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The influence of off-axis reinforcement on the tensile strength of an Ni–Al–Cr–C eutectic composite

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The tensile strength of an Ni–Al–Cr–C eutectic composite which has an Ni-rich matrix containing a dispersion of Ni_3Al precipitates has been measured both with a fully-aligned and a cellular microstructure. The fully-aligned eutectic behaves as a simple metal-matrix composite. The reduction in strength of the cellular eutectic due to the off-axis reinforcement is shown by a simple analysis to result from weakness of the fibre/matrix interface which is borne out by observations of the fracture surface.

1. Introduction

A number of defects may exist in directionally solidified eutectic alloys, for example, banding, grain boundaries, primary crystallization or cellular growth, which have a deleterious effect on the strength of the composite. These defects may induce fracture by virtue of the absence of one of the phases in a growth band or due to the misalignment of the phases with primary crystallization [1] or cellular growth.

This paper describes a study of the tensile behaviour of an Ni–Al–Cr–C pseudo-binary eutectic alloy in which a comparison was made between specimens with fully-aligned and those with cellular microstructures. The alloy has a composite microstructure with Cr_3C_2 fibres in an Ni-rich (γ) matrix with Cr ($\sim 4\%$) and some Al in solid solution containing a dispersion of Ni_3Al (γ') precipitate particles [2]. The orthorhombic Cr_3C_2 fibres invariably grow with their axes parallel to the crystallographic c-axis whilst the fcc matrix growth direction may be either $\langle 110 \rangle$ or $\langle 100 \rangle$. Under the conditions used in this work, fully-aligned growth was only maintained over less than half of the ingot after which it suddenly degenerated to a cellular type of morphology. Tensile specimens were machined from either side of this interface and their fracture behaviour was compared.

2. Experimental

The alloy used in this study was made up from Ni of 99.96% purity, Al of 99.995% purity, Cr_3C_2 of 99.5% purity and Cr of 99.9% purity with a composition (wt. %) of Ni, 79.0%; Al, 6.9%; Cr, 12.3%; and C, 1.8%. This was melted in pure alumina crucibles in a dynamic argon atmosphere and chill cast in air into a steel split-mould to form ingots 9 mm in diameter, and 180 mm long. For directional solidification the ingots were placed in contact with a water-cooled copper chill-block in a soft mould of alumina powder contained in an outer alumina sheath. This assembly was lowered through an RF coil at 351 mm h^{-1} in an argon atmosphere. The thermal gradient across the liquid/solid interface was measured as 13.5 K mm^{-1}. Directionally solidified ingots were produced that were free of such defects as porosity and banding.

A small flat was ground onto the side of each ingot and polished metallographically to enable tensile specimens to be cut with completely aligned or completely cellular microstructures. Tensile specimens were machined by grinding the shoulder profile and then spark machining the cylindrical gauge-length profile by means of a brass tool driven by a servo-motor onto the specimen rotating in a lathe, and in the final stages the spark energy was reduced in small steps to achieve a good surface finish. The specimens had reduced sections 5 mm in length with a diameter of 2.0 mm and were all finished uniformly with 600 grit silicon carbide paper prior to testing. An Instron testing machine was used

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for tensile tests and the strain was measured using a Weidemann-Baldwin microformer-type extensometer.

Fractured specimens were examined directly by scanning electron microscopy and longitudinal microsections with the surface profile protected by Cu- or Ni-plating were studied by optical microscopy.

3. Results and discussion

3.1. Morphology of the Ni–Al–Cr–C eutectic

The microstructure of the Ni–Al–Cr–C alloy consisted of faceted Cr_3C_2 fibres in a γ matrix that contained a high density (approximately 50% by volume) of cubic precipitates of γ' whose size was in the range 0.4 to 1.2 μm (Fig. 1). The volume fraction of Cr_3C_2 was 0.11 and the mean fibre spacing ~ 4 μm. The fibre spacing was not unique in this material, and the figure quoted is the square root of the reciprocal of the mean number of fibres per unit area in transverse sections.

The fully-aligned structure only grew for part of the distance from the chill and then suddenly degenerated to a cellular type of morphology (Fig. 2). This sudden breakdown is not characteristic of the aligned/cellular transition in binary eutectic alloys [3] which occurs more gradually due to the onset of constitutional supercooling although the morphology of the cellular structure is similar. Nevertheless, reducing the growth rate increased the distance from the chill over which there was fully-aligned growth in the normal manner but increasing the melt superheat and thus the thermal gradient did not have a reproducible effect on the position of the interface. The aligned/cellular transition in this alloy is not solely controlled by the onset of constitutional supercooling and may be influenced by morphological factors.

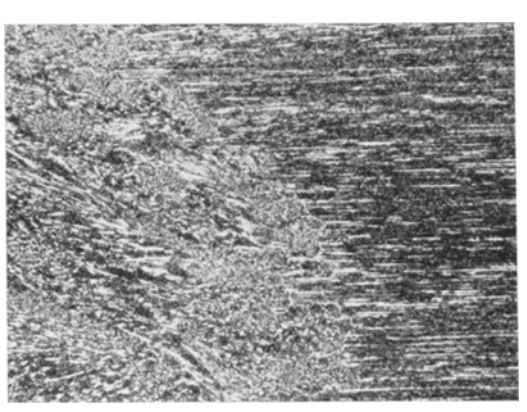

Figure 1 Longitudinal section of the aligned Ni–Al–Cr–C eutectic alloy showing carbide fibres and cubic precipitates of Ni_3Al in two orientations in the Ni-rich matrix, × 950.

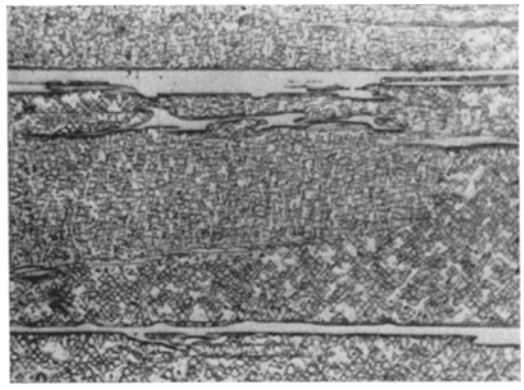

Figure 2 Longitudinal section of the directionally solidified Ni–Al–Cr–C alloy showing the sudden transition from the aligned to the cellular morphology, × 75.

Examination of the orientation of the matrix as revealed by the traces of the {100} planes of the γ/γ' interface [2] showed that closer to the chill [110] matrix orientations tended to predominate whereas in the cellular region [100] matrix orientations were more prevalent. The fibres invariably grow with [001] as the fibre axis and closer to the chill they grow parallel to the heat flow. The matrix is then nucleated by the fibres in a [110] orientation and the two phases have a simple orientation relationship with preferred interface planes [2]. Further from the chill, preferred [100] growth directions occur in the fcc matrix and the fibres grow in a variety of directions, compromising between growth in the heat flow direction and along the [110] matrix direction.

3.2. Tensile behaviour of the aligned alloy

The stress-strain curve of the fully-aligned Ni–Al–Cr–C alloy tested at room temperature is shown in Fig. 3 (curve a). The primary modulus (elastic fibres/elastic matrix) was 260 GN m^{-2}, the composite yield stress was 1150 MN m^{-2}, the secondary modulus (elastic fibres/plastic matrix) was 180 GN m^{-2}, and the UTS 1630 MN m^{-2}.

The fracture took place in the centre of the gauge-length, and the fracture surface was macroscopically transverse to the tensile axis. On a fine scale, the fracture surface consisted of

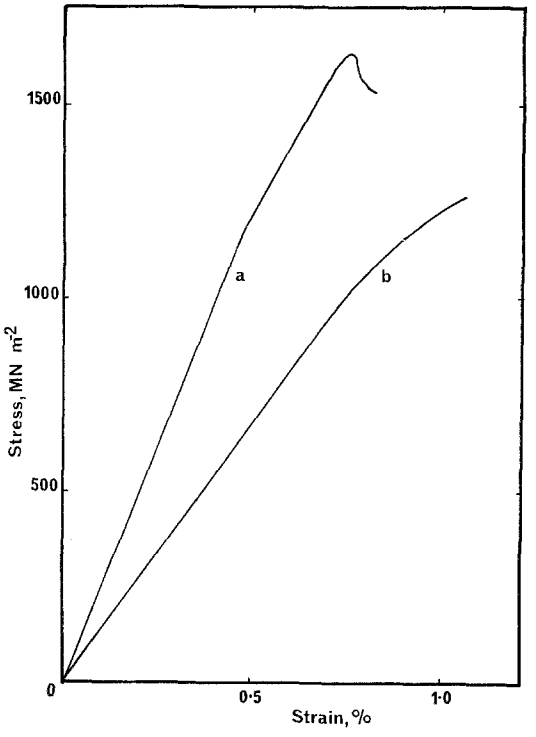

Figure 3 Stress-strain curves for the Ni–Al–Cr–C alloys tested at room temperature (a) fully aligned, (b) cellular.

elongated dimples of the matrix drawn out by ductile tearing surrounding broken Cr_3C_2 fibres (Fig. 4). A few areas of the fracture surface showed misaligned carbides where the fracture had evidently taken place on the fibre/matrix interface (Fig. 5). Detailed examination of the broken fibres showed that they had fractured in a brittle fashion in a similar manner to Al_3Ni [4] and Cu_6Sn_5 [5]. Part of the fibre fracture surface had a smooth profile and the remainder was roughened as the fracture energy was dissipated by creating additional surface area. The fracture behaviour of Cr_3C_2 was analogous to an amorphous glass rather than a crystalline material where better defined cleavage facets would be formed, and tests conducted on individual fibres chemically extracted from the matrix also showed brittle behaviour similar to a glass [6]. Micrographs of longitudinal sections showed that Cr_3C_2 fibres had been fractured into short lengths throughout the gauge-length (Fig. 6). The aspect ratio of the fractured segments was of the order of ten which indicates that stress transfer from the matrix to the fibres is very effective. In tensile tests in which the specimen was unloaded at 1350 MN m^{-2}, no broken fibres were observed.

Fibre fracture occurs close to the UTS in this material but this does not lead immediately to failure as, for example, in the Al–Al_3Ni eutectic [7] because the relatively high shear strength of the matrix makes stress transfer by shear of the matrix on planes of maximum resolved shear

Figure 4 Tensile fracture surface of the aligned Ni–Al–Cr–C alloy tested at room temperature showing elongated dimples of the matrix around broken Cr_3C_2 fibres, SEM, × 1200.

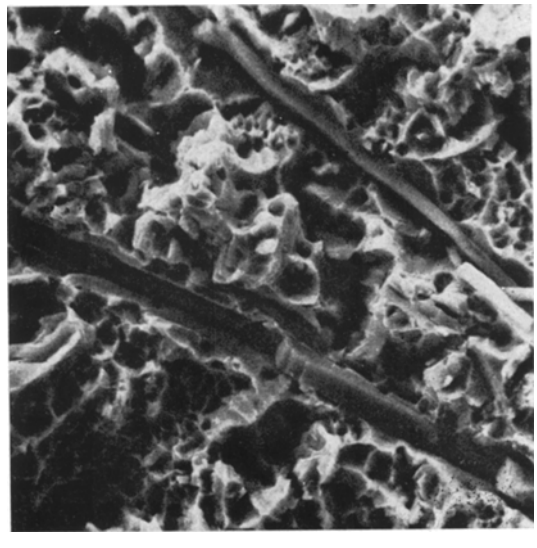

Figure 5 As Fig 4 showing fracture at the interface of a misaligned fibre, × 1200.

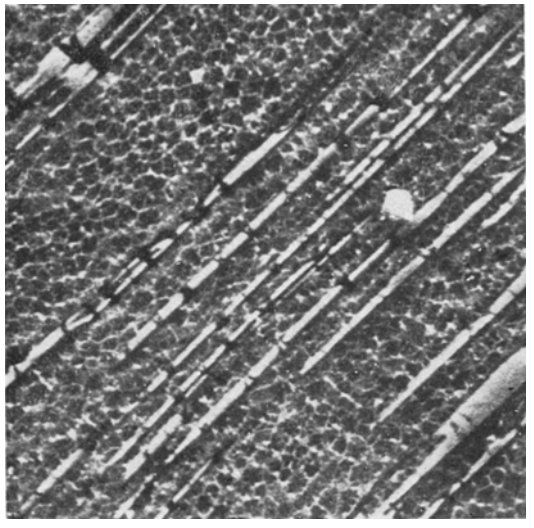

Figure 6 Longitudinal section of the Ni–Al–Cr–C alloy after tensile failure showing fractured carbides, SEM, × 2000.

stress more difficult. Although the composite fractures in a plane perpendicular to the tensile axis where the stress concentration from a broken fibre is greatest, the matrix has sufficient strength and ductility for failure not to occur until a large proportion of the fibres have been cracked into lengths corresponding to the critical stress transfer lengths. Furthermore, the actual stress at which a given fibre breaks is dependent on its size and this, too, results in the fracture process being more gradual in this material as the weaker fibres fracture at lower stresses than the stronger ones.

3.3. Tensile behaviour of the cellular alloy

The stress-strain curve of the Ni–Al–Cr–C eutectic alloy with a cellular morphology is shown in Fig. 3 (curve b). The primary modulus was 135 GN m^{-2} and the composite yielded at 1100 MN m^{-2}, after which failure occurred at 1260 MN m^{-2}. The lower primary modulus of the cellular alloy with respect to the aligned alloy may be rationalized in terms of the differences in the modulus of the matrix in different crystallographic directions. If a simple mean of the elastic compliances of Ni [8] and Ni_3Al [9] is taken, the ⟨100⟩ modulus of the matrix is 120 GN m^{-2} and the ⟨110⟩ modulus, 220 GN m^{-2}. Accepting a value of 370 GN m^{-2} for the modulus of Cr_3C_2 [10], a rule of mixtures calculation with a fibre volume fraction of 0.11 gives values of 145 GN m^{-2} and 235 GN m^{-2} for the primary modulus with [100] and [110] matrix orientations respectively. Ignoring possible effects due to the misaligned carbides, there is reasonable agreement with the observed values of the modulus of the cellular alloy if it has a predominantly [100] matrix and similarly the aligned alloys may have higher moduli as a smaller proportion of the matrix grows with the "soft" [100] orientation and more with the "harder" [110] orientation. There are discrepancies, and these may be due to positive deviations from the rule of mixtures that are expected if the Poisson's ratios of the component phases are not equal or to thermally induced residual stresses due to differential contraction of the phases. Nevertheless, the variation in primary modulus may be correlated with the change in matrix orientation associated with the transition from the aligned to the cellular morphology.

The fracture surface of the cellular alloy was macroscopically flat and in detail clearly showed the cellular nature of the morphology as failure had taken place at the fibre/matrix interface in regions where the fibres were tilted away from the tensile axis (Figs. 7 and 8). In regions where the fibres were closer to the tensile axis, a normal tensile failure characterized by elongated dimples was observed. Longitudinal sections showed that

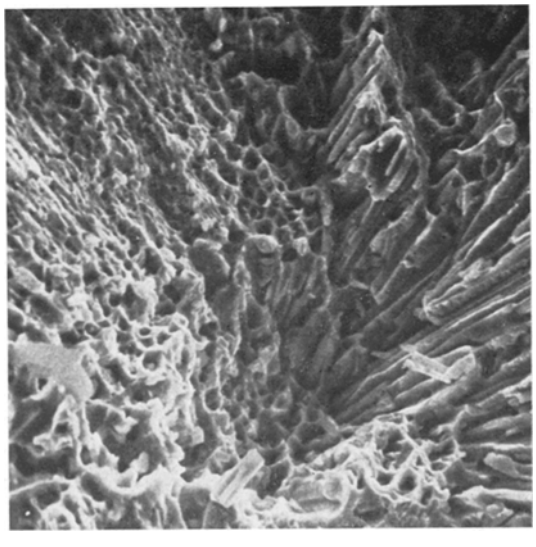

Figure 7 Tensile fracture surface of the cellular Ni–Al–Cr–C alloy showing failure at the fibre/matrix interface and elongated dimples, SEM, × 900.

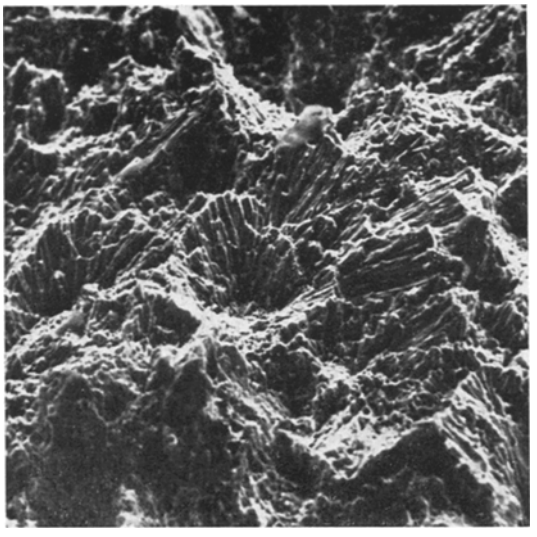

Figure 8 As Fig. 7 showing the cellular structure as revealed on the fracture surface, × 200.

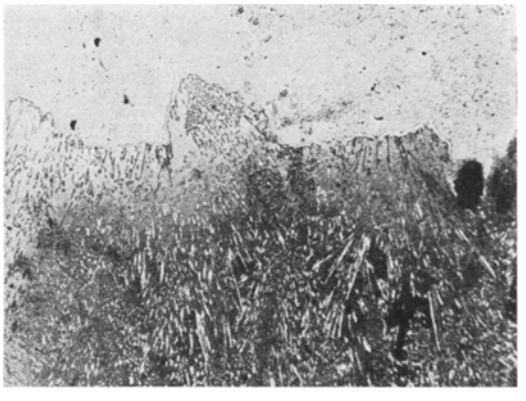

Figure 9 Longitudinal section through the fracture surface of the cellular Ni–Al–Cr–C alloy, × 65.

the fibres were not as extensively fractured as in the fully-aligned specimens (Fig. 9).

The strength of a composite shows a large reduction when tested away from the fibre axis [11]. If θ is the angle between the tensile axis and the fibre axis in a composite of UTS, σ_c, then failure initiated by fracture of the fibres occurs at a stress

$$\sigma = \sigma_c \sec^2\theta \,. \tag{1}$$

Failure by matrix shear on a plane parallel to the fibres requires a stress

$$\sigma = 2\tau_u \operatorname{cosec} 2\theta \tag{2}$$

where τ_u is the ultimate shear stress of the matrix. Failure of the matrix in tension (provided the fibre/matrix interface does not fail first) occurs at a stress

$$\sigma = \sigma_u \operatorname{cosec}^2\theta \tag{3}$$

where σ_u is the ultimate tensile strength of the matrix in plane strain. The consequent variation in composite strength is summarized in Fig. 10. Failure occurs by the mode which requires the lowest applied stress and the critical angle (θ_{crit}) above which the strength decreases rapidly is given by $\tan^{-1} \tau_u/\sigma_c$. In a cellular eutectic, in any one cell the fibres may be considered to vary continuously in orientation between $\pm\theta_{max}$ across the cell, where θ_{max} is the maximum inclination of the fibres to the tensile axis. Thus, if the composite is considered as a set of parallel slabs each containing fibres at an angle θ to the tensile axis with a small spread of orientations $d\theta$, then each element will have a strength $d\sigma$ given by

$$d\sigma = \sigma_c \sec^2\theta \, d\theta \tag{4}$$

between 0° and θ_{crit}, and a strength

$$d\sigma = 2\tau_u \operatorname{cosec} 2\theta \, d\theta \tag{5}$$

between θ_{crit} and θ_{max}. Hence the strength of the composite may be calculated by integrating across the cell between $\pm\theta_{max}$

$$\sigma = 2\left(\int_0^{\theta_{crit}} \sigma_c \sec^2\theta \, d\theta + \int_{\theta_{crit}}^{\theta_{max}} 2\tau_u \operatorname{cosec} 2\theta \, d\theta\right) \tag{6}$$

$$= 2\sigma_c \left[\tan\theta\right]_0^{\theta_{crit}} + \tau_u \left[\log_e \tan\theta\right]_{\theta_{crit}}^{\theta_{max}} . \tag{7}$$

Experimentally, θ_{max} was measured as 30° but τ_u was not known and therefore using $\theta_{crit} = \tan^{-1} \tau_u/\sigma_c$, θ_{crit} was adjusted in order to predict the observed reduction in strength and found to be ~ 10°. This gives a value of τ_u of 300 MN m^{-2} that is much lower than values quoted (~ 450 MN m^{-2}) for polycrystalline Duranickel [12] which has a composition similar to the matrix (Ni–4.5 Al). Furthermore, George *et al.* [13] found that in the Al–Al_3Ni eutectic tested away from the fibre axis, constraint effects raised the apparent value of τ_u by a factor

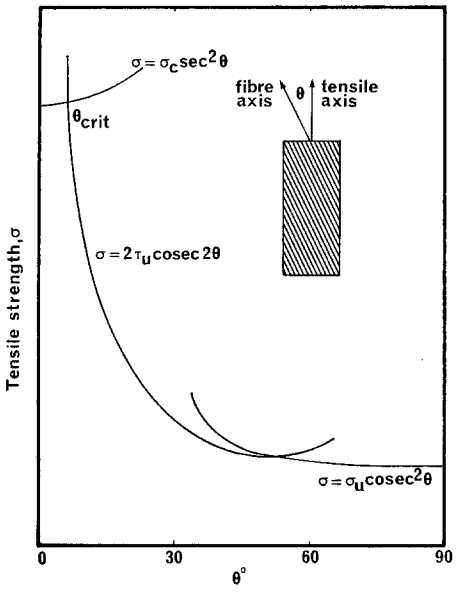

Figure 10 Variation of tensile strength with angle θ between aligned continuous fibres and the tensile axis.

of at least two with respect of bulk values. In the cellular Ni–Al–Cr–C alloy, however, experimental observations showed that fracture took place at the fibre/matrix interface which is evidently weaker than the matrix and this may account for the low value.

4. Conclusions

The strength of an Ni–Al–Cr–C eutectic composite which has Cr_3C_2 fibres in a γ,γ' matrix has been measured both with a fully-aligned and a cellular micromorphology. The fully-aligned eutectic behaved as a simple metal-matrix composite and fracture was initiated by brittle fracture of the Cr_3C_2 fibres. The cellular eutectic was weaker than the aligned eutectic, and using a simple analysis, the reduction in strength due to the off-axis reinforcement was shown to result from fracture at the fibre/matrix interface which was correlated with observations of the fracture surface.

This demonstrates the importance, not only of keeping the ratio τ_u/σ_c as high as possible to minimize the sensitivity of the composite strength to off-axis loading or misalignment of the reinforcing phase, but also of maximizing the strength of the fibre/matrix interface. This has many practical consequences for the application of eutectic composites, but relatively little attention has been paid to the properties of interfaces in composites with regard to their fracture behaviour.

Acknowledgements

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References

1. J. VAN DEN BOOMGAARD and A. M. J. VAN RUN, *J. Mater. Sci.* **8** (1973) 1095.
2. G. J. MAY, to be published.
3. R. W. KRAFT and D. L. ALBRIGHT, *Trans. Met. Soc. AIME* **221** (1961) 95.
4. N. D. BROOM and G. J. DAVIES, *Phil. Mag.* **28** (1973) 685.
5. G. J. DAVIES and N. D. BROOM, *ibid* **26** (1972) 1243.
6. N. D. BROOM, Department of Metallurgy and Materials Science, University of Cambridge, private communication.
7. B. CANTOR, G. J. MAY and G. A. CHADWICK, *J. Mater. Sci.* **8** (1973) 830.
8. H. B. HUNTINGTON, *Solid State Physics* **7** (1958) 213.
9. K. ONO and R. STERN, *Trans. Met. Soc. AIME* **245** (1969) 171.
10. G. V. SAMSONOV, "High Temperature Materials" No. 2, Properties Index (Plenum Press, New York, 1964).
11. E. Z. STOWELL and T. S. LIU, *J. Mech. Phys. Solids* **9** (1961) 242.
12. "Metals Handbook", Vol. 1, 8th Edn. (ASM and Chapman and Hall, 1961).
13. F. D. GEORGE, J. A. FORD and M. J. SALKIND, "Metal Matrix Composites", ASTM STP 438 (1968) p. 59.

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