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Abstract

A TEM study has been made of the effect of varying the magnitude of the applied stretch upon the distribution of S-phase after subsequent ageing of an 8090 Al-Li-Cu-Mg-Zr alloy. The S-phase has also been varied without stretching by means of a duplex ageing treatment. A comparison has been made of the yield stress and initial work hardening rates in tensile tests at temperatures between 21°C and 277°C of each of the microstructures produced.

Dynamic recovery occurs at the lowest temperature in material subjected to the highest degree of prior stretch, and at the highest temperature in unstretched material. It is concluded that the dislocation substructure introduced during the stretch acts as an effective dislocation sink during high temperature deformation of the aged alloys. The duplex-aged material thus exhibits a high resistance to dynamic recovery. The work-hardening rates observed at the highest temperatures appear to be dependent upon the size and amount of the S-phase present.

Introduction

The present investigation is concerned with an Al-Li-Mg-Cu-Zr alloy designated 8090. The main strengthening phase is δ' (Al_3Li) which, being a coherent, ordered phase encourages planar, inhomogeneous slip which may lead to premature failure. Cu and Mg are added to promote the formation of S-phase precipitates (Al_2CuMg) which not only contributes to the strength but also has the effect of homogenising the distribution of slip.

S-phase nucleates preferentially on stray dislocations and upon subgrain boundaries, and a more uniform distribution of the precipitate particles is produced if the solution-treated material is plastically stretched prior to ageing. The uniform dislocation networks generated act as nucleation sites for the precipitation of S-phase particles. Previous work (1) has shown that the dynamic recovery characteristics of 8090 are influenced by these microstructural changes, although it was not possible to determine whether the S-phase particles themselves or the dislocation network upon which they precipitated were responsible for the changes observed. The object of the present research is to study in a systematic way the effect of the degree of prior stretch upon the S-phase distribution and upon the elevated temperature tensile properties. A comparison has also been made with the properties of an alloy in which the S-phase distribution has been changed by means of a duplex heat-treatment (2) not involving a stretching operation.

Experimental Procedure

Material

The composition of the alloy is given in Table 1. The material was in the form of rolled plate of 25mm thickness.

Table 1: Composition of 8090 alloy

Element:	Li	Cu	Mg	Zr	Fe	Si	Na	Al
Wt. %	2.41	1.16	0.61	0.11	0.14	0.10	0.0015	bal.

Test-pieces

A series of specimen blanks were solution-treated at 530°C for 1hr, cold-water quenched and subjected to a plastic stretch of either 4% or 7%, and then aged to peak hardness (20 hr at 190°C). A second series of blanks were subjected to the following heat-treatment (2): solution treatment at 530°C for 1hr followed by a water quench, natural ageing for 24 hr and a final artificial age of 24 hr at 190°C.

Cylindrical tensile specimens of gauge length 18mm and diameter 5mm were machined from the blanks, such that their longitudinal axis was parallel to the rolling direction of the sheet. Tensile tests were conducted over a range of temperatures between 21°C and 277°C at a strain rate of 1 x 10^{-3} s^{-1}. Specimens were maintained at test temperature of 0.5 hr prior to commencement of testing. Work hardening rates were calculated from the gradient of the load/extension curve after 0.2% permanent strain.

Results

Fig.1 shows the variation in modulus-normalised 0.2% proof stress (PS) with temperature for the material in the various conditions.

Fig.2 shows the variation in modulus-normalised work hardening rate (whr) with temperature. The dotted line in figs.1 and 2 refer to the behaviour of unstretched (N) singly aged material, and the data are taken from Reference (1).

The distribution of S-phase in the various specimens was studied in the TEM, and figs.3 - 6 show the structures observed under either bright field or dark field conditions. Fig.3 (taken from Reference 1) shows the S-phase distribution in (N) material, and it is apparent that the distribution is very heterogeneous, the phase being primarily decorating the stray dislocations present. In the duplex-aged (D) material (Fig. 4) there is a fine, uniform array of S-phase , as has been reported in the literature (2), and the stray dislocations present have also been decorated with S-phase particles as is observed in the N material. In the 4% (Fig. 5) and 7% (Fig. 6) deformed material, the effect of stretching has also been to produce a uniform array of S-phase, this time in association with the dislocations introduced by the stretch.

It is seen that the particles of S-phase are acicular in habit, and the average length of the particles after the various treatments has been measured (fig. 7). The S-phase particle size is seen to be somewhat similar in the unstretched materials (N and D), and progressively increases as the degree of stretch is increased from 4% to 7%.

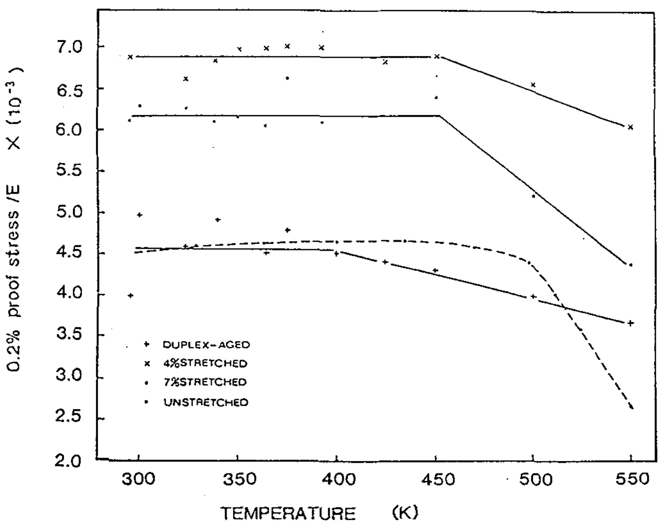


Fig.1. The variation in modulus-normalised 0.2% proof stress (PS) with temperature for the material in the various conditions.

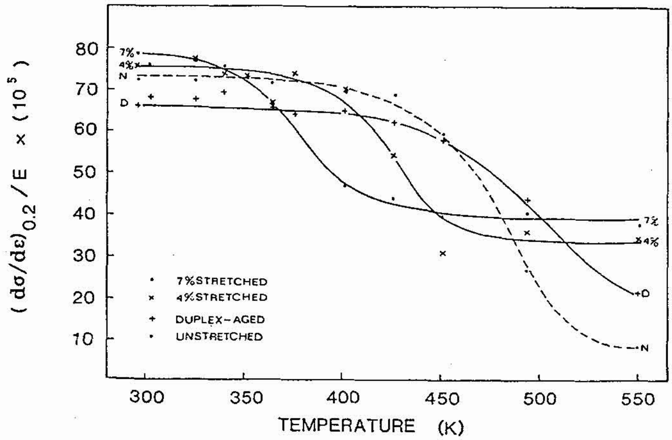


Fig.2. The variation in modulus-normalised work-hardening rate with temperature.

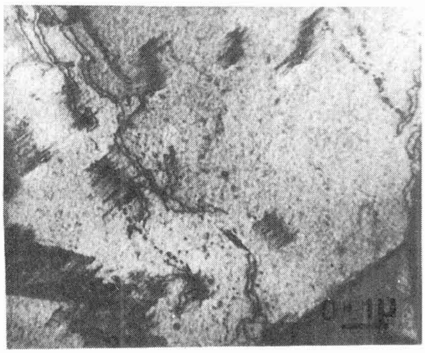
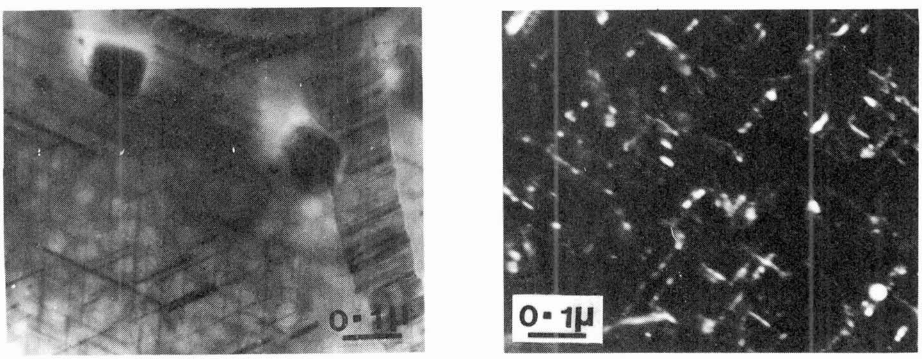
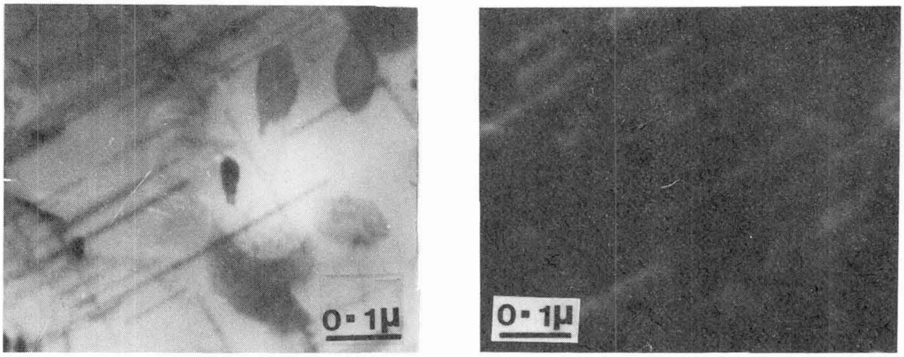


Fig.3. Showing the S-phase in unstretched, single-aged (N) material being principally associated with stray dislocations in the matrix. (from Reference 1)



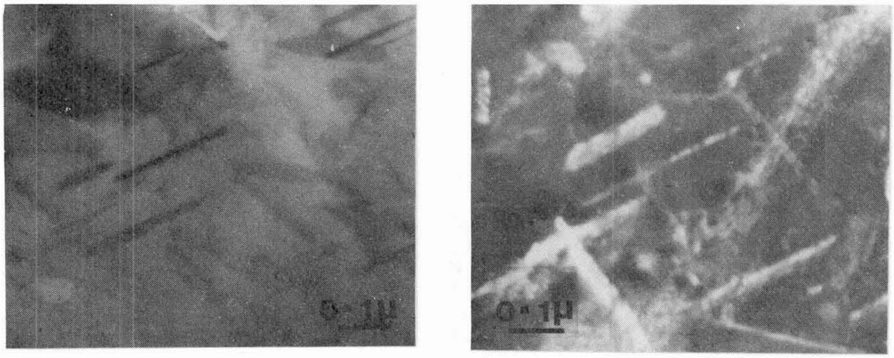
(a) (b)

Fig.4. Showing the S-phase in (a) bright field and (b) in dark field contrast in unstretched, duplex-aged (D) material.



(a) (b)

Fig.5. Showing the S-phase in (a) bright field and (b) dark field contrast in material given 4% stretch prior to peak ageing.



(a) (b)

Fig.6. Showing the S-phase in (a) bright field and (b) dark field contrast in material given 7% stretch prior to peak ageing.

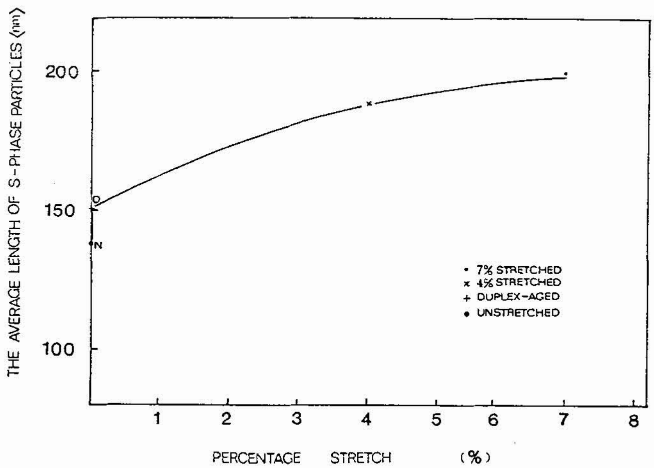


Fig.7. The average length of the S-phase particles in the various specimens.

Discussion

The distribution of S-phase

In the unstretched (N) material, the intragranular precipitation of S-phase is limited to the decoration of stray dislocations. The particles form serrated sheets of interconnected laths (Fig. 3), and a relatively low volume fraction of the phase appears to have formed.

In the stretched material a much more uniform array of S-phase particles is observed. The application of increasing stretch appears to produce an increasing size of S-phase particle , and it may be inferred that the dislocations introduced by the stretch are also acting as pipe diffusion channels for Mg and Cu, so that enhanced particle growth rates have determined the particle size.

The material which has been given the duplex ageing treatment (2) are seen to contain S-phase in two distributions - decoration of the stray dislocations as in the N materials, and also a uniform array of S-phase particles throughout the grains which have presumably nucleated at solute-vacancy clusters formed during the period of natural ageing after the quench. The low dislocation density has not therefore provided easy diffusion paths for particle coarsening, so that (fig.7) the average size of the S-phase particles is significantly less than that observed after 4% or 7% stretching.

The variation in yield stress with temperature.

Apart from some scatter in the data for the D material, the modulus-normalised yield stress of the specimens is constant up to about 450K, This implies a constant mechanism of yield, presumably the glide of dislocations by the operation of dislocation sources. The fall in yield stress at temperatures above 450K presumably reflects the onset of processes controlled by dislocation climb.

Contributions to the observed yield stress will arise from the δ'-phase, (this will be comparable in each of the specimens), the S-phase and the dislocation substructure introduced by the stretch. N material has precipitated only a low volume fraction of S-phase, so part of its yield strength will arise from the Cu and Mg still in solution. It has a low yield stress due to the absence of a dislocation substructure. In D, the Cu and Mg have precipitated as S-phase, with little net change in the yield stress.

In considering the stretched samples, an increase in yield stress is apparent, but it is notable that the 7% stretch results in a structure of lower yield stress than a 4% stretch. This presumably is because the softening arising from the increase in size of the S-phase particles (and thus coarser dispersion) outweighs the contribution to the strength from the dislocation substructure.

Temperature variation of the Work Hardening Rate.

At low temperature, the observed work hardening rate will arise from the accumulation of the statistically stored dislocations and the geometrically necessary dislocations. The latter will arise mainly from the presence of the (unsheared) particles of S-phase in the structure. Clearly the dislocation substructure introduced by the stretch remains, after ageing, a potent source of work hardening, in that the highest work hardening rate is observed in the material subjected to the highest degree of stretch (7%).

With increase in temperature, however, it appears that this same dislocation substructure begins to act as a dislocation sink, so that the work hardening rate begins to fall first in the 7% stretched material, whereas the D and N material (into which no dislocation substructure had been introduced) exhibit the greatest resistance to dynamic recovery. The 4% stretched material shows dynamic recovery characteristics intermediate between the 7% stretched and the unstretched materials.

The progressive fall in whr observed between 350°C and 550°C is therefore considered to arise from the climb of dislocations into the grain and subgrain boundaries. The differences in whr between the groups of specimens observed at the highest temperature is considered to arise from the differences in S-phase distribution, and it may be noted that the observed order of work hardening rates is the same as that of the particle size (Fig. 7). There will be differences in the distribution and behaviour of the geometrically necessary dislocations associated with the S-phase particles, so that N material, with the lowest volume fraction and smallest particles of S-phase thus exhibits the lowest whr at this temperature.

Summary and Conclusions

1. Solution-treated material should be subjected to a plastic stretch or to a duplex ageing treatment in order increase the volume fraction of S-phase precipitated.

2. The average particle-size of the S-phase increases as the percentage stretch increases up to 7%. This is attributed to particle coarsening by pipe diffusion of solute atoms along the dislocations introduced by the stretch.

3. The highest yield stress in the peak-aged alloy is observed in that subjected to 4% stretch. This is attributed to the particle-size of the S-phase being finer than that given a 7% stretch.

4. Dynamic recovery, as measured by the work-hardening rate at a strain of 0.2%, occurs at the lowest temperature in the material subjected to the highest degree of prior stretch. Maximum resistance to dynamic recovery is thus exhibited by the unstretched materials, N and D. It is concluded that the work hardening rate falls because the dislocation substructure begins to act as a dislocation sink.

5. The work-hardening rates at the highest temperatures studied are a maximum in those materials with the highest volume fraction and largest particles of S-phase.

Acknowledgements

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