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High temperature compressive behaviour of as-cast Al–Cu alloys of varying composition

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Al–Cu alloys with different copper contents ranging from 10 to 45 wt% were deformed in compression in the as-cast condition. Measurements of the strain-rate sensitivity parameter m using the crosshead speed jump technique show that alloys of composition close to the eutectic become superplastic after about 25% strain, with m increasing from about 0.2 at the beginning of deformation to about 0.5 or more, depending on composition. This transition to the superplastic state is accompanied by an important decrease of the flow stress during compression and it is associated by the breakdown of the initially lamellar structure of the eutectic phase in the highly deformed regions of the specimen. Moreover for the alloy with 37 wt% copper, the transition corresponds also to the degeneration of the dendritic primary phase and this alloy shows a particularly high value of m and a low steady state flow stress in spite of the large volume fraction of the hard $CuAl_2$ compound. Alloys of composition far away from the eutectic do not exhibit superplastic behaviour during compression, at least in the range of strain rate investigated and this is due to the large grain size and the important grain growth that occurs during deformation.

1. Introduction

Superplasticity is now becoming a very well known phenomenon that occurs in many materials with a fine and usually equiaxed structure. Many of them are eutectics in which the presence of phases generally of comparable volume fractions and different chemical compositions, inhibits grain growth. Without any special solidification conditions, the eutectic alloys have also the advantage of having an extremely fine as-cast structure which can be easily transformed into a fine equiaxed one by hot or cold working followed by recrystallization.

Recent studies on Al–Cu eutectic alloys, showed that this treatment can be realized by simply prestraining, preferably at high speed, (prior to superplastic deformation) the specimen, either in tension [1, 2] or in compression [3].

Several studies have been performed on the superplastic properties of the Al–Cu eutectic [4–10] and of some hypoeutectic and hypereutectic alloys [11, 12] having already a fine equiaxed structure. It was shown that alloys of a composition far from the eutectic do not exhibit superplastic behaviour [11] and this was attributed to excessive grain growth. Additions of small amounts of zirconium (0.5 wt%) to the Al–6 wt% Cu alloy, however, inhibit grain coarsening and make this industrially a well known superplastic alloy [13].

In the present work, as-cast Al–Cu alloys with different copper concentrations on both sides of the eutectic are used to investigate the structural changes and the corresponding evolution of the mechanical parameters in a compression test piece. This study is an extension of the previous investigation on the eutectic [3] and tests on specimens of this composition were also performed for comparison.

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2. Experimental procedure

Six aluminium–copper alloys with different copper contents (10, 17, 22, 33, 37 and 45 wt %) were prepared from eutectic (33 wt % Cu) and from either pure aluminium or copper, in the form of as-cast bars of 12 mm diameter solidified in a cylindrical copper mould. In this paper, the alloys will be designated only by a number corresponding to the nominal percentage of copper. Precise determinations of copper concentrations by chemical analysis gave however the following results, respectively: 9.80; 17.20; 22.05; 33.10; 36.90; and 45.4. In order to minimize the effect of different cooling rates, all the alloys were poured with the same superheat of about 100° C.

Compression specimens, 8 mm diameter × 8 mm high, were machined from the as-cast bars. Compression tests without lubrication were performed on an Instron testing machine at constant velocity using a compression unit specially designed for the machine. The tests were performed at 515° C using an electrical furnace, the temperature being controlled to within ± 1° C. After deformation, the specimens were water-quenched directly in the furnace, then sectioned, polished and etched for metallographic observations.

The same etching reagent (10% sodium hydroxide in water) was used for all the alloys investigated. Metallographic observations on specimens in the as-cast conditions were made as well.

Tests were performed either at constant velocity or using the crosshead speed jump technique for the determination of the strain-rate sensitivity parameter $m (= (\partial \log \sigma / \partial \log \dot{\epsilon})_T)$ according to the Backofen *et al.*'s method [14].

3. Experimental results

Fig. 1 shows optical micrographs of the structure of the alloys in the as-cast condition. For the hypoeutectic alloys (Figs. 1a, b and c) the dark regions are of eutectic composition and the light ones correspond to the Al–Cu solid solution supersaturated with copper (α-Al). Owing to the high cooling rates, the interlamellar spacing is less than 1 μm and the two phases of the eutectic cannot be optically resolved. For the hypereutectic alloys (Figs. 1e and f) the dark regions are also of eutectic composition and the light ones correspond now to the $CuAl_2$ intermetallic compound. The eutectic mixture in these alloys is coarser and is more or less degenerated. All the as-cast structures, except for the eutectic composition, show a dendritic characteristic of the primary phase, which is either the α-Al solid solution for the hypoeutectic alloys or the intermetallic compound for the hypereutectic ones. The eutectic structure (Fig. 1d) is of cellular type with approximately parallel lamellae in a cell and a high density of faults in the cell boundaries. The structure resolved by scanning electron microscopy observations show an interlamellar spacing of about 0.4 μm.

True stress against engineering strain is plotted on Fig. 2 for specimens of each alloy deformed in compression at a crosshead speed of 8.3×10^{-2} mm sec^{-1}. Stress was taken as the average pressure without correction for both frictional effects and strain rate increase during the constant velocity tests. Each curve, except that corresponding to alloy 10, shows a maximum stress at about 4% strain, after which the stress decreases to a comparatively lower value (steady state) which does not significantly change with strain. For alloy 45 however, this steady state value increases slightly with strain. The stress for alloy 10 does not show a peak – it increases continuously with deformation.

The effect of alloy composition on the peak stress and the steady state stress taken at 30% engineering strain is shown in Fig. 3. On the abscissa are shown the composition of the alloys in terms of the copper content and the volume fraction of the $CuAl_2$ intermetallic compound, as determined from the equilibrium diagram. It can be noticed from the curves that the ratio of the peak to the steady state stresses is maximum at 37%; henceforth it decreases very rapidly with further increase in the copper content. The same trend is also observed by decreasing the concentration from 37%, this ratio becoming smaller than 1 for a copper content less than about 12%. Alloy 45 deforms at excessively high steady state stress, compared to alloy 37, an increase of the copper content by only 8% leads to a stress increase from about 12 to about 32 MPa at 30% strain.

Fig. 4 shows the evolution of the strain-rate sensitivity parameter m with engineering strain for the six alloys investigated. For the determination of m, the speed jump technique was used in which specimens were deformed at an initial velocity of 3.3×10^{-3} mm sec^{-1}, and the velocity was then increased by a factor of 2.5 and finally decreased again to the initial value. The procedure was repeated up to 60% engineering strain.

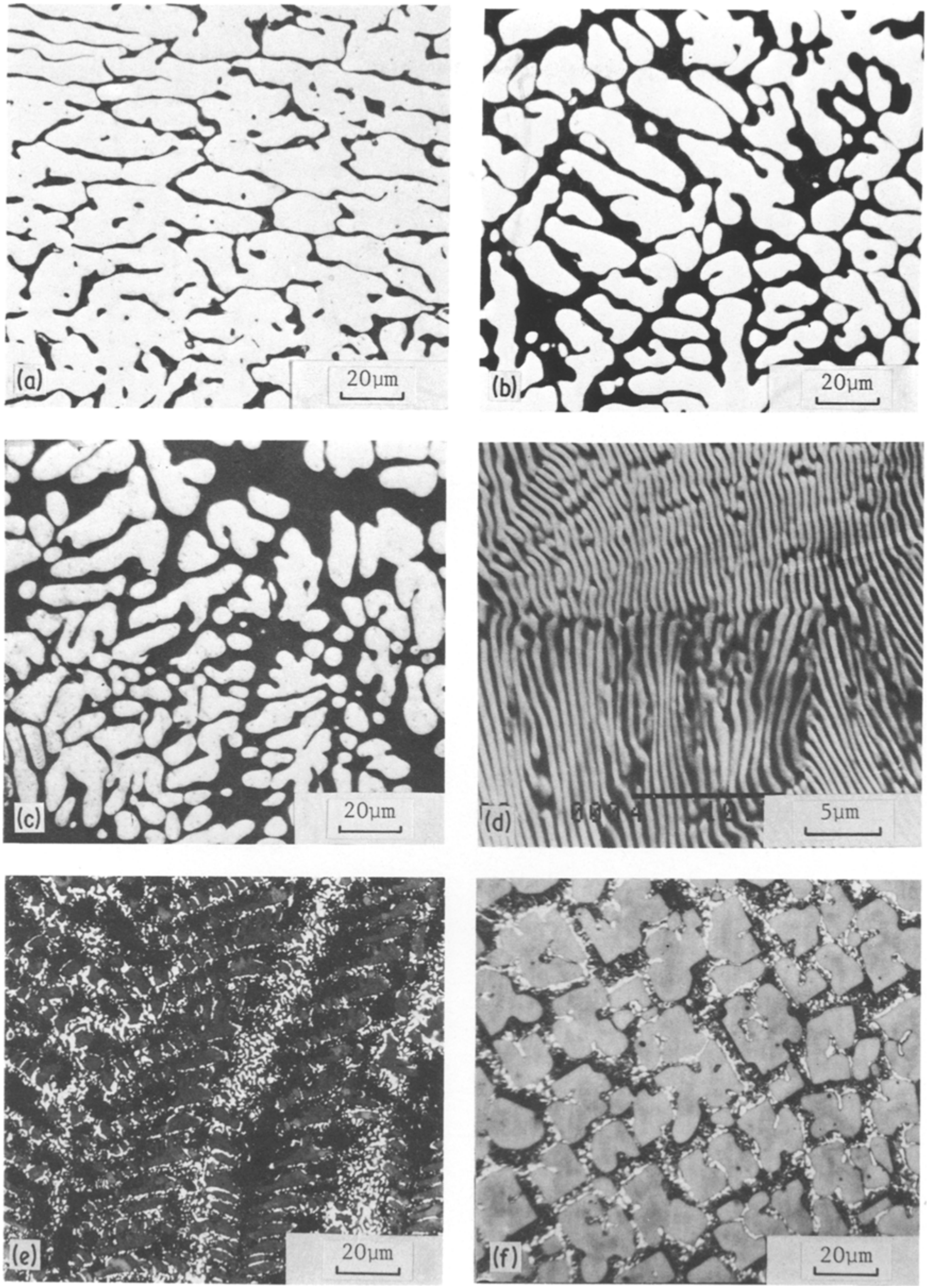

Figure 1 Micrographs showing the microstructure in longitudinal section of as-cast Al–Cu alloys with different copper content. (a), (b), (c), (d), (e) and (f), refer to alloys 10, 17, 22, 33, 37, 45, respectively.

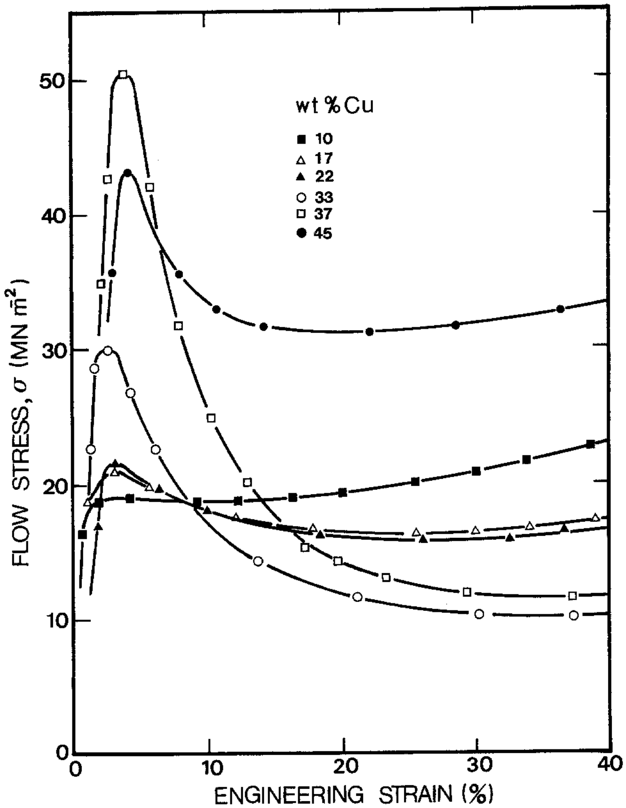

Figure 2 True stress against engineering strain for as-cast Al–Cu alloys with different copper content compressed at 515° C with a crosshead speed of 8.3×10^{-2} mm sec^{-1}.

The figure shows that for alloys with a Cu content greater than 22%, m increases from a low initial value of about 0.2 to 0.25, depending on the composition, to a steady state value characteristic of superplastic behaviour. For alloys 33 and 45, this increase occurs during the first 20% deformation, whereas for alloys 22 and 37, the steady state value is reached only after about 40% strain. It is to be noted that alloy 37 has a particularly high value of m which is close to 0.7. The evolution of m with strain for alloy 10 shows a different trend: it decreases slightly at the beginning of deformation towards a constant value after about 10% strain. For alloy 17, m increases slowly up to 0.35 in the steady state which is achieved after about 50%.

Results of the previous tests were used to correlate m to the copper content or the volume fraction of the $CuAl_2$ phase in the alloy. Fig. 5 shows this relation by means of two plots representing the m value in both the steady state and the as-cast condition. The latter values were determined by extrapolation of the curves of Fig. 4 to zero strain.

The change in behaviour of the as-cast materials during compression and the corresponding increase

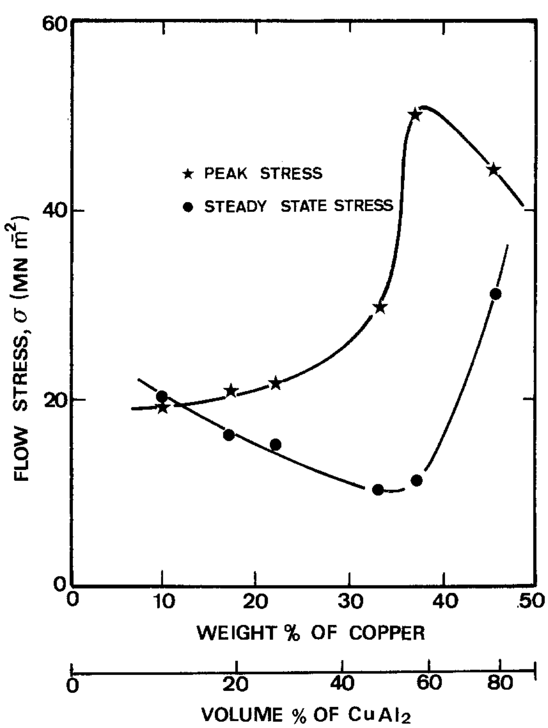

Figure 3 Effect of copper content on both the peak stress and the steady state stress (taken at 30% strain) for compression tests performed at 515° C with a crosshead speed of 8.3×10^{-2} mm sec^{-1}.

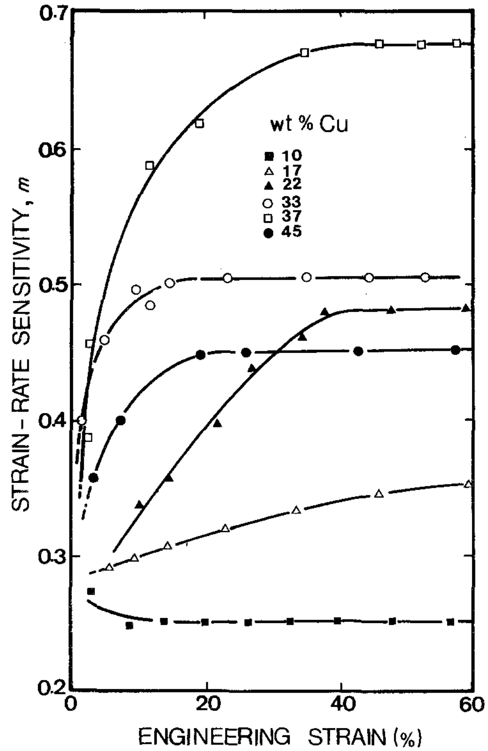

Figure 4 Evolution of the strain-rate sensitivity parameter (m) with deformation for as-cast Al–Cu alloys with different copper content (m measured by incremental speed-jump $3.3 \times 10^{-3} \rightleftarrows 8.3 \times 10^{-3}\,mm\,sec^{-1}$).

of m can also be expressed by plots of stress against strain rate both in the initial stage of deformation and in the superplastic region. For the former plot, specimens were deformed at different velocities, and the peak values of stress were recorded and plotted against the corresponding strain rate. For the latter, identical specimens were prestrained up to 60% at $8.3 \times 10^{-3}\,mm\,sec^{-1}$ and thereafter each of them was subjected to a different speed for which the steady state stress was determined.

Fig. 6 shows such plots for alloy 22. The maximum slopes of the curves are close to 0.3 and 0.45 for slightly deformed and highly deformed specimens, respectively. These values are in agreement with the results of Fig. 5. The figure shows that the higher the strain rate, the lower the ratio between peak stress and steady state stress. This situation was found to occur also on the eutectic and the explanation was given elsewhere [3].

Comparison of the $\log \sigma$–$\log \dot{\epsilon}$ relations for the hypoeutectic alloy 22 and the hypereutectic alloy 37 is shown in Fig. 7. The same procedure as described previously was used to obtain these plots. The curves have the classical sigmoidal shape with a higher m value for alloy 37 (close to 0.65) than for alloy 22 (close to 0.45). It is to be noticed that the curves intersect, which shows that at low strain rate, the stress for alloy 37 is lower than that for alloy 22 whereas the latter becomes softer than the former at a higher strain rate. The last result seems to be logical according to the higher volume fraction of the harder $CuAl_2$ phase in alloy 37. At low strain rates, however, the higher strength of alloy 22 might be attributed to the larger grain size owing to a more rapid grain growth of the softer phase.

Fig. 8 shows micrographs of the structure of deformed specimens for the alloys under study. These specimens were compressed at a constant crosshead speed of $8.3 \times 10^{-2}\,mm\,sec^{-1}$ up to 37% engineering strain, then quenched directly in the furnace and sectioned parallel to the stress axis. The micrographs were taken in the most severely deformed regions of the section. The structure consists of the two terminal phases: the light α-Al phase and the dark $CuAl_2$ phase.

Comparison with micrographs of Fig. 1 shows that an important evolution of the structure has

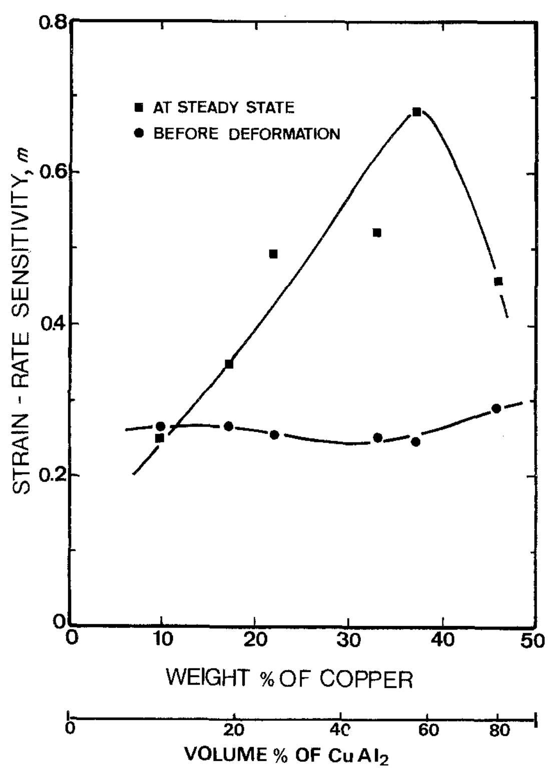

Figure 5 Effect of copper content on both the initial (in the as-cast) and the steady-state strain-rate sensitivity.

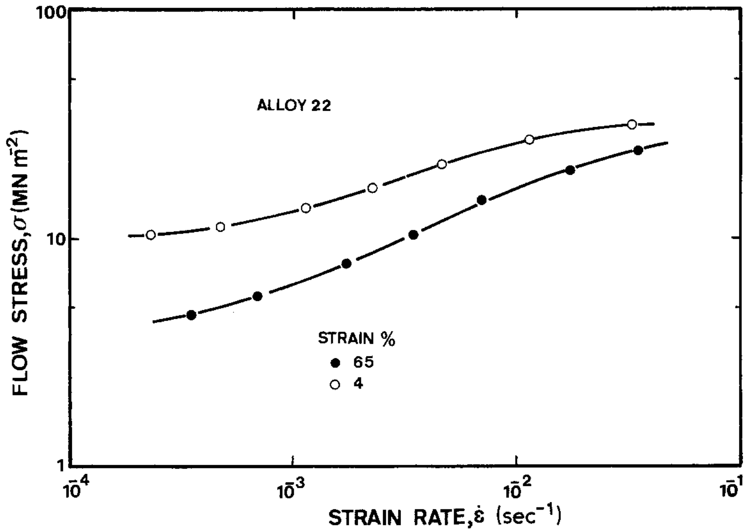

Figure 6 Log–log σ against $\dot{\epsilon}$ for alloy 22 after two different amounts of predeformation at 515° C with a crosshead speed of 8.3×10^{-3} mm sec^{-1}.

occurred during deformation. The eutectic mixture has degenerated: for the hypoeutectic alloys, the microstructure consists of grains of the $CuAl_2$ phase dispersed in the α-Al matrix (Figs. 8a, b and c). Both the eutectic and alloy 37 show a microduplex structure with phases of comparable volume fractions (Figs. 8d and e); it is to be remarked, however, that alloy 37 has a smaller phase size than that of the eutectic. In the case of alloy 45 (Fig. 8f), the predominant phase is the $CuAl_2$ compound; it is formed both by large grains corresponding to the primary phase and small ones resulting from the degeneration of the eutectic.

4. Discussion

The previously mentioned transformation of the structure during compression is associated with an important evolution of the mechanical properties. For alloys with compositions close to the eutectic, this evolution is accompanied by a decrease of

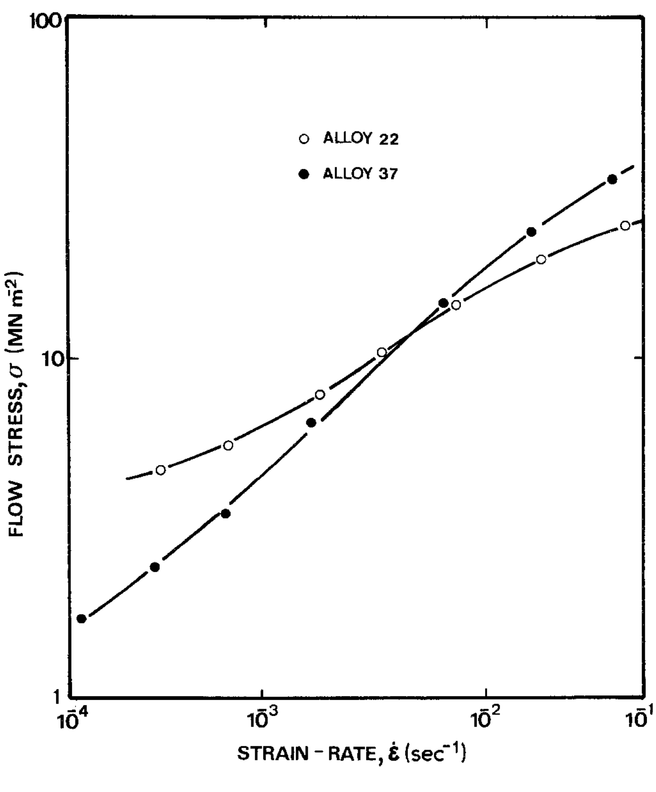

Figure 7 Log–log σ against $\dot{\epsilon}$ for alloys 22 and 37 after the same amount of predeformation (35%) at identical test conditions (515° C and 8.3×10^{-3} mm sec^{-1}).

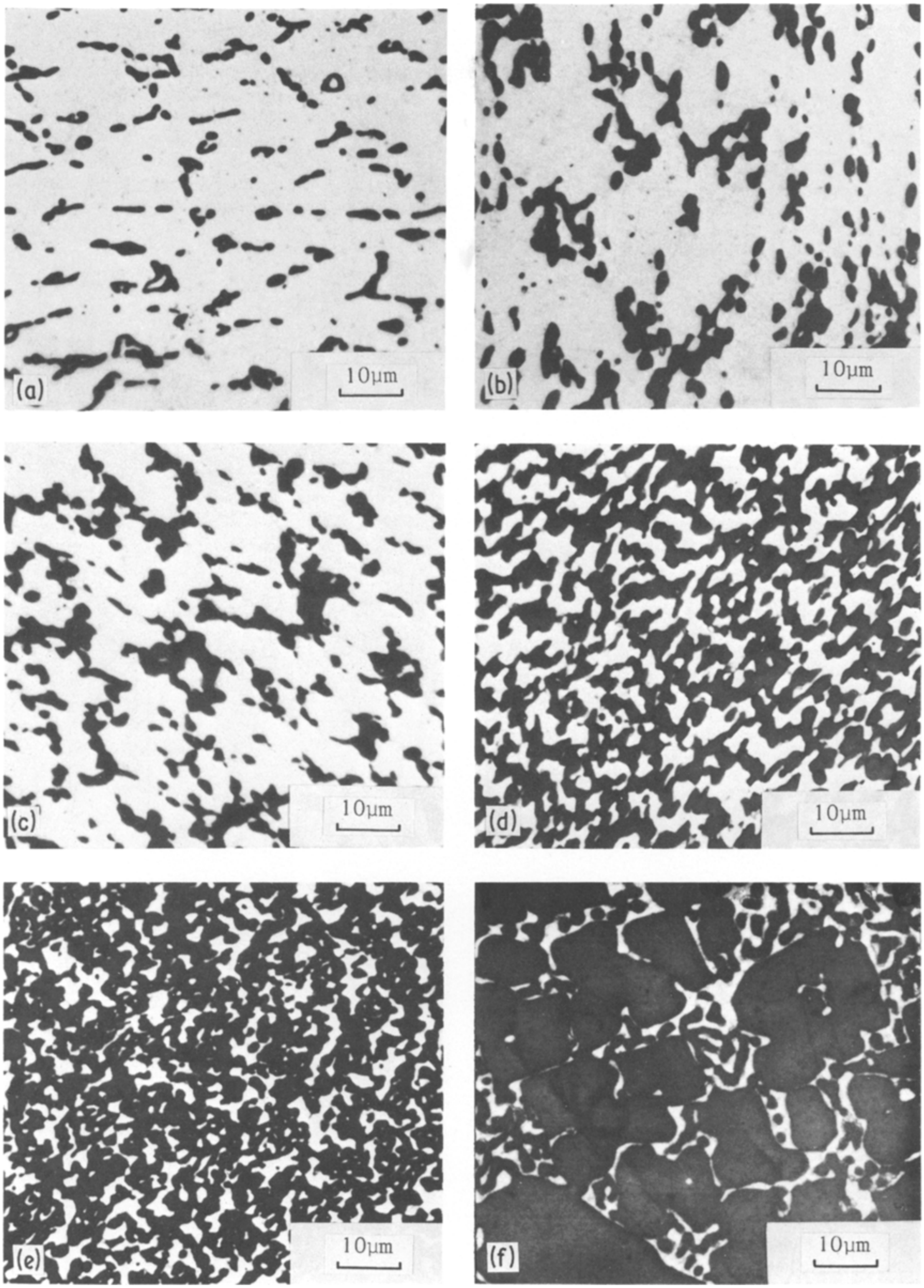

Figure 8 Optical micrographs showing microstructure in longitudinal sections of different alloys after 37% deformation at 515° C and with 8.3×10^{-2} mm sec^{-1}. (a), (b), (c), (d), (e) and (f), refer to alloys 10, 17, 22, 33, 37, 45, respectively.

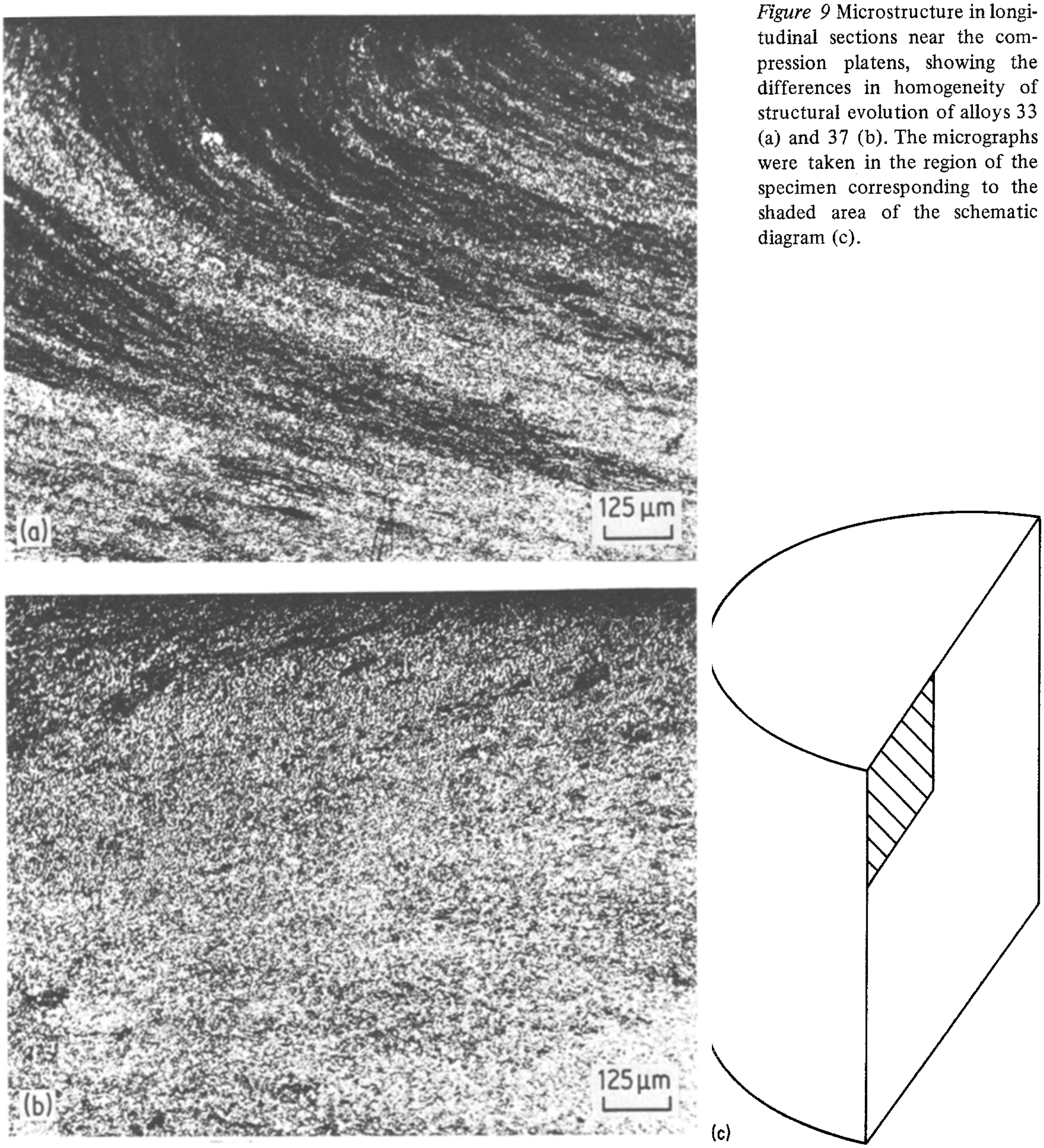

Figure 9 Microstructure in longitudinal sections near the compression platens, showing the differences in homogeneity of structural evolution of alloys 33 (a) and 37 (b). The micrographs were taken in the region of the specimen corresponding to the shaded area of the schematic diagram (c).

the flow stress at about 4% strain and a corresponding increase of the strain-rate sensitivity coefficient to values characteristic of superplastic behaviour. However, for the different alloys, marked differences are found in the magnitudes of the peak and steady state stresses, the values of the strain-rate sensitivity parameter and the strains necessary to achieve the superplastic conditions.

The peak stress in the early stages of deformation can be attributed to the presence of the eutectic mixture which has to deform in order for the lamellar structure to degenerate into an equiaxed one. This degeneration which occurs by the breakdown of the continuity of the phases needs stresses high enough to deform the hard intermetallic compound.

The peak stress phenomenon was already reported on the as-cast Al–Cu eutectic alloy by different authors [1–3] and also on other alloys both in the solid [15] and the semi-solid state [16] when the structure evolves from interconnected phases to more individualized ones. This explanation concerning the origin of the peak is confirmed by the absence of a peak in the stress–strain curve for alloys with low copper concentrations in which the amount of eutectic is small compared to that of the primary phase. Moreover the magnitude of the peak increases with an increase of the eutectic volume fraction. For hypereutectic compositions however, the peak stress continues to increase with the increase of copper content up to 37% and decreases thereafter. The very high peak observed for alloy 37 can be attributed to the superimposed effect of the presence of the primary dendritic $CuAl_2$ compound which has also to deform in order to transform the initial structure into a very fine equiaxed one (Fig. 8). For alloy 45, the decrease of the peak stress is due to the predominant effect of the decreasing volume fraction of the eutectic. Indeed in this alloy, the primary $CuAl_2$ compound seems to be only slightly deformed as shown by comparison of the structure before and after compression, deformation being concentrated in the interdendritic eutectic mixture.

Both alloys 33 and 37 show a very low steady state stress (about 10 and 12 MPa, respectively, for a strain rate of $1.6 \times 10^{-2} \sec^{-1}$) due to the extremely fine structure being developed during the initial stages of deformation. In spite of the structure of alloy 37 being finer than that of the eutectic, the steady state stress is higher for the former and this is explained by the larger volume fraction of the intermetallic phase. It is to be noted that the structure of alloy 37 shows two different ranges of grain sizes for the intermetallic compound; the smaller grains resulting from the eutectic degeneration and the larger ones being developed through the evolution of the primary dendrites.

This evolution, however, did not seem to take place in alloy 45, as deduced by comparison of the structure before and after deformation (Fig. 1 and 8) and the reason for this is not obvious. In this alloy, the presence of large grains of the $CuAl_2$ phase restricts deformation to the eutectic region leading to high local strain rates and thus high steady state stress. Moreover this localization of deformation results in strain-induced grain growth which is probably the reason of the stress increase during deformation.

Stress increase in alloy 10 is also explained by grain coarsening which is in this case due to the low volume fraction of the intermetallic compound. Such an effect indeed decreases continuously with an increase of copper content and it completely vanishes for alloys of nearly eutectic composition.

It is necessary here to remember that the curves in Fig. 2 correspond to the average pressure without correction for frictional effects and strain rate increase during the tests. If the curves are corrected for these effects, the real stress decrease after the peak would be more important and the plots would show a continuous decrease or a lower increase in stress with increasing strain. Indeed for alloy 10, because of the low m value, (Fig. 4), the recorded stress is not significantly affected by the strain rate increase during the constant velocity test.

The value of m for the alloys in the as-cast condition is very low (0.23) and it is not significantly affected by the composition. This value is characteristic of non-superplastic behaviour and is in agreement with previous results concerning deformation of as-cast materials [6] and materials of similar structure in welded joints [17].

The m value for alloy 10 remains very low during deformation being close to the initial one, in spite of the important structural evolution: indeed after deformation, the structure consists of fine particles of the $CuAl_2$ phase more or less uniformly dispersed in the α-Al matrix. This structure does not lead, however, to superplastic behaviour for the strain rate of the tests and this is probably due to the large matrix grain size which increases during deformation as a consequence of agglomeration of the $CuAl_2$ particles.

This agglomeration phenomenon also occurs in the other alloys, and this observation is similar to those already reported [12, 18]. For alloys containing comparable volume fractions of phases, however, (close to the eutectic composition) the agglomeration also leads to a change in grain size but it still lies in the range of superplasticity for the strain rate of the tests.

The steady state value of m for alloy 17 is about the same as that previously observed for alloy in the initially equiaxed condition [11, 12] and it was found to be associated with relatively high elongations of about 300% [11].

Alloys 22, 33, 37 and 45 exhibit high steady state values of m ranging from about 0.45 to 0.68. For alloys 33 and 45, this steady state is achieved after only 20% deformation while for the other two alloys, higher strains (about 40%) are required. This difference in behaviour can be attributed to the different degree of localization of deformation: alloys 33 and 45 indeed show very localized deformation with two different aspects; in the eutectic it occurs heterogeneously at the intercell boundaries in the zones of maximum shear as already observed [3]; in alloy 45, the localization occurs in the eutectic mixture uniformly distributed in between the coarse undeformed dendrites of the $CuAl_2$ primary phase, which represents a large volume fraction of the structure. Conversely for the other two alloys, deformation shows no localization, the structure being completely transformed throughout the whole volume of the specimen. This difference in homogeneity of structural evolution for alloys 33 and 37 is clearly seen in Fig. 9 which shows typical micrographs of longitudinal sections near the compression plates of specimens deformed under identical conditions (40% strain at 8.3×10^{-2} mm sec^{-1}). Because of localization of deformation, the macroscopic strain necessary to achieve steady state value of m for alloys 33 and 45 is lower than that for alloys 22 and 37. The differences in magnitude of steady state values of m for the four above mentioned alloys tested under identical conditions can be attributed to different factors: grain size in the transformed regions of the specimens, high local strain rates in these regions and volume fraction of each phase. Each of these factors can have incremental or decremental effects on the measured value of m, so that it is difficult to predict the actual magnitude of m in each case. These measured values which are indicative of superplasticity are to be considered as representative of the global transformed and non-transformed structure of the specimens.

5. Conclusion

The compressive behaviour of as-cast Al–Cu alloys with different copper contents was investigated under conditions of superplasticity and the main results are as follows:

1. Alloys of composition close to the eutectic become superplastic during compression. This transition manifests itself by an important decrease of the flow stress and a corresponding increase of the strain-rate sensitivity coefficient.

2. This superplastic characteristic is achieved by the different alloys after different amount of compression, depending on the degree of homogeneity of the structural evolution during deformation.

3. Alloys with composition far from the eutectic do not behave superplastically as deduced from measurements of the strain rate sensitivity parameter and this can be attributed to the large grain size obtained in these alloys.

4. Observations of different regions of compressed samples show that the transformation of the as-cast structure is highly homogeneous except for the eutectic alloy, in which the transformation is concentrated in shear bands inclined to the compression axis.

5. The behaviour of the alloy containing 37 wt % Cu is of particular interest from an engineering point of view: the transition from the as-cast condition to the superplastic state being associated with homogeneous transformation of the structure into an equiaxed one, high strain-rate sensitivity and low flow stress.

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References

1. S. HORI and N. FURUSHIRO, Technol. Rep., Osaka University No. 1097 August 5, (1972) 75.
2. M. W. A. BRIGHT, D. M. R. TAPLIN and K. W. KERR, *J. Eng. Mater. Technol.* 97 (1975) 1.
3. G. A. NASSEF, M. SUERY and A. EL-ASHRAM, *Met. Technol.* 9 (1982) 355.
4. M. J. STOWELL, J. L. ROBERTSON and B. M. WATTS. *Met. Sci. J.* 3 (1969) 41.
5. B. M. WATTS, M. J. STOWELL and D. M. COTTINGHAM, *J. Mater. Sci.* 6 (1971) 228.
6. D. L. HOLT and W. A. BACKOFEN, *Trans. ASM* 59 (1966) 755.
7. C. P. CUTLER, J. W. EDINGTON, J. S. KALLEND and K. N. MELTON, *Acta. Metall.* 22 (1974) 665.
8. R. D. S. WHITLEY, *Z. Metallkd.* 64 (1973) 522.
9. S. HORI and N. FURUSHIRO, *Bull. J. Inst. Met.* 14 (1975) 673.
10. G. RAI and N. J. GRANT. *Metall. Trans.* 6A (1975) 385.
11. S. HORI, N. FURUSHIRO and S. KAWAGUCHI, in Proceedings of 19th Japan Congress on Materials Research, Japan, 1976 (Society for Materials Science, Japan, Kyoto, 1976).
12. J. R. CAHOON, *Met. Sci. J.* 9 (1975) 346.
13. K. MATSUKI, K. MINAMI, M. TOKIZAWA and Y. MURAKAMI, *ibid.* 13 (1979) 619.
14. W. A. BACKOFEN, L. R. TURNER and D. H. AVERY, *Trans. ASM Q.* 57 (1964) 980.
15. M. SUERY and B. BAUDELET, *Rev. Phys. Appl.* 13 (1978) 53.
16. M. SUERY and M. C. FLEMINGS, *Met. Trans.* 13A (1982) 1809.
17. C. HOMER, J. P. LECHTEN and B. BAUDELET, *Metall. Trans.* 8A (1977) 119.
18. K. A. PADMANABHAN, *Trans. Ind. Inst. Met.* 26 (1973) 41.

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