JOURNAL OF MATERIALS SCIENCE 22 (1987) 3005–3010

The effect of matrix properties on reinforcement in short alumina fibre–aluminium metal matrix composites

C. M. FRIEND
Materials Technology Group, Royal Military College of Science, Shrivenham, Wiltshire SN6 8LA UK

The effect of the alloy matrix on room-temperature strengthening in δ-alumina-reinforced aluminium alloys has been investigated. Alloy matrices fell into two families exhibiting significantly different fibre-strengthening response. The first gave rise to little or no improvement in the room-temperature strength, while the second gave significant improvements by up to 300%. It is shown that a simple Rule of Mixtures (ROM) strength analysis, modified to account for the discontinuous and random orientation of the reinforcement, can adequately explain these responses. Little or no reinforcement occurs when the matrix properties result in a high value for the critical volume fraction V_{CRIT} which must be exceeded to produce any increase in strength. However, by careful selection of the matrix alloy V_{CRIT} can be reduced, thus giving significant reinforcement of the room-temperature strength. This analysis shows that for optimum room-temperature reinforcement the matrix alloys should exhibit a low rate of work-hardening. In certain alloys reinforcement levels were in excess of those predicted by the ROM analysis. It is proposed that this occurs in relatively low-strength matrices as a result of dispersion strengthening of the matrix due to the presence of the fibre array.

1. Introduction

In recent years considerable interest has been shown in the reinforcement of aluminium alloys by short δ-alumina fibres [1]. Work has shown that these composites exhibit good interfacial bonding and improvements in the room-temperature modulus and high-temperature modulus and strength [2]. However, in the volume fractions investigated experimentally, these systems exhibit a number of anomalies in their room-temperature strengths. A feature of many of the alloy–fibre combinations investigated is little or no improvement in the room-temperature strength despite obvious improvements in modulus. This behaviour has been observed in a number of alloys reinforced by δ-alumina fibres, in particular Al–Si alloy [1, 2]. Equally as strange is that in other alloys containing similar volume fractions of reinforcement considerable improvements in room-temperature strength are observed, some as great as 200% compared with the properties of the unreinforced alloy [3]. Dinwoodie *et al.* [2] have discussed ways in which the room-temperature strength can be empirically optimized; however, little work has been conducted to ascertain the fundamental reasons for this anomalous room-temperature strength behaviour. This paper presents the results of such a study with an analysis of the room-temperature strength of these composites.

2. Experimental details

Composite materials were prepared by a pressure infiltration technique [1] using a number of aluminium alloy matrices and oriented preforms of short δ-alumina fibres (Saffil, from ICI (Mond Div.)). Table I contains the compositions of these alloys and the processing variables employed during the casting of the materials, and Table II shows typical properties of the Saffil alumina fibre. The densities of the preforms were in the range 0.66 to 0.71 g cm^{-3}, which produced composites with a fibre volume fraction (V_f) of ~ 0.25. To allow comparison of the properties of the unreinforced alloys and composites, the volume of metal poured into the die during casting was in excess of that required for infiltration of the preform. This made available both composite and unreinforced alloy processed under identical thermal/pressure conditions. Following each cast the unreinforced alloy was machined from the composite, producing two discs 100 mm in diameter and 15 mm thick. Round tensile specimens of 6 mm diameter and 30 mm gauge length were machined from these discs with their axes parallel to the plane of the disc, and these specimens were then tested using a conventional screw-driven tensile machine. Optical and scanning electron microscopy was employed to characterize both the alumina fibre preforms and the final composites.

3. Results

Fig. 1 shows a scanning electron micrograph of a typical uninfiltrated δ-alumina fibre preform which confirms the interlocked pseudo-three-dimensional random nature of the preform, and therefore the resulting pseudo-random fibre arrangement in the cast

0022–2461/87 $03.00 + .12

TABLE I Alloy compositions and casting parameters

Alloy	Die preheat (°C)	Preform preheat (°C)	Casting temperature (°C)	Casting pressure (MPa)	Duration of applied pressure (sec)
Al–4.0 Zn–2.0 Mg	520	520	1000	25	10
Al–12 Si	515	515	950	38	10
Commercially pure aluminium	520	520	1000	25	10

composite. Fig. 2 shows typical microstructures of the cast composites of the Al–4.0 Zn–2.0 Mg and Al–12 Si (wt %) alloys. All the composites were of good quality with little evidence of porosity or the ingress of dross or inclusions. The grain sizes in the matrices of the composites were generally smaller than those of the unreinforced alloys cast under the same conditions, and in the case of the Al–Si matrix the eutectic was generally finer and showed some signs of a divorced eutectic with preferential nucleation of the silicon plates around the fibres.

Table III contains the average tensile data for both unreinforced and reinforced alloys. The composites showed two distinct types of behaviour. In the case of the Al–Zn–Mg and Al–Si alloys the 0.25 V_f of Saffil resulted in little or no reinforcement. In the Al–Zn–Mg matrix there was a 3% reduction in strength compared with the unreinforced alloy, and in the Al–Si matrix a 15% improvement in room-temperature strength. The second type of behaviour was shown by the commercially pure aluminium matrix, which exhibited an increase in strength by over 200% on the addition of 0.25 V_f Saffil fibre. All the composites exhibited low ductilities compared with the unreinforced matrix alloys, typically less than 1%.

4. Discussion

It is clear from the results presented above that short-fibre strengthening of aluminium alloys is highly dependent on the alloy matrix employed. These results are consistent with previous observations. The lack of reinforcement in Al–Si alloys has been observed previously [1–3] with reductions in strength of between 25 and 30% observed in some alloys on the introduction of Saffil fibres [3]. The improvement in room-temperature strength has also been previously observed in fibre-reinforced commercially pure aluminium [3]. The question which remains to be answered, however, is what is the reason for these significantly different strengthening responses in aluminium alloy composites?

The strength of composites can be analysed by a simple Rule of Mixtures (ROM) approach. However, in the case of discontinuous randomly oriented fibre arrays some modification is required to account for the lack of reinforcing efficiency due to (a) the discontinuous nature of the fibres and (b) the three-dimensionally random orientation of the reinforcement. A simple ROM analysis for uniaxial continuous-fibre composites above a critical volume fraction of fibres produces the following equation for the composite strength:

$$\sigma_c = \sigma_{uf} V_f + \sigma_m^*(1 - V_f) \quad (1)$$

where σ_c is the composite strength, σ_{uf} is the strength of the reinforcing fibres, V_f is the volume fraction of fibres and σ_m^* is the matrix stress at the fibre failure strain.

This equation must be modified for discontinuous reinforcement to account for the finite length of fibre required to transfer stress from the matrix to the fibre. In this case an extra term relating the fibre length (l) to the critical transfer length (l_c) must be introduced. Thus

$$\sigma_c = \sigma_{uf} V_f \left(l - \frac{l_c}{2l}\right) + \sigma_m^*(1 - V_f) \quad (2)$$

To account for a non-axial distribution of fibres an additional term must be introduced to further reduce the fibre reinforcing efficiency. Thus

$$\sigma_c = C\sigma_{uf} V_f \left(l - \frac{l_c}{2l}\right) + \sigma_m^*(1 - V_f) \quad (3)$$

where C is an empirical constant describing the inefficiency of reinforcement due to a distribution of fibre orientations.

A number of values have been suggested for C, two of the most widely employed forms of the strength equation being

$$\sigma_c = \tfrac{3}{8}\,\sigma_{uf} V_f \left(1 - \frac{l_c}{2l}\right) + \sigma_m^*(1 - V_f) \quad (4)$$

for thin laminae i.e. a planar random orientation, and

$$\sigma_c = \tfrac{1}{5}\,\sigma_{uf} V_f \left(1 - \frac{l_c}{2l}\right) + \sigma_m^*(1 - V_f) \quad (5)$$

for a completely random fibre array.

These ROM analyses are based on the assumption that the fibres do not fracture until a unique stress level (σ_{uf}) is reached, where they all fracture together. This is clearly an over simplification for these randomly oriented alumina fibres, which have a distribution of fibre strengths and which exhibit a distribution of fibre stresses due to their random three-dimensional

TABLE II Typical properties of Saffil fibre [4]

Composition	96 to 97% Al_2O_3	Crystal structure	δ-alumina
Mean diameter	3 μm	Elastic modulus	300 GPa
Strength	2000 MPa	Failure strain	0.67%
	Fibre length 500 μm		

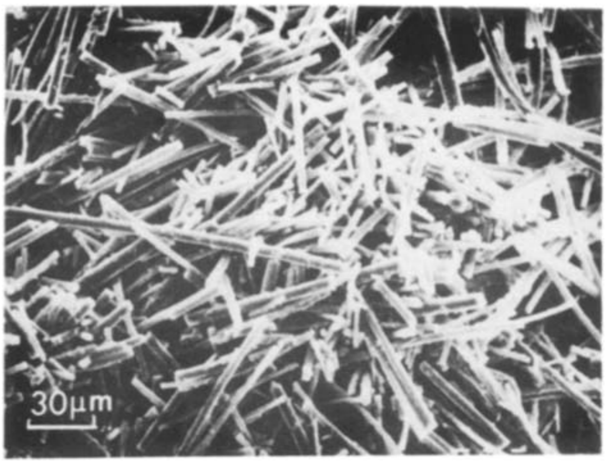

Figure 1 Scanning electron micrograph of an uninfiltrated Saffil preform.

orientation. These equations do, however, allow some analysis of the anomalous room-temperature strength behaviour in short alumina fibre-reinforced aluminium alloys.

Fig. 1 shows the typical arrangement of fibres in a Saffil preform with a pseudo-three-dimensionally random orientation of fibres. If this is the case, Equation 5 should describe the strength of these composites above a critical volume fraction of reinforcement. The ROM strength diagrams for these composite systems can therefore be calculated using Equation 5 and a consideration of the decrease in residual matrix strength with increasing volume fraction of reinforcement:

$$\sigma = \sigma_{um}(1 - V_f) \quad (6)$$

where σ_{um} is the strength of the unreinforced matrix alloy.

The data required for this calculation are, σ_{um}, σ_m^*, σ_{uf}, l and l_c. σ_{um} and σ_m^* have been evaluated for the experimental alloys from tensile tests on the unreinforced matrix alloys and are contained in Table III. The grain sizes of the composite matrices were finer than those of the unreinforced alloys cast under the same conditions as a result of the solidification taking place within the constraint of the fibre array. However, the values of σ_{um} and σ_m^* measured on

TABLE III Average tensile data for unreinforced and reinforced alloys

Alloy	UTS (MPa)	σ_m^*(MPa)	No. of casts
Unreinforced Al–Zn–Mg	273	226	6
0.25 V_f Al–Zn–Mg	266	–	6
Unreinforced Al–Si	143	88	4
0.25 V_f Al–Si	165	–	4
Unreinforced aluminium	55	50	2
0.25 V_f aluminium	178	–	2

unreinforced alloys should give reasonable measures of these properties in the composite matrices. Values for σ_{uf} and l were obtained from the published literature on Saffil fibres [4] and are shown in Table II. Values for l_c were determined using the equation

$$l_c = \frac{\sigma_{uf} d}{2\tau} \quad (7)$$

where d is the fibre diameter and τ is the shear strength of the fibre–matrix interface.

A value for the fibre diameter was obtained from published literature on Saffil (Table II); however, the interfacial shear strengths are more difficult to establish. For the purpose of this calculation the shear strengths of the interfaces were assumed to be τ_{ym}, the shear yield strengths of the matrix alloys, which were also assumed to be half the tensile yield strengths. Thus Equation 5 can be rewritten as

$$\sigma_c = \tfrac{1}{5}\sigma_{uf} V_f \left(1 - \frac{\sigma_{uf} d}{2l\sigma_{ym}}\right) + \sigma_m^*(1 - V_f) \quad (8)$$

where σ_{ym} is the tensile yield strength of the unreinforced matrix alloy. The ROM strength diagrams for the experimentally investigated alloys can therefore be calculated using Equations 6 and 8 and the data in Tables II and III.

Figs 3 and 4 show the results of such calculations for the Al–Zn–Mg and Al–Si composites cast in the present work. Both systems show good agreement between the modified ROM calculations and the experimentally determined composite strengths, suggesting that Equations 6 and 8 can be used to describe the strength against V_f behaviour in these two

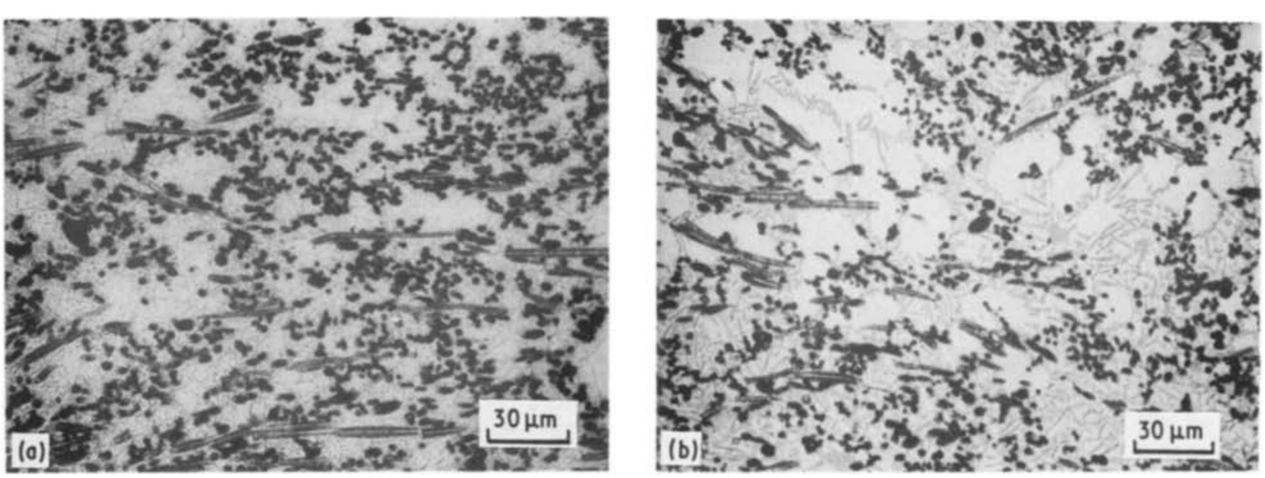

Figure 2 Optical micrographs of composites: (a) 0.25 V_f Saffil/Al–4.0 Zn–2.0 Mg, (b) 0.25 V_f Saffil/Al–12 Si.

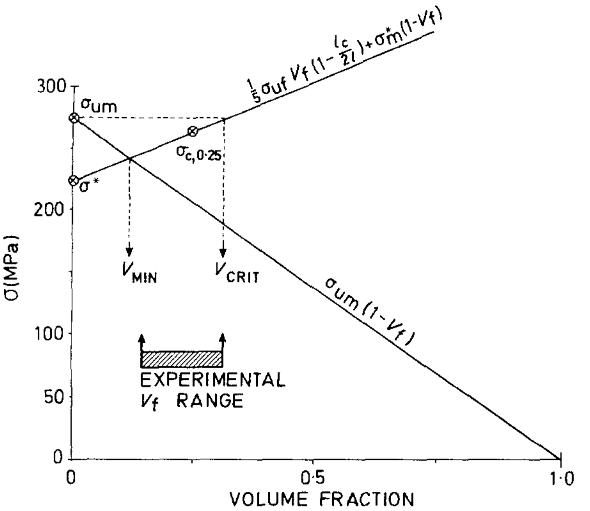

Figure 3 ROM prediction for Saffil-reinforced Al–4.0 Zn–2.0 Mg composites: (⊗) experimental data.

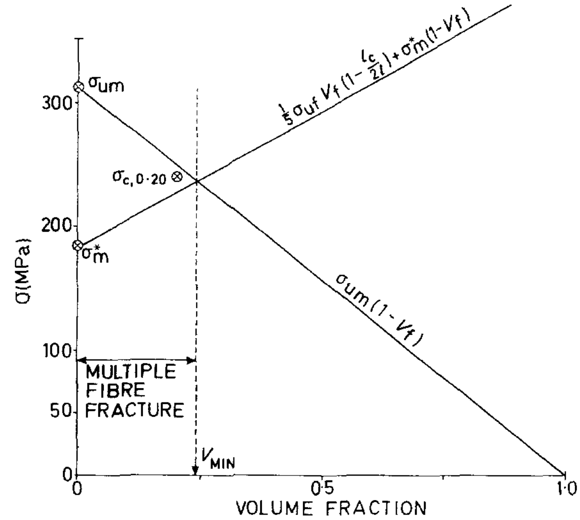

Figure 5 ROM prediction for Saffil-reinforced Al–7 Si [3].

composite systems. Superimposed on Figs 3 and 4 are the typical volume fraction ranges investigated experimentally using the preform and pressure infiltration technique ($V_f \sim 0.15$ to 0.30). If this V_f range is compared with the strengthening behaviour it is clear why these systems exhibit little or only limited reinforcement of the matrix strength.

The ROM diagram shows two important volume fraction parameters: (i) V_{MIN}, the critical volume fraction at which the failure mode of the composite changes from a multiple fibre-fracture mode to instantaneous failure of the composite following fibre fracture; and (ii) V_{CRIT}, the volume fraction which must be exceeded to produce reinforcement of strength above that of the unreinforced alloy. It is clear from the calculated ROM diagrams that the commonly investigated V_f ranges lie in the region of V_{MIN} and V_{CRIT}. Considering the Al–Zn–Mg composite system (Fig. 3) it is clear that the usual volume fractions investigated lie below V_{CRIT}.

The composites produced using this alloy would therefore be expected to show no reinforcement and would exhibit strengths below that of the unreinforced matrix alloy. It would require volume fractions in excess of 0.30 before reinforcement would be expected.

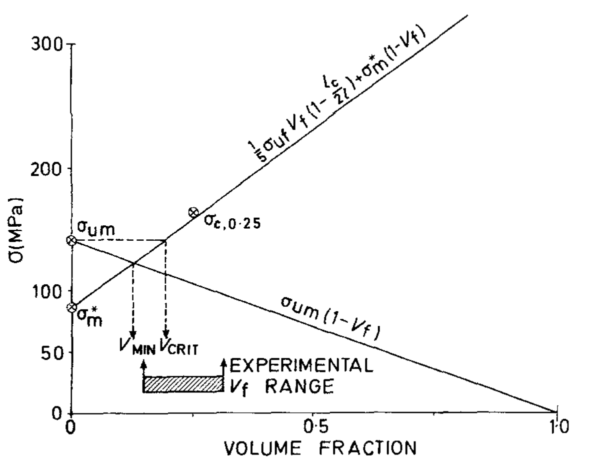

Figure 4 ROM prediction for Saffil-reinforced Al–12 Si composites: (⊗) experimental data.

Considering the Al–12Si composite system (Fig. 4) it is clear why only limited reinforcement is observed. The commonly investigated V_f range for this alloy lies only partially above V_{CRIT}, and therefore only higher V_f composites would exhibit strengthening. The reinforcement in such composites would, however, be small due to the low volume fraction dependence of the composite strength (as a result of the low efficiency of reinforcement in pseudo-three-dimensionally random fibre arrays).

Using these diagrams also gives a qualitative understanding of the low failure strains in these composites. Composites with V_f greater than V_{min} exhibit instantaneous failure following fibre fracture. Since fibre fracture occurs typically at strains below 1%, the composite materials with $V_f > V_{min}$ would therefore be expected to exhibit failure strains no greater than this level. This must therefore be the qualitative reason for the low composite failure strains.

These diagrams also suggest that if the V_f of a composite is less than V_{min} multiple fracture of the reinforcing fibres should be observed prior to final failure. Harris and Wilks [3] have observed multiple fracture of the reinforcement in a 0.20 V_f Saffil/Al–Si composite. Table IV contains the tensile properties of their matrix alloy and composite, and Fig. 5 shows the ROM prediction for their system. From Fig. 5 it is clear that they observed multiple fracture because their experimental volume fraction of reinforcement lies below V_{min}.

Since this simple ROM analysis appears to adequately model the room-temperature strength behaviour of these composites, it can be used to predict the properties of the component phases which are required to produce a composite exhibiting extensive room-temperature reinforcement. The critical parameter to assess the level of reinforcement at a given volume fraction of fibre is V_{CRIT}. If V_{CRIT} is small, the experimentally accessible volume fractions will have

TABLE IV Tensile data for unreinforced and 0.20 V_f Saffil reinforced Al–7 Si alloy [3]

$\sigma_{um} = 312$ MPa	$\sigma_m^* = 184$ MPa	$\sigma_c = 237$ MPa

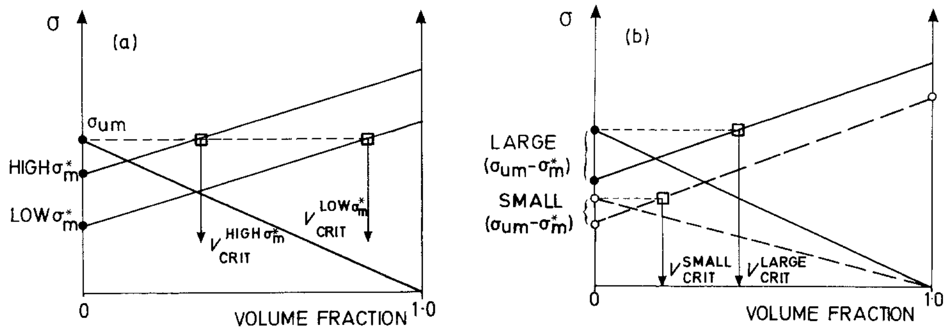

Figure 6 (a) Effect of σ_m^* on V_{CRIT} (σ_{um} constant). (b) Effect of the stress difference ($\sigma_{um} - \sigma_m^*$) on V_{CRIT}.

$V_f > V_{CRIT}$ and therefore exhibit reinforcement of the room-temperature strength. At V_{CRIT}, $\sigma_c = \sigma_{um}$, therefore rearranging Equation 8 gives

$$V_{CRIT} = \frac{(\sigma_{um} - \sigma_m^*)}{\frac{1}{5}\,\sigma_{uf}[1 - (\sigma_{uf}d/2l\sigma_m)] - \sigma_m^*} \quad (9)$$

This indicates that V_{CRIT} is highly dependent on the properties of the matrix alloy. It is clear from Equation 9 that V_{CRIT} depends primarily on the value of σ_m^* and the stress difference ($\sigma_{um} - \sigma_m^*$). If σ_m^* is increased (for a constant σ_{um}) or if both σ_{um} and σ_m^* change such that the difference ($\sigma_{um} - \sigma_m^*$) decreases, then V_{CRIT} is reduced to lower reinforcement volume fractions. This is illustrated schematically in the ROM diagrams in Fig. 6. This suggests that the matrix alloys which exhibit reinforcement at room temperature will be those with a small value for the stress difference ($\sigma_{um} - \sigma_m^*$); these are typically alloys which exhibit a small difference between their yield stress and ultimate tensile stress (UTS), i.e. alloys which have a low rate of work-hardening. Fig. 7 schematically illustrates the stress–strain curves of two matrix alloys, one which would be expected to show little strengthening and the other which would exhibit significant reinforcement of the room-temperature strength.

A suitable matrix material to test this hypothesis is commercially pure aluminium. This exhibits extensive plastic deformation with a low rate of work-hardening, a low yield strength and a low UTS. The combination of these properties produces a low value for ($\sigma_{um} - \sigma_m^*$) (5 MPa in the case of the aluminium cast in this work). Fig. 8 shows an ROM prediction for this matrix. This analysis predicts a V_{CRIT} of ~ 0.025, and composites within the volume fraction range normally investigated would be expected to exhibit reinforcements of between 168 and 250%. This calculated behaviour is qualitatively consistent with both measurements in the present work, which indicated a reinforcement of 324% for a 0.25 V_f material, and measurements by Harris and Wilks [3] which indicate a reinforcement of 359% in a 0.20 V_f material. However, although qualitatively consistent, the experimental data show significant levels of reinforcement above that predicted by the ROM. The strength of the commercially pure aluminium composite cast in the present work was 55 MPa above the level predicted by the simple ROM analysis. This behaviour is similar to that obtained on analysing the data of Harris and Wilks [3] for a 0.20 V_f Saffil–Commercially pure aluminium composite. Fig. 9 shows an ROM analysis based on their data. In this case the composite strength is 65 MPa above that predicted by the ROM. Since the ROM analysis does not appear to fully describe the room-temperature strength behaviour in commercially pure aluminium matrices, it is important to analyse the terms in Equations 5 and 8 to identify any factors which may give rise to this improved reinforcement.

If one analyses the physical significance of the empirical constant C, it is clear that this parameter simply describes the reduction in reinforcing efficiency of a given volume fraction of fibre due to its three-dimensionally random orientation. A value of $C = 0.2$ has been shown above to be appropriate to describe this effect in an as-cast Al–Zn–Mg alloy and in Al–Si alloys. It would therefore be inappropriate to modify this constant for the commercially pure aluminium matrix. Instead the equations must be inspected further to identify other factors which may contribute

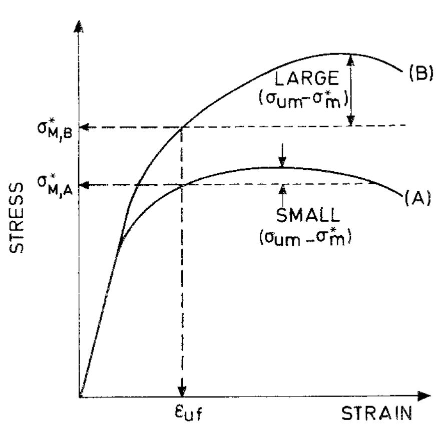

Figure 7 Schematic stress–strain curves for composite matrices which exhibit (A) significant reinforcement, (B) limited reinforcement.

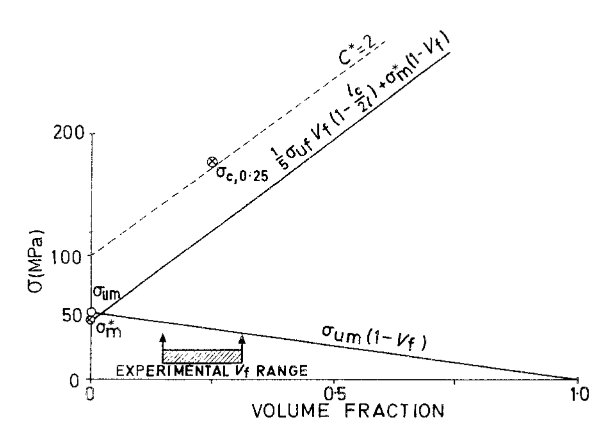

Figure 8 ROM prediction for Saffil-reinforced commercially pure aluminium: (⊗) experimental data.

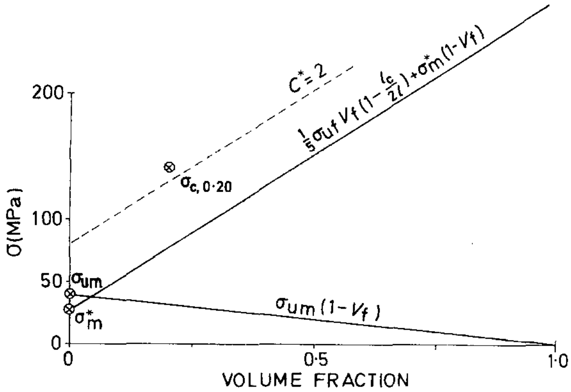

Figure 9 ROM prediction for Saffil-reinforced commercially pure aluminium [3].

to this reinforcement effect. If this is done there are clearly two terms which can be modified to account for the improved reinforcement. These are (i) the interfacial shear strength which alters the term $(1 - l_c/2l)$, and (ii) σ_m^*. An estimate has been made of the interfacial strength in terms of the yield strength of the unreinforced matrix alloy which appears to be a good approximation for Al–Zn–Mg and Al–Si alloys. There is, however, no evidence that this approximation applies equally to commercially pure aluminium matrices. The ultimate value of the term $(1 - l_c/2l)$ is unity, where the composite behaviour approaches that of a continuously reinforced material due to a low value of l_c. If this maximum value for $(1 - l_c/2l)$ is employed it produces a higher strength composite, but cannot by itself result in the observed composite strengths. However, if the matrix property σ_m^* is modified, good agreement can be obtained. The ROM strength equation then becomes

$$\sigma_c = \tfrac{1}{5}\sigma_{uf}V_f\left(1 - \frac{l_c}{2l}\right) + C^*\sigma_m^*(1 - V_f) \quad (10)$$

where C^* is an empirial constant.

This behaviour is plotted in Figs 8 and 9 for the commercially pure aluminium composites, where it is clear that good agreement is obtained between the experimental and predicted data if a value of $C^* = 2$ is employed. The values of C^* for the investigated alloys therefore lie between 1 for the Al–Zn–Mg and Al–Si alloys, and 2 for commercially pure aluminium. The physical significance of the constant C^* is not entirely clear. It results in an effective σ_m^* (and also therefore an effective σ_{um}) which is much larger than that measured in the unreinforced matrix alloy. At present one can only speculate why this higher value arises only in commercially pure aluminium matrices. There is no evidence of fibre dissolution resulting in either solution or precipitation hardening of the matrix; however, it is possible that in relatively weak matrices such as commercially pure aluminium the presence of the fibres may have a dispersion-strengthening effect on the matrix in addition to the fibre reinforcement, which raises σ_m^* of the matrix. This latter hypothesis is consistent with microhardness tests carried out in Al–Mg, Al–Si and Al–Cu matrices containing fibre arrays [1].

5. Conclusions

The strength of pseudo-randomly reinforced discontinuous alumina fibre–aluminium alloy metal matrix composites can be described by a simple Rule of Mixtures analysis, modified to account for the discontinuous and random orientation of the reinforcing fibres. Such an analysis shows that in many aluminium alloy composites V_{CRIT} is relatively high, which results in the observed lack of reinforcement of room-temperature strength. Reinforcement is only observed where the matrix alloy exhibits a small stress difference $(\sigma_{um} - \sigma_m^*)$, in which case improvements of 200 to 300% are predicted over the strength of the unreinforced matrix alloy. In such systems the reinforcement can also be greater than that predicted by the simple Rule of Mixtures analysis. It is proposed that this occurs as a result of dispersion strengthening of the matrix alloy by the presence of the fibre array.

This relatively simple Rule of Mixtures analysis can therefore be used to describe the complex interplay which takes place between the matrix properties and fibre orientation in these composites, as well as indicating the failure mode operating during their deformation.

Acknowledgements

The author thanks A. Nixon, J. Bradfield and J. Crocker for experimental assistance, Docent R. Warren for useful discussions and J. Dinwoodie (ICI, Mond Division) for the suppy of Saffil preforms.

References

1. T. W. CLYNE, M. G. BADER, G. R. CAPPLEMAN and P. A. HUBERT, *J. Mater. Sci.* **20** (1985) 85.
2. J. DINWOODIE, E. MOORE, C. A. J. LANGSAN and W. R. SYMES, in Proceedings of 5th International Conference on Composite Materials, San Diego, California (Metallurgical Society Inc., Warrendale, Pennsylvania, 1985) p. 671.
3. S. HARRIS and T. E. WILKS, in Proceedings of 1st European Conference on Composite Materials, Bordeaux, France, September 1985, edited by A. R. Bunsell, P. Lamicq and A. Massiah (European Association for Composite Materials) p. 95.
4. Data sheet on R. F. Saffil (ICI, Mond Division, 1984).

Received 13 October 1986
and accepted 19 January 1987