THE EFFECT OF PRIOR STRETCHING ON THE ELEVATED TEMPERATURE TENSILE BEHAVIOUR OF THE ALUMINIUM-LITHIUM ALLOY 8090

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Abstract

Specimens of 8090-T8 alloy were stretched by 3% and 7% prior to ageing at 190°C. The tensile and proof strengths were increased for the 7% stretched alloy at low temperatures, but at above 210°C mechanical properties were independent of the extent of prior stretching. Engineering stress - engineering strain curves showed apparent work-softening at high temperatures, but this was due to a combination of a low work-hardening rate and an increased tensile strain in the necked region at these temperatures.

Introduction

An important property of a potential aerospace aluminium alloy is the ability to retain strength at temperatures of 100-200°C, either for sudden temperature excursions or prolonged soaking at elevated temperature. Early work by Pridham, Noble and Harris (1) showed that strength in the Al-Li-Cu-Mg alloy 8090 was retained up to 150°C, both for the stretched (3%, -T8) and unstretched (-T6) alloys. While the stretched alloy was stronger at temperatures below 150°C, at higher temperatures there were no strength differences between the two forms of the alloy. Xia and Martin (2) showed how the high temperature tensile behaviour of peak aged 8090 was affected by the extent of pre-ageing stretch applied to the alloy. They showed that the mechanical properties, especially the modulus-normalised work-hardening rate, of an alloy stretched by 7% began to fall at a lower temperature than those of an alloy stretched by 4%. This they ascribed to dynamic recovery, which was accelerated by the more extensive dislocation substructure in the 7% stretched alloy.

Thus alloys which have been substantially stretched may have higher strengths at room temperature but may not retain these advantages at higher temperatures. Thermomechanical treatments, such as the increased stretch described by Miller et al (3), which give good results at room temperature may not be as suitable for components which are likely to be aerodynamically heated.

Current practice with 8090-T62 sheet (4) shows that strength is retained to 150°C after a 10 hour soak, and to 120°C after a 1000 hour soak, this showing superior strength retention to that exhibited by 2014-T6.

The aim of the present work is to examine the effect of 3% and 7% stretch on the tensile properties of the Al-Li-Cu-Mg 8090 alloy in the temperature range up to 250°C.

Experimental

The material was in the form of 6 mm plate and had a composition of (weight per cent) Li 2.36, Cu 1.08, Mg 0.67, Zr 0.11, Fe 0.05, Si 0.03, balance Al. Solution treatment was for 1 hour at 520°C, followed by water quenching. The specimens, immediately after solution treatment, were either stretched 3% or 7% prior to age hardening for 20 hours at 190°C.

Tensile testing at temperatures between 25°C and 240°C was carried out on cylindrical specimens of 3 mm diameter and 13 mm gauge length at an initial strain rate of $2.56 \times 10^{-4}\ s^{-1}$. Specimens were soaked in the furnace for 15 minutes before testing. All specimens were machined with their tensile axis parallel to the direction of rolling of the plate.

Additional tensile tests were carried out on flat specimens, 6.5 mm wide and 3 mm thick, with a gauge length of 25 mm and an initial strain rate of $6.67 \times 10^{-4}\ s^{-1}$.

Fracture surfaces were examined in a JEOL JSM680 scanning electron microscope at 25 kV.

Transmission electron microscopy was carried out on discs, cut from areas remote from the necked region of the tensile testpieces and with the foil normal parallel to the tensile axis. Thin foils were produced by electrolytic jet polishing using a nitric acid-methanol mixture at -15°C. The discs were then examined in a JEOL JEM-100CX microscope at 100 kV.

Results

Results of tensile testing of the round specimens are shown in Table 1 and schematic engineering stress - engineering strain curves are illustrated for the 150-210°C temperature range in Fig. 1. These show that the increased stretch raises the tensile and proof strengths of the alloy at room temperature, while at temperatures of 210°C and above, both the tensile and proof strengths of the two alloys are practically identical.

Fig. 1 shows that the general shape of the engineering stress - engineering strain curve changes with temperature. Taking the 3% stretched case, the rate of work hardening drops as the temperature rises. At a test temperature of 210°C, the curve has a shape characterised by a near-identical proof and tensile strength and an apparent work-softening stage. For the 7% stretched case the tensile test curve at 180°C shows the same general shape as the 210°C curve for the 3% stretched sample.

Since, in spite of the closeness of the tensile and proof strengths, the high temperature tensile specimens showed good ductility, additional tests using flat specimens were carried out to determine the extent of uniform elongation. These results are shown in Table II and indicate that the uniform elongation does not significantly increase with increasing testing temperature.

Table I Tensile Properties in Temperature Range 25-240°C

Test Temp (°C)	3% stretch			7% stretch		
	Tensile Strength (MPa)	0.2% Proof Strength (MPa)	Ductility (%)	Tensile Strength (MPa)	0.2% Proof Strength (MPa)	Ductility (%)
25	442	365	8	503	419	10
100	428	361	18	434	414	14
150	384	353	26	392	385	26
180	354	324	20	365	355	29
210	321	310	22	329	311	23
240	315	300	28	310	290	21

Table II Results of Additional Tensile Testing

Test Temp (°C)	3% stretch			7% stretch		
	Tensile Strength (MPa)	0.2% Proof Strength (MPa)	Ductility* (%)	Tensile Strength (MPa)	0.2% Proof Strength (MPa)	Ductility* (%)
160				376	369	19 (11)
190	363	354	23 (13)			
200				346	344	25 (10)
250	220	216	30 (15)			

* Uniform ductility in parentheses

Fractographs are shown in Fig.2 and indicate the transition from the characteristic layered structure in the room temperature tested sample to a more homogeneous dimpled structure for those tested at 240°C.

Transmission electron microscopy of foils taken from the specimens tensile tested at 210-240°C showed a small increase in the diameter of the δ' particles but no evidence of dislocation pairs cutting the ordered δ' particles.

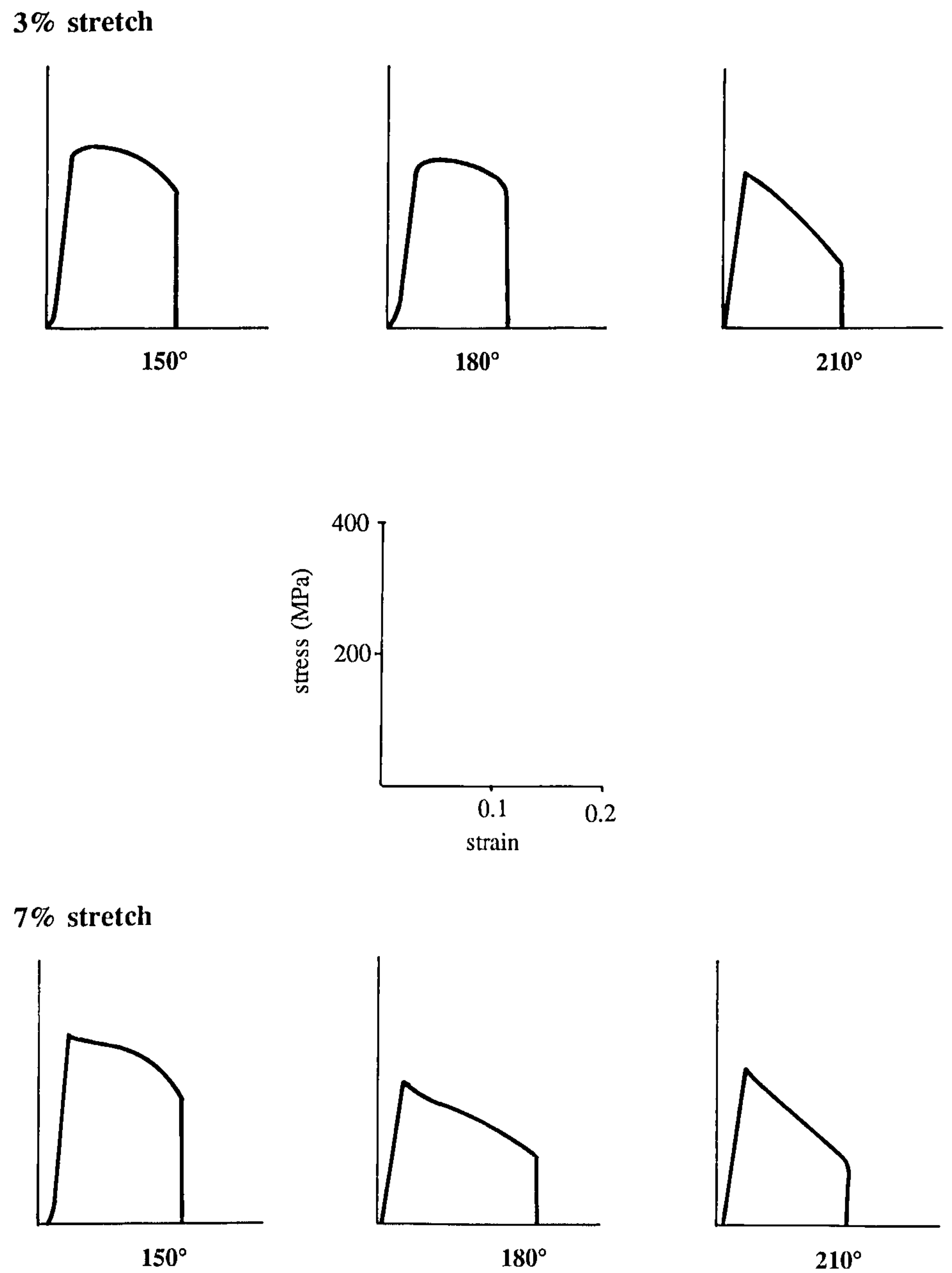
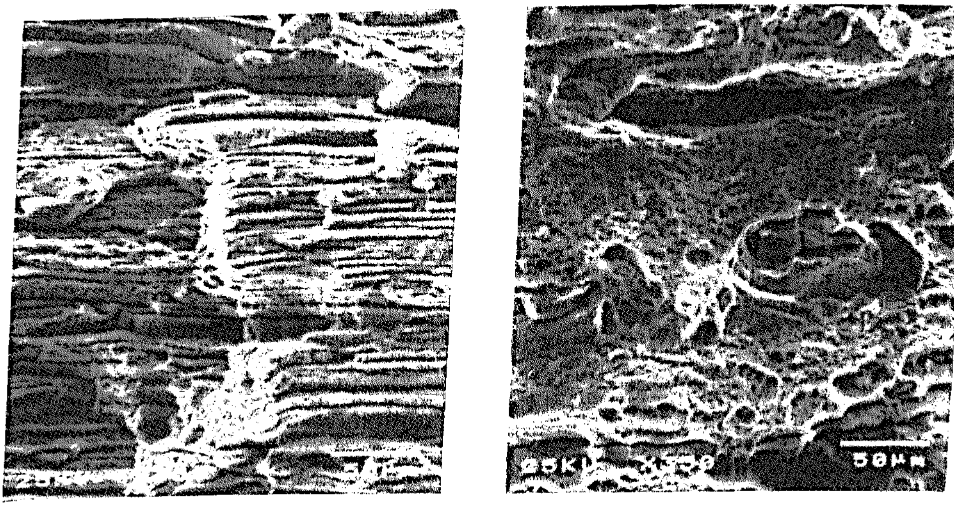


Figure 1 Engineering stress - engineering strain curves (schematic) for 8090-T8 stretched 3% and 7% before ageing in the temperature range 150-210°C



a) b)

Figure 2 3% stretched samples tested at a) room temperature and b) 240°C

Discussion

In alloys hardened by particles that are by-passed by dislocations, the work hardening rate will tend to decrease with increasing temperature, as the extra dislocations introduced by work hardening are increasingly removed by temperature-sensitive processes such as climb and cross-slip. For 8090 it will be the S (Al_2CuMg) particles that will be by-passed, thus homogenising the slip. Damerval, Lapasset and Kubin (5) showed that the work-hardening rate of 8090 in the underaged condition (6 h/170°C) reduced substantially, as the temperature rose from -196°C to 160°C. This confirmed the results of Stewart and Martin (6) on a model single crystal of Al with 0.31 volume per cent Si particles, which indicated a sharp drop in work hardening rate in the temperature range of 80-180°C and an absence of work hardening at 330°C.

However none of these describe conditions where the alloy work-softens. In precipitation hardened alloys localised work-softening occurs when precipitates are cut rather than by-passed (7), producing heterogeneous slip. This could be shown in the stress-strain curves of peak-aged binary Al-Li single crystals by the presence of serrations (8), although other mechanisms have been indicated as explaining serrations more generally in Al-Li alloys (9-11). Binary Al-Li alloys (12) show concentration of slip on to few slip planes (heterogeneous slip), which is the reason for the introduction of ternary and quaternary elements to commercial alloy compositions. In the case of 8090 this work softening could only occur in the absence of S particles and would be shown by an increased number of dislocations cutting the δ' particles, something that does not happen in the temperature range of this work. Thus the apparent work-softening shown in the engineering stress - engineering strain curves (Fig.1) must have a different origin. Table II indicates that the increased ductility is confined to the necked area of the testpiece, since the extent of uniform elongation hardly changes as the temperature rises from 190°C to 250°C for the 3% stretched specimens. For a material with a low value of the strain-rate sensitivity index m, fracture follows rapidly when a neck forms and there is little additional specimen extension. Large extensions in the necked area are associated with higher values of m, leading to the phenomenon of superplasticity when m > 0.4 (13). For materials below their temperature range for superplasticity, such as 8090 at the temperatures of the present work, m will rise with increasing temperature (14) and this will be normally associated with an increased tensile elongation (15). Thus the decreasing load seen as the specimen plastically deforms in Fig.1 is associated with the slow formation of a neck during tensile testing at temperatures where m will be higher and there is little or no work hardening. This confirms the increased ductility of 8090-T852 forgings, tensile tested at temperatures above 200°C, seen by Wan, Smallen and Carter (16), even though the tensile and proof strengths converged at these higher temperatures.

The effects of the different levels of stretch can be summarised as an increased proof strength (Table I) and the change to an apparent work-softening stress-strain curve occurring at a lower temperature (Fig.1) for the 7% stretched alloy. The latter would be consistent with an increased rate of dynamic recovery in the alloy which had a greater initial dislocation density. Xia and Martin (2) showed that the rate of dynamic recovery was increased by the increased stretch, as shown by the drop in the modulus normalised work-hardening rate occurring at a lower temperature for the alloy with the greater stretch. Later work by Xia and Martin (17) indicated that this behaviour applied for 8090, but did not apply for 8091, where the temperature at which the modulus-normalised work-hardening rate decreased with temperature was independent of the extent of stretching. They ascribed this to the presence of fully mobile dislocations in the 8090 alloy, while the greater volume fraction S in the 8091 alloy with its greater copper content immobilised all dislocations introduced by the stretch. The reduced room temperature proof strength with 7% stretch compared with 4% stretch (2), ascribed to an increase in the average length of S precipitates in the 7% stretched alloy, was not noted in the present work, and the increased strength with increased stretch seen in Table I at room temperature is similar to that reported by Miller et al (3) in 8090, even allowing for the underageing in their 6% stretched samples, as well as the early work of Pridham, Noble and Harris (1), which contrasted 3% stretched and unstretched alloys.

Conclusions

a) 8090-T8 alloy has an increased tensile and proof strength at room temperature when stretched by 7% prior to ageing, compared with that when a 3% stretch is applied.

b) At temperatures of 210°C and higher the mechanical properties of 8090-T8 are independent of the extent of prior stretching.

c) The change in shape of the engineering stress - engineering strain to one showing apparent work-softening occurs at a lower temperature (180°C) for a 7% stretched alloy than for a 3% stretched alloy (210°C). This is probably due to an increased rate of dynamic recovery in the 7% stretched alloy.

d) The apparent work softening seen at high temperatures is due to an increased tensile strain in the necked region during deformation.

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