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Strain inhomogeneities in eutectic composites

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Transmission and replica electron microscopy have been performed on deformed samples of the as-quenched Ni–W eutectic composite, which deforms extensively by slip in both phases. Two types of strain inhomogeneities were found for samples loaded above the yield point. Longitudinal plastic strain inhomogeneities were a result of the banding of matrix slip. This banding sometimes caused sufficient stress concentrations in the reinforcement phase that it yielded locally at applied stresses well below its macroscopic yield point. Transverse plastic-strain inhomogeneities in the matrix were caused by the preferential yielding of the matrix at or near the reinforcement/matrix interface. The matrix strain gradient caused by this local yielding did not appear to persist for large strains due to the fairly long slip-line length relative to the interfibre spacing. A second mode of transverse strain inhomogeneity was the local yielding of only a fraction of the W fibres intersected by each slip band. Yielding at the reinforcement/matrix interface could be seen to occur at both fibre corners and near ledge-type defects on the flat faces. Because of the severe compatability constraints imposed by the bicrystal-like deformation of the composite, secondary slip systems operated locally for plastic strains as low as 0.4%.

1. Introduction

For most purposes the deformation of each phase of a eutectice composite may be treated as internally homogeneous. Consequently, little has been written regarding systematic longitudinal or transverse strain inhomogeneities which might be found. Gould and Martin [1] have proposed a strengthening mechanism based on the formation of loops around fibres, and yielding has been analysed in terms of pile-ups [2], or the stress necessary to operate dislocation sources in the matrix of a eutectic composite [3]. None of these investigators report direct observations of strain inhomogeneities, however. The best graphic evidence for strain inhomogeneities has been the high dislocation densities found near the fibres in composites rapidly cooled from high temperature, which presumably developed to relieve elastic misfit strains [4]. Also, Pattnaik and Lawley [5] have concluded from studies of deformed $Al–CuAl_2$ eutectic that the matrix dislocation density is higher near the lamellar interface. In the present paper we discuss the occurrence of transverse and longitudinal strain inhomogeneities in eutectic composites. Taken together, the presence of transverse and longitudinal strain gradients means that use of the rule of mixtures and the isostrain assumption of composite behaviour must be used with care in discussing eutectic composites.

2. Experimental procedure

The nickel–45.5 wt % tungsten eutectic was used for this study because of its ductility and prior characterization [6-8]. Details of alloy preparation and directional solidification have been given previously [6, 7]. The alloy as directionally solidified contains approximately 6 vol % tungsten fibres in a nickel–tungsten solid-solution matrix, which itself contains precipitates of WNi_4. For the present study the material was quenched from 1050° C, a temperature above the peritectoid, so that the WNi_4 precipitates were dissolved. Compression specimens were machined from the quenched bars. Flats were ground on the sides of the compression specimens, and these were polished electrochemically. The compression specimens were then loaded to successively higher stresses, separated by unloading and surface replication. The replicas were shadowed with chromium and viewed in a JEM-T7 electron microscope. Some compression specimens were sectioned at various loading points, thin foils were prepared, and the foils were viewed in a Philips EM-300 electron microscope using a double-tilting specimen holder.

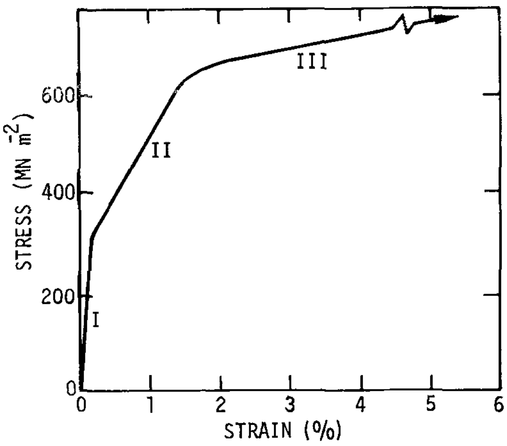

Figure 1 Stress–strain curve of early part of the deformation of the as-quenched Ni–W eutectic in compression.

3. Results

The early portion of the plastic stress-strain curve of the Ni–W composite is given in Fig. 1. The regions I, II and III are marked to correspond to the strain ranges where both phases are elastic, the matrix is plastic and the reinforcement elastic, and both phases are elastic, respectively [9]. The results discussed in this paper are from specimens strained 0.4% beyond the end of stage I. At this point the composite is approximately at the midpoint of stage II: the matrix is plastic and the reinforcement remains elastic, according to this simplified approach.

At this strain, slip lines can be seen in the matrix (Fig. 2) but slip lines can also occasionally be seen in the reinforcement. The slip lines in the reinforcement are continuous with slip lines in the matrix and also are found only where concentrations of matrix slip occur (Fig. 2). In this composite the matrix slip is slightly banded in character, and slip in the reinforcement can be found only where a band intersects the reinforcement. Portions of the tungsten reinforcement which are intersected by the matrix slip band may or may not be forced to slip, as can be seen for adjacent intersected fibres in Fig. 2. Therefore, it is clear that the proximity of fairly heavy matrix slip concentrations may cause the intersected fibres to slip locally well below the point at which they might be expected to slip homogeneously. Intersected portions of the fibres may remain elastic, but if slip occurs in the reforcement, it is almost always continuous with matrix slip lines. When a portion of a fibre yields, its reduced increment of load-carrying ability with increased strain is compensated for by matrix hardening. Thus the portion of the fibre which has yielded cannot elongate at a higher rate than the

Figure 2 Replica of specimen strained 0.4% beyond the proportional limit. One of the two fibres has yielded, but only where it has been intersected by the slip band in the matrix. Slip on three matrix slip systems can be identified, at one point being nucleated at the reinforcement/matrix interface (arrow).

neighbouring unyielded material, and the localized fibre yielding cannot be detected externally.

Even where the matrix has strained only 0.4% past the proportional limit, slip on systems in addition to the primary system can be seen. Fig. 2 shows that, within the matrix slip band, slip on at least three systems can be identified. The reinforcement/matrix interface seems to be a preferred nucleation site for the slip, as seen in Fig. 2 for slip on one of the secondary systems. The early production of slip on secondary systems probably contributes to work-hardening of the matrix by a latent-hardening mechanism. This increment of flow stress is entirely independent of the pile-up mechanisms usually invoked to explain deformation behaviour of eutectics of varying reinforcement spacing.

Transmission electron microscopy of thinned foils provides additional information on the nature of the initiation of slip at or near the reinforcement/matrix interface. Both the corners of fibres and their flat faces can act as nucleation sites for slip (Figs. 3 and 4). The corner sources generate loops which expand under stress. The loops are truncated during the preparation of the thin foil, thereby producing the characteristic dislocation array of Fig. 3. Fig. 4 shows that the matrix slip dislocations which nucleate on the flat reinforcement/matrix interfaces are associated with planar defects lying permanently in that interface. No attempt was made to ascertain the Burgers or displacement vector of the interface defects in strong contrast in Fig. 4, but their nature can be deduced from the thickness-fringe displacement which occurs at the defect. Displacement of thickness fringes has been associated with ledge-type interfacial defects by Gleiter [10] for the case of grain boundaries, and by Garmong and Rhodes [11, 12] for the case of eutectic interface ledge-like defects, such as those sometimes found on curved portions of a eutectic interface. Therefore, the interfacial defects shown in Fig. 4 are most probably interfacial ledges of some type. In Fig. 4 it should be noted that the matrix slip dislocations appear to originate only at the interfacial ledge defect, and not at any point on the otherwise nearly smooth portion of the interfacial plane.

No mechanism could be positively identified for the production of matrix slip dislocations at fibre corners and at interfacial ledges. From Figs. 3 and 4 it is clear, however, that whatever the mechanism, it must be capable of producing extended trains of dislocations, not just one or two before exhaustion. Detailed studies of the mechanisms of production of lattice slip dislocations from grain-boundary sources suggest by analogy that for each matrix slip dislocation a eutectic interfacial dislocation must be produced for a continuous source to operate [13, 14]. In the present investigation a search was made for such a process with limited success. Near operating ledge sources features with weak contrast could

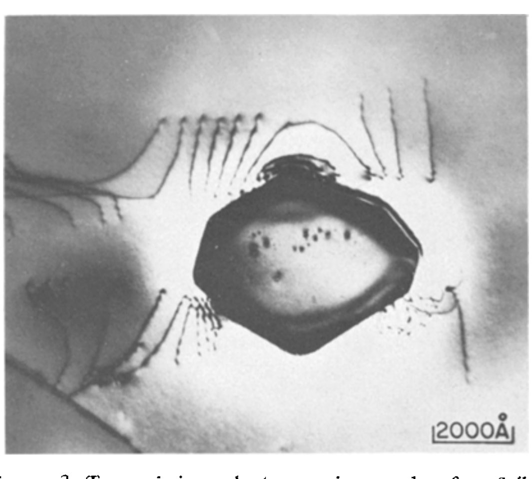

Figure 3 Transmission electron micrograph of a foil prepared from a transverse section of the composite strained 0.4% beyond the proportional limit. Matrix slip dislocations nucleate at both a sharp corner and on the flat face.

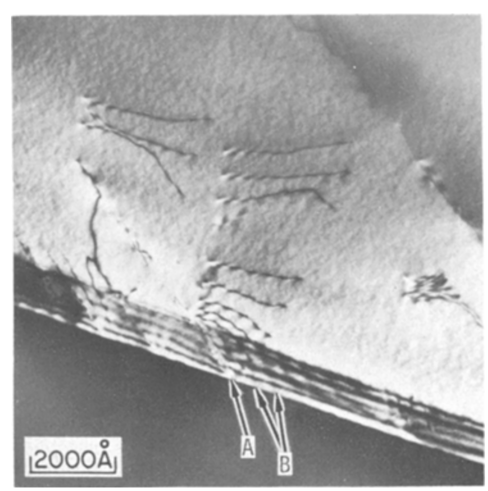

Figure 4 Transmission electron micrograph of longitudinal-section foil from a specimen strained 0.4% beyond the proportional limit. Matrix slip dislocations nucleate at the linear interface defect A, which can be identified as a ledge-type defect by the thickness-fringe displacements. Faint features B are associated with operating ledge-type sources and may be interfacial dislocations.

be observed to lie in the interfacial boundary (Fig. 4), but no condition of observation could be found in which the features produced images of sufficient strength and clarity that detailed analysis of their character could be undertaken. These interfacial features do not produce thickness fringe displacements, but beyond this it was impossible to determine whether they could be uniquely associated on a one-to-one basis with the matrix slip dislocations being generated at the ledge-type defect.

In this and previous extensive investigations [6-8] of the Ni–W eutectic composite system no sources of matrix dislocations have been found which are entirely independent of reinforcement/matrix interfaces and grain boundaries. While it is always difficult to prove experimentally a negative assertion, this result, coupled with the fairly frequent observation of reinforcement/matrix interfacial flat-face ledge sources and corner sources, suggests that the major part of matrix slip at low strains in this system is nucleated in the interface or the near-interfacial regions.

4. Discussion

The strain inhomogeneities found in this composite after small strains may be distinguished as transverse and longitudinal. Transverse inhomogeneities are a result of the nucleation of slip near reinforcement/matrix interfaces. Because the slip length of the matrix is quite long in this material, the inhomogeneity is not pronounced. That is, even though slip nucleates near the interface, it soon spreads over wide areas of the matrix, and the stress concentrations related to the impingement of a matrix slip line against a normally full elastic reinforcement may be sufficient to cause yielding of the reinforcement. The stress supported by a composite is often expressed by the following equation, known as the rule of mixtures or law of combined action: $\sigma_c = V_r\sigma_r + V_m\sigma_m$, where σ is stress, V is volume fraction, and r and m refer to reinforcement and matrix. If the individual stresses are taken to be the *in situ* average stresses of the phases, then the equation must be correct. The present results suggest that σ_m could be further expressed as $\sigma_m = V_{me}\sigma_{me} + V_{mp}\sigma_{mp}$, where the additional subscripts e and p refer to regions which are elastic and plastic due to the matrix strain inhomogeneity. Such an approach has been previously suggested for other types of composites [15]. In the present case this approach would be somewhat unrealistic, since the transverse matrix inhomogeneity is localized for only a brief period of strain. However, in the case of the reinforcement, the equation $\sigma_r = V_{re}\sigma_{re} + V_{rp}\sigma_{rp}$ might be useful in precisely describing the stress-strain curve of the composite, if the transverse fraction of the reinforcing phase that is plastic at any strain could be evaluated.

The longitudinal strain inhomogeneity appears to be related to the slight banding of matrix slip found in the as-quenched Ni–W eutectic. As seen in Fig. 2, slip in both the matrix and the reinforcement may be longitudinally localized, with extensive deformation in some regions and virtually none in longitudinally adjacent regions. Normally, the deformation strain ϵ_c of a composite is assumed constant and homogeneous throughout the length of the composite. In the present case, however, the total strain over the entire gauge length of the test specimen might better be expressed by a longitudinal series rule of mixtures, $\epsilon_c = l_e\epsilon_e + l_p\epsilon_p$, where l is a metallographically determined length fraction. The logical concomitant of this equation is a realization that some regions of the reinforcement may be elastic to higher strains than others. The break in the stress–strain curve of this composite between stages II and III of deformation (Fig. 3), is undoubtedly the onset of general yielding of the reinforcement, as distinct from the premature yielding of those regions of the reinforcement in which stress concentrations are created by impingement of matrix slip bands. Thus, the effectiveness of normally whisker-like reinforcement phases may be markedly diminished due to localized yielding in the presence of matrix stress concentrations in alloys where the matrix deformation is banded. In the present case, the as-quenched Ni–W eutectic alloy exhibits slightly banded slip, but in the same alloy given subsequent ageing treatment slip becomes quite heavily banded [8]. This heavy banding apparently causes even earlier localized yielding of the reinforcement due to the large stress concentration [6]. Moreover, the generally higher level of matrix flow stress caused by the ageing treatment exerts even greater constraints and thereby stress concentrations in the reinforcement away from the slip bands. As a result the reinforcement yield strain, as measured by the observation of the macroscopic stage II/stage III transition, decreases with further ageing. Generally, strengthening the matrix of a eutectic composite may adversely affect the yield strength of the reinforcement through greater localization of the matrix slip and through a generally higher level of matrix elastic constraint on reinforcement deformation. Since a number of the eutectic composite superalloys have precipitation-hardened matrices [4], the expected banding of slip in these systems could seriously reduce the extent of the elastic deformation of the reinforcement.

5. Conclusion

Two types of strain inhomogeneity have been observed in deformed Ni–W eutectic composites Longitudinal strain inhomogeneities are present as a result of banded matrix slip. Transverse strain inhomogeneities are caused by preferential slip production at the reinforcement/matrix interface of the composite and by the yielding of only a fraction of the fibres intersected by slip bands. The important technical result is that matrix strengthening procedures may lead to decreased reinforcement yield or ultimate strength, even in the absence of chemical effects.

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