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To cite this version:

HAL Id: jpa-00226559
https://hal.archives-ouvertes.fr/jpa-00226559
Submitted on 1 Jan 1987

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THE EFFECT OF SUPERPLASTIC DEFORMATION ON THE TENSILE AND FATIGUE PROPERTIES OF Al-Li (8090) ALLOY

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ABSTRACT

Superplastic deformation of the new generation of aerospace aluminium-lithium alloys has generated considerable interest in the aerospace industry not only with the potential savings in component weight and manufacturing costs but also for development of novel designs. Even though many papers have addressed the superplastic deformation characteristics of these alloys it is essential for their exploitation to determine and to understand the effect of the forming process on their mechanical properties. This paper presents results on the effect of superplastic strain and post form heat treatment on the tensile and fatigue performance of the Al-Li-Cu-Mg-Zr (8090) alloy with a low and high copper content of 1-1.2 wt% and 1.56 wt% respectively. Both alloys were superplastically formed biaxially under an imposed back pressure to prevent intergranular cavitation.

For the low copper containing alloys the tensile properties decreased with superplastic strain and were independent of quench rate. In contrast the tensile properties of the high copper containing alloy were quench rate sensitive. The as-formed and aged strength were initially lower than those determined for the low copper alloy although both the 0.2% proof stress and the tensile strength increased with superplastic strain. Incorporation of a solution heat treatment with a cold water quench prior to ageing resulted in a strength increase of 50 MPa over that determined for the low copper alloy and the tensile properties decreased with superplastic strain. The reduction in strength with superplastic strain is related to changes in the ageing kinetics resulting from dynamic recrystallisation and grain growth, which was similar for both alloys. The higher strength of the high copper alloy following re-solution heat treatment is due to enhanced homogeneous S phase precipitation. The S phase precipitation was reduced in the as-formed and aged material due to precipitation of a copper rich phase during the slower air cool, resulting in the lower strength.

The fatigue performance, under sinusoidal loading and stress ratio (R) = 0.1, was similar for both alloys and the endurance limit was slightly reduced with increased superplastic strain.

The results are compared to those for the aluminium-lithium alloy 8091, which has a higher copper content and with those for the superplastic aluminium alloys Supral 220 and 7475B. The implication of the results for manufacturing practice is also discussed.

INTRODUCTION

Superplastic forming offers the aerospace industry considerable potential not only for reducing weight and manufacturing costs, but also for the development of novel designs of aircraft structure. Additional
weight savings offered by the new generation of aluminium-lithium alloys has generated further interest. Although many papers have addressed the superplastic forming characteristics of a variety of Al-Li alloys, produced via ingot or powder metallurgy, at present there is no Al-Li alloy sheet commercially available which has been specifically manufactured for optimum superplastic formability. However production quality Al-Li-Cu-Mg-Zr (8090) alloy sheet is available and this product has, during its pre-production development, exhibited adequate superplastic capability for the production of envisaged aerospace structures [1]. The intergranular cavitation created during the superplastic deformation being suppressed/inhibited by application of an imposed back pressure during the forming operation. The availability of this alloy together with its attractive combination of properties and low quench rate sensitivity has ensured continued research at the RAE particularly with regard to the effect of superplastic strain on the mechanical properties. This paper presents results on the effect of composition extremes within the 8090 specification, in particular copper content, together with the effects of superplastic strain and post form heat treatment on the tensile and fatigue behaviour. The data is also compared with some preliminary results for the 8091 alloy which has basically the same composition but a higher copper content.

MATERIAL AND TEST PROCEDURES

The 8090 and 8091 sheets were commercially manufactured and supplied in the T4/T3 condition by Alcan International Ltd. The sheets had an unrecrystallised microstructure and their compositions, as determined at the RAE, are given in Table 1. The 8091 sheet was manufactured from pre-production quality ingot.

Table 1 Compositions of Sheets

<table>
<thead>
<tr>
<th>Major Alloying Elements Wt%</th>
<th>Li</th>
<th>Cu</th>
<th>Mg</th>
<th>Zr</th>
</tr>
</thead>
<tbody>
<tr>
<td>mm</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>8090</td>
<td>1.60</td>
<td>2.38</td>
<td>1.20</td>
<td>0.54</td>
</tr>
<tr>
<td>3.00</td>
<td>2.36</td>
<td>1.00</td>
<td>0.66</td>
<td>0.14</td>
</tr>
<tr>
<td>3.00</td>
<td>2.32</td>
<td>1.56</td>
<td>0.65</td>
<td>0.15</td>
</tr>
<tr>
<td>8091</td>
<td>2.0/3.0</td>
<td>2.36</td>
<td>1.98</td>
<td>0.78</td>
</tr>
<tr>
<td>8090 min</td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Spec max</td>
<td>2.20</td>
<td>1.00</td>
<td>0.60</td>
<td>0.04</td>
</tr>
<tr>
<td>Spec max</td>
<td>2.70</td>
<td>1.60</td>
<td>0.78</td>
<td>0.16</td>
</tr>
</tbody>
</table>

Property boxes were superplastically formed under an imposed back pressure of 3.45 MPa using forming temperatures and pressure-time cycles previously determined to suppress cavitation [1]. Tensile and fatigue test pieces were machined from the side walls and the base of these boxes. The tensile test pieces had a 30 mm gauge length and 6 mm gauge width and were tested to BS 4A4 using a 20 mm transducer extensometer. As slight variations in thickness occurred along some gauge lengths the average cross sectional area within the extensometer and the original cross sectional area at the point of fracture were used to calculate the 0.2% proof stress and tensile strength respectively. The fatigue test piece had overall dimensions 158 mm x 32 mm and an hour glass profile with a 9 mm minimum gauge width. Each set of test pieces covered a range of superplastic strains, calculated
at the minimum gauge width, due to slight thickness variations across
the base of the property boxes. The superplastic thickness strain
\( \varepsilon = \ln \frac{t_{\text{f}}}{t_{\text{i}}} \) where \( t_{\text{i}} \) and \( t_{\text{f}} \) are the initial and final thickness
respectively. The fatigue tests were carried out using sinusoidal
loading for a stress ratio \( (R) \) of 0.1 at a frequency of 120 Hz. The
test pieces were either post form aged for 24 h at 185°C or solution
heat treated for 20 minutes in the temperature range 515-540°C, cold
water quenched and aged. The as-formed and heat treated surfaces were
not removed prior to testing.

RESULTS AND DISCUSSION

Initial forming trials and microstructural examination confirmed that
the production quality sheet could be deformed superplastically under
the conditions previously determined [1] and that intergranular cavi-
tation was suppressed up to the maximum strain required ie 300% strain.
With increasing strain however the as-formed surface not in contact
with the die became 'rippled' with the ripples parallel to the final
rolling direction and correlating to the initial aligned microstruc-
ture.

The results, Fig 1, for the 1.6 mm and 3.0 mm thick 8090 sheets with
a copper content close to the specification minimum, and similar to
that for the alloys used in the previous study, confirm the benefit
of increasing the ageing time from 5 h to 24 h at 185°C and re-
emphasise the low quench rate sensitivity. The latter being extremely
important in reducing manufacturing problems associated with distortion
of complex shaped thin section components following rapid quenching.
The overall reduction in the 0.2% proof stress and tensile strength
with superplastic strain cannot be prevented in that it is a conse-
quence of the deformation process causing a loss of the sub-grained
structure due to recrystallisation and a slight increase in grain size,
Fig 2. The greater loss observed in tensile strength than 0.2% proof
stress may be misleading in that the tensile strength is ductility
dependent and for strains greater than \( \sim 1.0 \) the ductility was
reduced from 5-6% to 1-2%. This low ductility is considered a
function of the slightly tapered gauge lengths and the surface condi-
tion rather than any inherent embrittlement especially as for this
orientation recrystallised material is more ductile than unrecrystal-
lied material and a ductility of 8% has been reported for 1 mm thick
recrystallised sheet [2].

Increasing the copper content to 1.56 wt%, which is close to the
specification maximum, resulted in a dramatic difference in the post
formed tensile properties as clearly seen by comparison of Figs 1 and
3a, although there was no significant change in grain size, cf Fig 2.
The marked quench rate sensitivity was independent of forming at
either 515°C or 530°C. At low superplastic strains the as-formed and
aged properties were similar to those for the alloy with 1.2 wt%
copper but with increasing levels of strain an increase in strength
rather than a decrease occurred. Solution heat treating the as-formed
material at 515°C, 530°C or 540°C prior to ageing increased the
strength of the low strained material by 50 to 70 MPa and resulted in
a similar behaviour to that for the 1.2 wt% copper alloy in that the
strength decreased with superplastic strain. As a consequence the
strength differential observed at low strains with quench rate
decreases as the superplastic strain is increased with the strength
becoming quench rate independent after approximately 0.9 thickness
strain. The increased strength of the richer copper material was also
accompanied by an increase in the ductility from 4-5% to 8% although
this was reduced to 6% for material tested in the as-formed and aged
condition.
The post formed tensile properties for superplastically formed 8091 alloy sheet, Fig 3b, show the same behaviour as observed for the high copper 8090 alloy.

Plots of the strength data versus final thickness for the two sheet gauges investigated, Fig 4, revealed a marked, rather than gradual transition-in strength as indicated in Fig 3 for the alloys containing 1.56 and 1.98 wt% copper after air cooling. With the transition occurring at nominally the same thickness for both alloys and independent of initial gauge thickness the reduction in strength is not related to a specific amount of superplastic strain but to a microstructural change corresponding to quench rate. Microstructural studies confirmed this by revealing the presence of a copper rich grain boundary precipitate together with AlCuFe rich and AlZrTiCu containing matrix particles in the as-formed and aged material below the transition thickness Figs 5 and 6. The copper rich precipitate was not present in the material above the transition thickness or following the rapid cold water quench from 515°C, 530°C or 540°C. Precipitation was prevented if the initial cooling rate to below 300°C was greater than 5°C s⁻¹.

It is interesting to note that the as-formed and aged base line strengths decreased with increased copper content whereas the reverse was true if the material was solution heat treated prior to ageing. The increase in strength with copper content is well documented for the Al-Li-Cu-Mg-Zr system and is a consequence of S phase (Al₁₂CuMg) precipitation. This in turn encourages homogeneous deformation and improved ductility by inhibiting planar slip. The higher Mg+Cu also reduces the lithium solubility at the ageing temperature increasing the volume fraction of the homogeneously precipitated δ' (Al₃Li) which provides the major strengthening contribution in these alloys. In the low copper containing alloys the microstructure was predominantly δ' although a small amount of heterogeneously precipitated S phase was observed associated with sub-grain boundaries and defect sites such as dislocation loops and helices for both quench rates. In the high copper containing alloys a well developed homogeneous dispersion of S phase occurred together with the δ' after rapid quenching but not in the low strained as-formed and aged material. This is consistent with the formation of copper-rich grain boundary precipitates depleting the solute supersaturation and so reducing the S phase formation, strength and ductility.

The effect of superplastic strain and post form heat treatment on the fatigue behaviour of 8090 and 8091 sheet is shown in Fig 7. The results for the 8090 sheets, Figs 7a and 7b, revealed no significant difference in the fatigue performance for the two copper levels investigated following heat cycling at 530°C carried out to simulate the forming process and that no improvement occurred even when the high copper variant 8090 was cold water quenched rather than air cooled even though the tensile strength was increased by 70 MPa. Superplastic forming 8090 at 515°C caused an overall reduction in the fatigue performance of as-formed and aged material with a greater reduction determined for the high copper variant. The endurance limit at 5 x 10⁷ cycles being reduced from 200 ± 10 MPa to 135 ± 15 MPa and 110 ± 15 MPa for the low and high copper variants respectively. Forming at 530°C or re-solution heat treatment prior to ageing recovered the loss in fatigue resistance in the low copper variant (Fig 7a) but only partially recovered that in the high copper variant giving an endurance limit of 162 ± 10 MPa. The reason for this is unclear and is under further investigation. The fatigue performance of the superplastically formed 8091 (Fig 7c) was slightly better than that for the 8090 in the low stress-high life regime with an endurance...
limit at $5 \times 10^7$ cycles of $220 \pm 25$ MPa. As observed for the high copper 8090 alloy no improvement in the fatigue performance occurred with increased strength following the faster quench rate. Lack of material prevented heat cycled control data and consequently it was not possible to ascertain whether the fatigue performance of 8091 was reduced with superplastic strain as observed for the high copper 8090.

The results are compared with those for the superplastic aluminium alloys Supral 220 [3,5] and 7475E [4] in Figs 8 and 9. It can be seen that for use in the as-formed and aged condition the low copper 8090 has the advantage as its tensile properties are more consistent with superplastic strain. However if the as-formed components can be re-solution heat treated higher strengths can be achieved in the high copper 8090 and 8091, although these strengths are lower than those achieved for 7475E. The thickness of this latter alloy however is restricted to less than 3 mm as a consequence of the special thermo-mechanical processing required to make it superplastic. The 0.2% proof stress advantage observed for Supral 220 over high copper 8090 and 8091 at thickness strains greater than 0.5 will tend to be negated as a range of superplastic strains will exist in the formed component. The data in Fig 9 reveals that all the alloys have a similar fatigue performance following re-solution heat treatment and ageing. It should however be remembered that this fatigue performance is also achieved in low copper 8090 formed at 530°C and aged only. Although some structures will be strength or fatigue designed the majority will be stiffness critical and thickness limited. In these situations 8090 and 8091 alloys with their 10-15% higher elastic modulus and their lower density will out perform Supral 220 or 7475E. The lack of quench rate sensitivity in the low copper 8090 is also extremely important when the production of multi-sheet structures is considered not only because of distortion on quenching but also because of problems in achieving high cooling rate in internal structure.

In conclusion variation in the copper content in 8090 from 1.0 wt% to 1.56 wt% did not affect the superplastic capability of the alloy but the higher copper content did cause the alloy to become quench rate sensitive. After re-solution heat treatment and ageing higher strengths with improved ductility were achieved in the high copper alloy due to additional S phase precipitation.

REFERENCES


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Fig 1 Effect of postform heat treatment and superplastic strain on the tensile properties of 8090 (1.2 wt% Cu) aged 24 h at 185°C [solid symbols - CWQ open symbols AC].

Fig 2 Effect of superplastic strain copper content on the grain size of 8090.
Fig 3 Effect of forming temperature, post form heat treatment and superplastic strain on the tensile properties of A) 8090 (1.56 wt% Cu) and B) 8091 aged 24 h at 185°C [solid symbols - CWQ, open symbols - AC].
Fig 4 Effect of quench rate and final thickness on the tensile properties of superplastically formed 8090 and 8091 aged 24 h at 185°C [(L) - low copper, (H) - high copper].
Fig 5 Effect of superplastic strain and quench rate on the microstructure of 8090 (1.56 wt% Cu). Etch Kellers Reagent.
Fig 6 Effect of quench rate A] cold water and B] air cool. on precipitation in 8090 (1.56 wt% Cu) using back scattered scanning electron microscopy.
Fig 7  Effect of superplastic strain and post form heat treatment on the fatigue of 8090 and 8091 aged 24 h at 185°C under sinusoidal loading and R = 0.1.
Fig 8 Comparison of the tensile properties of superplastically formed aluminium alloys in the -T6 condition.

Fig 9 Comparison of the fatigue performance for superplastically formed aluminium alloys in the -T6 condition after ~1.2 superplastic strain.