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In-situ hot microhardness study of cavitation in superplastic 60/40 microduplex brass

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The hardness of the individual phases of the 60/40 brass have been measured *in situ* over the temperature range 20 to 700° C. Activation energy values for the alloy and its phases were determined from the hot hardness data. Such data are utilized to understand the occurrence of cavitation during superplastic deformation of this alloy. It is also suggested how cavitation may be minimized by lowering the deformation temperature to 250° C where the alloy still exhibits large elongations.

1. Introduction

Usually, for a material to show superplasticity it must have a fine ($< 10\,\mu$m) equiaxed grain size which remains stable at the temperature of deformation. This type of microstructure can be obtained by producing a mixture of two phases in which the phases are present in approximately equal proportions. Although a large number of alloys have been rendered superplastic in this way, only a few are available commercially [1–3]. The reason for this is that many superplastic alloys based upon better-known alloys which have been used for many years in industry (e.g. brasses and steels) often fail at relatively low elongations ($< 400\%$) due to extensive grain-boundary cavitation. Cavitation in α/β brasses has been studied extensively [4–9] and in the case of 60/40 brass, cavities were found to be nucleated preferentially at the α/β boundaries and triple points. In this paper an *in situ* hot microhardness study of cavitation in 60/40 microduplex brass is reported and suggestions are made as to how it may be minimized by lowering the deformation temperature to 250° C where the alloy still exhibits large elongations.

2. Experimental details

The composition and proportion of α and β phases in 60/40 brass vary with temperature, as indicated in Table I. Knowing this, two additional alloys were made of compositions 61.4 Cu–Zn and 54.6 Cu–Zn to represent these α and β phases, respectively, over the temperature range 300 to 700° C. These special alloys were then tested at various temperatures to obtain Young's modulus and tensile data on the individual phases (in 60/40 brass) which they represent. All material was in the form of extruded bars of 5/8 in. diameter for tensile specimens and 1 in. diameter for hardness specimens. Experimental details of hardness, Young's modulus and tensile testing have already been described in a previous paper [10].

3. Results

In situ hardness of α and β phases in 60/40 brass at various temperatures is shown in Fig. 1. At low temperatures, the β phase is harder than the α phase, the two phases being equal in hardness at 250° C, and above this temperature the β phase is much softer than the α phase. At the superplastic temperature, 600° C, the hardness of α and β phases is 17.8 and 2.8 Hv, respectively, and the ratio $H(\alpha)/H(\beta) = 6.4$ (Table II). Hot hardness results for 60/40 brass under different microstructural conditions, showing the effect of grain size and shape, are given in Fig. 2. Above 350° C ($0.5\,T_m$), the as-extruded fine-grain-size material has a lower hardness value than the annealed one, whereas the lamellar structure is harder than either. This is the reverse of the Hall–Petch relationship and is due to an increased contribution of grain-boundary diffusion and grain-boundary sliding to flow with decreasing grain size, i.e. with an increase in specific grain-boundary area [4]. From hot hardness and Young's modulus data, activation energy values for the alloy and its phases were determined [11] and are given in Table III. The total error in determining the activation energy, Q, was about ± 2 kcal mol^{-1}. The mechanical properties of the individual phases in 60/40 brass are summarized in Table II. In this paper, the three alloys 60 Cu–Zn, 61.4 Cu–Zn and 54.6 Cu–Zn will be referred to as $B_{\alpha\beta}$, B_α and B_β, respectively.

4. Discussion

Chandra *et al.* [7] measured the extent of sliding at α/α, β/β and α/β boundaries in 60/40 brass and showed that maximum sliding occurred at the α/β interfaces while the β/β boundaries exhibited very little sliding. Near the superplastic temperature, 600° C, we obtained an activation energy, Q, of 21.2 kcal mol^{-1} for the $B_{\alpha\beta}$ alloy. Now as grain boundary sliding (GBS) is the predominant mode of deformation in this alloy and sliding occurs mainly at α/β boundaries, this value of

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TABLE I Composition and proportion of α and β phases in 60 Cu–40 Zn alloy at different temperatures

Temp. (°C)	Composition (wt %)				Proportion, $\alpha:\beta$
	α phase		β phase		
	Cu	Zn	Cu	Zn	
300	63	37	53.5	46.5	68:32
400	61	39	54	46	86:14
500	61	39	55.5	44.5	88:12
600	62.5	37.5	57	43	55:45
700	64	36	59	41	20:80

Q can be referred to as the "activation energy for interphase diffusion" and is similar to the activation energy for deformation of the β phase, the latter being 22 kcal mol^{-1}. At 600° C, the β phase is in a disordered state and interdiffusion of copper and zinc atoms takes place within it [12]. Thus, so far as the diffusion accommodation of interphase-boundary sliding is concerned, the major contribution will come from the β phase, whereas the α phase with an activation energy of 42.4 kcal mol^{-1} will remain relatively inactive.

The hardness value of the phases is important: firstly, because it gives a measure of the ease or difficulty of plastic deformation by dislocation glide in the individual phases and secondly, the variation of hardness with temperature provides information about the rate of diffusion in each phase. From Table II, the mechanical properties of the phases indicate that the plastic deformation of the β phase is easier than the α phase at all temperatures between 300 and 700° C. Thus, even the dislocation accommodation of interphase-boundary sliding would be easier on the beta-phase side compared to the alpha-phase side of the α/β interfaces. Furthermore, the low value of activation energy for β phase could also imply that a thermally activated motion of dislocations will occur more readily in this phase in comparison to the α phase which requires a higher value of activation energy.

Thus, the two phases of 60/40 brass are "incompatible" at the superplastic temperature 600° C, i.e. their mechanical properties and activation energy values differ, and the alloy cavitates because of unequal contributions made by the two phases to the accommodation of interphase-boundary sliding during superplastic deformation. The stability of the duplex structure of a superplastic alloy at high temperature ($> 0.5\,T_m$) will be greater, the greater the difference in composition of the two phases. In the case of 60/40 brass, the two phases do not differ much in composition and rapid grain growth occurs during superplastic deformation (grain size after superplastic deformation was $\simeq 40\,\mu$m) which leads to large strains arising from GBS and makes the accommodation processes even more difficult.

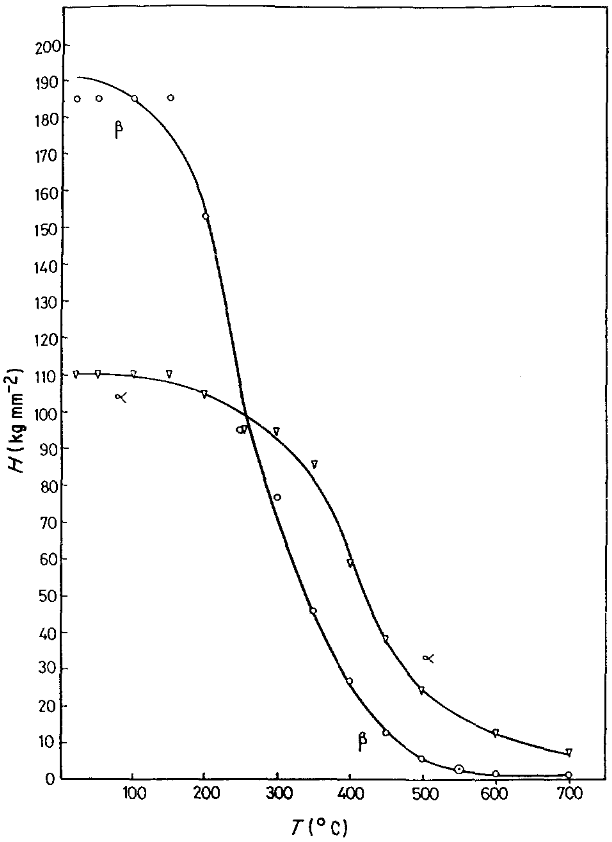

Figure 1 Hardness – temperature curves for the α and β phases, *in-situ*, of the 60/40 brass [10]. Load = 10 g, t = 15 sec, in a vacuum.

The amount of cavitation in $B_{\alpha\beta}$ was found to decrease with an increase in temperature [13], in agreement with other studies. According to Humphries and Ridley [8], if the external stress level reflects the magnitude of the stress concentrations which can develop in the material, the threshold stress for cavity nucleation will be exceeded at more potential nucleation sites as the stress level rises (i.e. as the temperature

TABLE II Mechanical properties of 60/40 brass and its individual phases at different temperatures

T (°C)	Ductility (%)			Hardness (10^3 p.s.i.*)					UTS† (10^3 p.s.i.)		
	α	β	$B_{\alpha\beta}$	*in situ*		α (B_α)	β (B_β)	$B_{\alpha\beta}$	α	β	$B_{\alpha\beta}$
				α	β						
300	40	63	47	135	109	89.6	66.8	85.3	32.4	12.9	25.3
400	25	50	29	83.8	38.3	54	20.6	34.1	32.2	10.1	14.9
500	25	180	27	34.8	8.8	22.6	3.7	11.9	6.8	0.6	4.2
600	54	172	178	17.8	2.8	12.2	1.6	4.4	3.6	0.5	1.2
700	148	151	65	9.9	2.0	4.8	1.1	1.9	0.7	0.4	0.6

*10^3 p.s.i. = 6.89 N mm^{-2}.

†UTS stands for ultimate tensile strength.

TABLE III Activation energy values for the alloys [11]

Alloy	Activation energy Q (kcal mol^{-1})	Temp. range (T_m*)
60 Cu-Zn ($B_{\alpha\beta}$)	21.2 ± 2	0.67–0.8
61.4 Cu-Zn (B_α)	42.2 ± 2	0.6 –0.8
54.6 Cu-Zn (B_β)	43.0 ± 2	0.5 –0.7
	22.0 ± 2	0.7 –0.8

*T_m = melting point in K.

decreases), thus increasing the cavity nucleation rate. On the other hand, if cavity growth was controlled by lack of diffusional accommodation, then decreasing the temperature would lead to lower diffusivities. The hardness of both phases of the $B_{\alpha\beta}$ alloy decreases with increasing temperature (Fig. 1). Consequently, during grain-boundary sliding the dislocation and diffusion processes in the deformed region near the grain boundaries would be faster at higher temperature than at lower temperatures.

At 500° C, the α phase is quite hard (22 Hv) and there is a large proportion of it ($\alpha : \beta = 88 : 12$) in $B_{\alpha\beta}$ alloy. Moreover, the α phase is less ductile (25% elongation), whereas in the β phase an elongation of 180% could be obtained at this temperature. Therefore, more cavitation occurs at this temperature than at 600° C. At 700° C, the two phases are very ductile and the ratio $H(\alpha)/H(\beta)$ is lower than its value at 600° C (see Table II) and this leads to less cavitation at 700° C. The alloy consists of a large proportion of β phase and also exhibits lower ductility at this temperature than at 600° C, probably because of rapid grain growth and hence a smaller number of α/β boundaries. Both these factors will make for less grain-boundary sliding. The individual phases were found to be elongated along the tensile axis.

At 250° C (0.45 T_m), the hardness of both phases in $B_{\alpha\beta}$ alloy is nearly the same (95 Hv, Fig. 1). The alloy was tested at different strain rates at this temperature and gave an optimum elongation of about 110% (Fig. 3). However, the alloy still exhibited cavitation at triple points involving both phases. It was felt that the grain size of the starting material was large ($\simeq 7\,\mu$m) in comparison to the very fine-grained Zn–Al eutectoid (grain size $\leqslant 1\,\mu$m) which shows superplasticity without any significant amount of cavitation at 250° C (0.7 T_m). Walser *et al.* [14] have developed a thermal mechanical process to obtain a mixture of cementite in ferrite consisting of ferrite grains $< 1\,\mu$m and cementite particles $< 0.1\,\mu$m in size. Such a steel exhibited elongations up to 750% when deformed at 650° C. The alloy does not show any cavitation during superplastic deformation. Hot hardness data on ferrite and austenite phases of 0.19 C steel and on nominally pure iron and cementite of white cast iron have been obtained by Gove [15]. At 700° C, H(cementite)/H(ferrite) = 97/22 = 4.4.

Recently, we proposed a criterion for cavitation in two-phase superplastic alloys [10]. According to this criterion, in a superplastic alloy consisting of two phases which are ultrafine (grain size $\leqslant 1\,\mu$m) and stable at the superplastic temperature, cavitation will not occur whether these phases are compatible with each other or not, i.e. they have different mechanical and diffusive properties or not. Therefore, we suggest that some kind of thermomechanical processing should be developed, or a powder metallurgical technique used to make an ultrafine microstructure (grain size $\leqslant 1\,\mu$m) in duplex 60/40 brass. Then the deformation processes may be confined mainly to the large grain-boundary area available and high ductility may be obtained without any significant amount of cavitation occurring around 250° C (0.45 T_m).

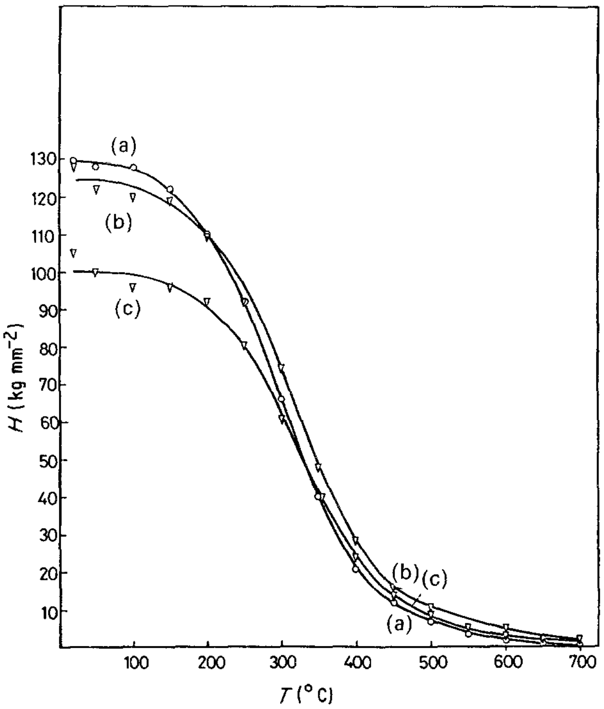

Figure 2 Hardness – temperature curves for 60/40 brass. (a) As-extruded, (b) 3 h at 800° C, air cool (lamellar structure), and (c) 1 h anneal at 600 to 620° C. Load = 100 g, t = 15 sec, in a vacuum.

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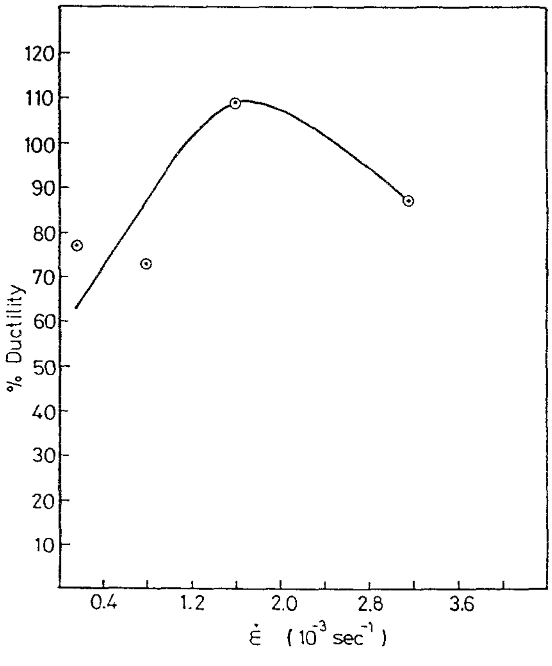

Figure 3 Variation of ductility with strain rate for 60/40 brass at 250° C.

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