentally that a relatively large variation of interfacial strength in the metal matrix–metal fibre composites may not cause an essential change in the fracture toughness of the composite.

Acknowledgement

The authors wish to thank Dr J. Thynne for his valuable advice on the style of presentation of the paper and also Mr E. G. Golofast and Mr A. A. Bastanov for their help in the experimental work.

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Received 12 May and accepted 4 July 1975.