JOURNAL OF MATERIALS SCIENCES 13 (1978) 2380–2384

Grain-boundary sliding and intergranular cavitation during superplastic deformation of α/β brass

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Intergranular and interphase cavitation in binary alpha/beta brass has been investigated in tension at 600° C under conditions of superplastic deformation. The sites for nucleation of cavities has been studied by quantitative metallography and the cavities are observed to nucleate preferentially at α–β interfaces. The process of cavitation is associated with grain boundary sliding and cavity nucleation occurs at points of stress concentrations in the sliding interfaces. Measurements of grain and phase boundary sliding at various interfaces demonstrate that sliding occurred on α–β boundaries more readily than on α–α and β–β interfaces. The predominance of α–β interface cavitation is believed to be as a result of greater sliding at the α–β boundary and of an unbalanced accommodation of sliding adjacent to this type of boundary.

1. Introduction

Grain boundary sliding is one of the major strain contributing mechanisms of flow in superplastic deformation. Since grain boundary sliding is also the process which is primarily responsible for the intercrystalline failures common to creep resistant materials [1] it is interesting that intergranular cavitation does not occur in most superplastic alloys. However, several binary alloys of copper [2, 3] and aluminium [4, 5] and some iron–carbon [6, 7] alloys have been found to cavitate at grain and phase boundaries during superplastic straining. The studies of Dunlop *et al.* [5] in Al-bronze and recently Sagat and Taplin [8] in ternary brass have shown that cavities form preferentially at α–β interfaces or triple junctions involving both phases. The α–α sites are less favoured and very few cavities form at β–β sites.

An important microstructural feature which might have a significant effect on ductility at creep temperatures is the type of boundary at which sliding occurs. To date, very little attention has been given to the separate roles played by the interface boundaries (e.g. α–α) and the interphase boundaries (e.g. α–β), although some preliminary results on grain boundary sliding in β and β–γ brasses [9] are available. Recently, Eberhardt and Baudelet [10] have undertaken the sliding measurements at phase boundaries in the lamellar Pb–Sn eutectic and found that the ratio between the maximum rate of phase boundary sliding and the interlamellar spacing in two Pb–Sn alloys was approximately constant. However, their studies could not give any information about the roles which various interface and interphase boundaries might play in high temperature fracture of multiphase materials as Pb–Sn alloys do not cavitate on straining above 0.5 T_m. Since cavitation is generally associated with grain and phase boundary sliding, it is likely that differences in the rate of sliding at single phase and two-phase boundaries might influence the cavitation and thus final fracture behaviour. No attempt has been made to examine the influence of relative proportions and type of interfaces on flow and fracture characteristics of duplex alloys from the standpoint of interface sliding. The aim of the present investigation is to study the sliding at various grain and phase boundaries in α–β brass under conditions of superplastic deformation and discuss its effect on intergranular cavitation and high temperature deformation behaviour.

TABLE I Composition, average grain size and α–β proportions of binary brass

Cu	Zn	Pb	Bi	Average grain size, d (μm)	α/β ratio at 600° C
Remainder	40.5	0.004	0.03	32	48:52

2. Experimental

2.1. Materials

The material studied was an industrial binary brass prepared by Imperial Metal Industries, Birmingham, UK. The actual composition of this alloy is given in Table I. The as-received hot extruded alloy had an average grain size of 2.8 μm. A stable grain size of 32 μm was obtained by annealing in argon at 700° C for $1\frac{1}{2}$h. The α and β phase sizes were then 30 to 34 μm respectively.

In general, alloys which possess superplasticity, have the size of the phases either both very small or vastly different. In either case there is a predominance of one type of interface. At the optimum temperature for superplasticity, 873 K, the present alloy consists of a duplex α–β structure having approximately equal volume fractions of alpha and beta phases and thus various interfaces (interphase and ordinary grain boundaries) are present in significant quantities.

2.2. Mechanical tests and grain-boundary sliding measurements

Mechanical testing was carried out in uniaxial tension on an Instron Universal testing machine fitted with a push-button speed selector gear box. The test temperature was maintained by a three-zone split, vertical, furnace. All tests were conducted at 600° C and a constant true strain rate of $7.7 \times 10^{-5}\,\text{sec}^{-1}$. Under these conditions of temperature and strain rate, the present alloy has been observed to show superplasticity reported in detail elsewhere [11]. All specimens used for grain boundary sliding measurements were strained in high purity argon atmosphere and the tests were interrupted at different levels of strain. Before testing the specimens were electropolished in 1/3 nitric acid; 2/3 methanol. The progress of grain boundary sliding was monitored by measurements of the average step height v at α–α, α–β, and β–β boundaries by means of a two-beam interferometer working at a lateral magnification of 700×. By using monochromatic (yellow) light having a wavelength of 0.6 μm, it was possible to measure steps of 0.1 μm or less. The height of the step, v, was measured at 100 randomly selected bound-

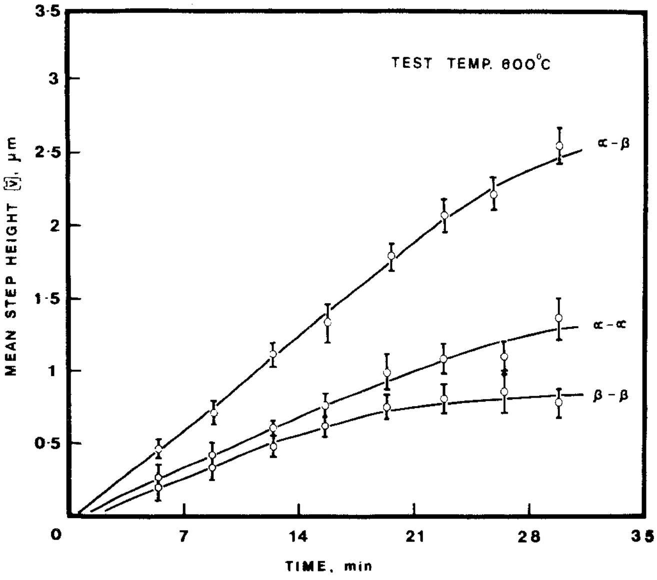

Figure 1 Mean step height ($\bar{v}$) at α–α, α–β and β–β boundaries plotted against time of straining for binary brass specimens deformed at 873 K and $\dot{\epsilon} = 7.7 \times 10^{-5}\,\text{sec}^{-1}$.

aries which gave the mean step height, $\bar{v}$, typically of the order of ± 5% at the 95% confidence limit. The experimental technique for measuring the height of step is described in more detail in [12] and [13].

After testing the specimens were rapidly cooled and sectioned longitudinally for optical and quantitative metallography. Polishing was carried out mechanically and by the method of attack skid-polishing described elsewhere [14] to preserve the true cavity dimensions and provide a damage free surface.

TABLE II Grain and phase boundary sliding rates at various interfaces in α–β brass strained at 600° C and $\dot{\epsilon} = 7.7 \times 10^{-5} \text{sec}^{-1}$

Type of boundary	Sliding rate (cm sec^{-1})
α–α	8.6×10^{-8}
α–β	1.5×10^{-7}
β–β	6.8×10^{-8}

3. Results

Grain boundary sliding at α–α, α–β and β–β boundaries was measured as a function of deformation time at constant true strain rate of $7.7 \times 10^{-5} \text{sec}^{-1}$ at 600° C. Fig. 1 records the amount of sliding in terms of the mean step height, $\bar{v}$, plotted against time. The figure indicates that $\bar{v}$ increases linearly with time in the initial stages of deformation at α–α, α–β and β–β boundaries but only a gradual decrease in mean step height was observed at higher strains. This kind of curve between sliding displacement and deformation time is commonly observed [15]. The sliding rates (after seven minutes of straining) were, however, different at different boundaries, and these rates have been documented in Table II. It was found that there was little significant difference between ratios of rates of sliding at the three types of boundary at different levels of tensile strains. The sliding rate ($\dot{\epsilon}_{\text{gbs}}$) was calculated from the measurements of mean step height $\bar{v}$ using the relation $\bar{v}$/(time of the test) [16].

The optical micrographs (Figs. 2a and b) of binary brass deformed in tension at a strain rate of $7.7 \times 10^{-5} \text{sec}^{-1}$ and 600° C show that both number and volume fraction of cavities increase with strain. The influence of the average sliding displacement on the per cent cavitation of α–α, β–β and α–β boundaries is recorded in Figs. 3a to c. The results indicate that the per cent cavitation of individual boundaries increases as the extent of grain boundary sliding increase and the α–β boundaries exhibit much greater cavitation than α–α and β–β interfaces under similar conditions of deformation.

4. Discussion

It is widely recognized that grain boundary sliding is a prerequisite for the nucleation of grain boundary cavities [17] and that nucleation results when sufficient stress concentration is generated to create new surfaces. The gradual development of cavitation with increasing strain during deformation of present α–β brass is consistent with nucleation by grain boundary sliding. The occurrence of sliding along at various grain and phase boundaries leads to stress concentrations sufficient to nucleate stable voids. The results (Fig. 3) indicate

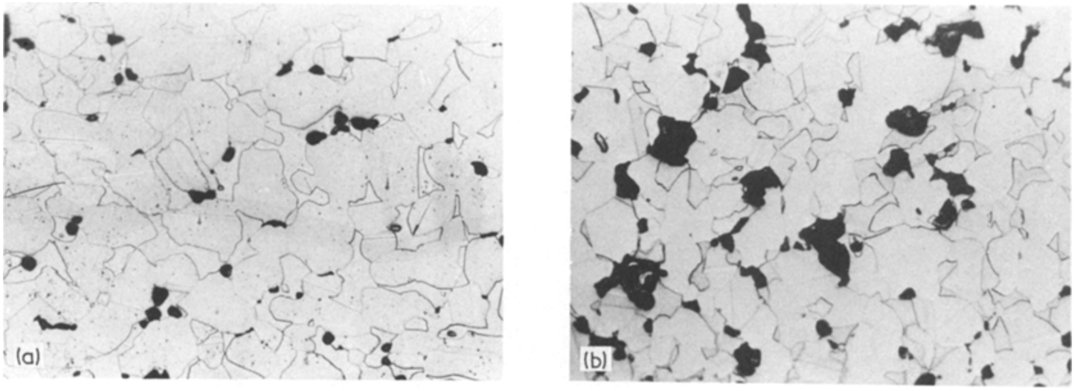

Figure 2 Optical micrographs of α–β brass strained at 873 K and $\dot{\epsilon} = 7.7 \times 10^{-5} \text{sec}^{-1}$ to (a) 50% elongation and (b) 95% elongation. Tensile axis is horizontal. Voids form at various grain and phase boundaries and both number and volume fraction of voids increase with strain (× 450).

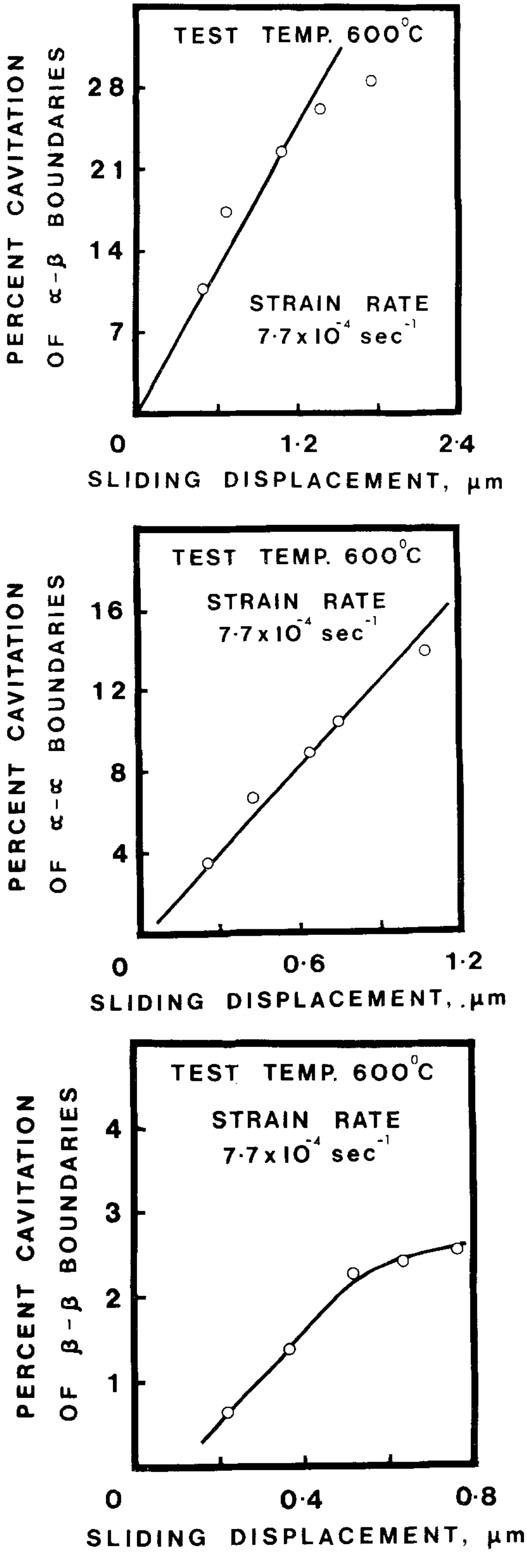

Figure 3 The influence of average sliding displacement on the per cent cavitation of α–α, α–β and β–β boundaries in α–β brass strained at 873 K and $\dot{\epsilon} = 7.7 \times 10^{-5}\,\text{sec}^{-1}$.

that cavities are formed preferentially on α–β boundaries during the high temperature straining of α–β brass. The preponderance of α–β interface cavitation could arise from one or more of the following causes:

(i) different phase boundaries may require different energies to nulceate a cavity;

(ii) different rates of sliding may occur at α–α, α–β and β–β boundaries;

(iii) a different ease of accommodation of sliding may apply to each type of boundary.

The critical energy (E) for nucleation of a cavity is proportional to γ_s^3/σ^2 where the surface energy (γ_s) defines the stable cavity size (r) in terms of the applied stress, σ, as given by $r_{\text{critical}} = a\gamma_s/\sigma$ (a is a constant). These relations suggest that γ_s must have a strong effect on cavity nucleation and if similar stress concentrations arise at α–α, α–β and β–β phase boundaries, that interface which possesses the lowest energy will cavitate first.

The energy ($E_{\alpha-\beta}$) required to create a cavity at an α–β interface per unit area of interface occupied by the cavity is:

$$E_{\alpha-\beta} \simeq \gamma_{s\alpha} + \gamma_{s\beta} - \gamma_{\alpha\beta}$$

where $\gamma_{\alpha\beta}$ is the surface energy of the α–β boundary, and $\gamma_{s\alpha}$ and $\gamma_{s\beta}$ are the free surface energies of the α and β phases respectively.

Similarly the energies $E_{\alpha-\alpha}$ and $E_{\beta-\beta}$ to create cavities at the α–α and β–β interfaces are:

$$E_{\alpha-\alpha} \simeq 2\gamma_{s\alpha} - \gamma_{\alpha}$$

$$E_{\beta-\beta} \simeq 2\gamma_{s\beta} - \gamma_{\beta}$$

Where γ_α and γ_β are the surface energies of grain boundaries within α and β respectively. Since $\gamma_{\alpha-\beta} \simeq 0.7\,\gamma_\beta$ [18] and $\gamma_s \simeq 3\gamma_{\text{gb}}$ [19] the above equations become:

$$E_{\alpha-\beta} \simeq 2.3\gamma_\alpha + 3\gamma_\beta = 5.2\gamma_\beta$$

$$E_{\alpha-\alpha} \simeq 5\gamma_\beta$$

$$E_{\beta-\beta} = 5\gamma_\beta$$

These calculations suggest that if similar stress concentrations arise at different interfaces then cavity formation will not be greatly favoured on any one boundary type. Hence, while the surface energy term may be important, it does not explain why cavities form at α–β interfaces preferentially.

The predominance of α–β interface cavitation is, therefore, believed to be either the result of greater sliding at the α–β boundary, or of a different ease of accommodation of sliding to each type of boundary. The results of extent of sliding at α–α, β–β and α–β boundaries (Fig. 1) show that maximum sliding occurred at the α–β interfaces while β–β boundaries exhibited very little sliding. In fact sliding rate (Table II) at α–β interfaces is found to be about $1\frac{1}{2}$ times that at α–α and about $2\frac{1}{2}$ times that at β–β boundaries. Since maximum amount of cavitation is associated with the maximum amount of sliding [16], the results of the present investigation are consistent with the predominance of α–β interface cavitation and very small extent of cavitation at β–β boundaries.

Apart from the differences in sliding characteristics at the various boundaries, the preferential sites of cavitation can also result from the mismatch in recovery rate between the α and β phases across the interfaces. The low stacking fault energy of the α phase makes dislocation processes like cross-slip, climb and node unpinning more difficult. It has also been found that β phase is softer (lower flow stress) than α at high temperatures, resulting in greater accommodation of grain boundary sliding by the β phase. The result of this difference in recovery rate in the two phases is that the relief of stress concentration by plastic flow on the α phase side of the interface may be relatively difficult compared with the β phase side, so the cavities nucleate because of insufficient flow in the α phase. It can, therefore, be concluded that the predominance of α–β interface cavitation is the combined result of greater sliding and unbalanced accommodation at this type of phase boundary.

Acknowledgements

This work has been supported by the National Research Council and Defence Research Board of Canada. Discussions with Dr S. Sagat and Professor G. Brinson are gratefully acknowledged. The authors are also grateful to Professor R. C. Gifkins of the University of Melbourne for his critical review of the manuscript.

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Received 20 January and accepted 18 April 1978.