THE MECHANICAL PROPERTIES OF UNSTRETCHED ALUMINIUM-LITHIUM ALLOY 2095 (WELDALITE™ 049) IN THE LOW TEMPERATURE AGED CONDITION

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Abstract

For the Al-Cu-Li-Mg-Ag-Zr alloy X2095 it has been shown that alloys aged at 125°C for up to 208 hours in the unstretched condition show characteristic underaged microstructures with low proof strengths and, for their proof strengths, relatively low fracture toughnesses. This is believed to be because of the relatively heterogeneous slip pattern in the absence of extensive T_1 (Al_2CuLi) precipitation. Confirmatory tests of stretched specimens, which had been aged for substantially less than the 120 hours required for extensive T_1 precipitation, indicated the same low proof strength and similar fracture appearance.

Introduction

The Al-Cu-Li-Mg-Ag-Zr alloy X2095 (Weldalite™ 049) has been designed (1) to have suitable tensile properties in the artificially aged and stretched (-T8), artificially aged and unstretched (-T6) and naturally aged (-T3 and -T4) conditions, as can be seen in Table I. Compared to other Al-Li alloys Weldalite™ has a much greater response to natural ageing, as well as having the ability to give good strength and ductility after artificial ageing, both in the stretched and unstretched conditions. In this way the alloy can be used in welded constructions (2), where it is not possible to carry out a full heat treatment after welding, and in forgings (3), where the extent of stretching will not be uniform in all parts of the forged component and significant variation of mechanical properties with amount of post-solution treatment stretching would be undesirable.

The amount of fracture toughness data is limited, but early work by Cho et al (4) indicated that the plane strain fracture toughness in the -T8 condition ranged from 18.6-25 MPa $\sqrt{m}$ for proof strengths of 610-586 MPa. Although these data were not from compositions optimised for good fracture toughness, it could be seen that, although the fracture toughness was high for its strength level compared with that of conventional aluminium alloys, as well as early Al-Li alloys such as 2020-T6, the absolute values of fracture toughness were sufficiently low to merit investigations into methods of increasing the fracture toughness, either by changes in composition or heat treatment.

In an attempt to increase the fracture toughness of X2095, while retaining as much of the high strength as possible, Blankenship and Starke (5) age hardened the alloy, after a stretch, at 125°C for 120 hours. The ageing temperature was selected (6) to be below the δ' solvus and yet above the minimum temperature for T_1 (Al_2CuLi) precipitation. This would ensure T_1 precipitation within the alloy subgrains and produce a reduced density of T_1 plates on subgrain boundaries. The plane strain fracture toughness achieved was 29 MPa$\sqrt{m}$ at a 0.2% proof strength of 660 MPa.

Table I Tensile Properties of X2095 in Various Heat Treated Conditions

Heat Treatment	Tensile Strength (MPa)	0.2% Proof Strength (MPa)	Ductility (%)	Fracture Toughness (L-T) (MPa$\sqrt{m}$)	Reference
-T3	529	407	16.6	36.9	(7)*
-T4	591	438	15.7		(7)
-T6 (underaged)		607		30.0	(7)
-T6	718	683	3.8		(7)
-T8	715	696	5.5		(7)
-T8	700	660	8	20	(6)**
-T8	681	667	6.5	28.2	(8)***

* (7) used an alloy with 6.2% Cu and 1.37% Li

** (6) used an alloy with 4.8% Cu and 1.24% Li

*** (8) used an alloy with 4.1% Cu and 1.49% Li. Their fracture toughness was a K_Q value , but close to K_{Ic}.

The present work aims to examine how low temperature ageing would proceed in the absence of a stretch after solution treating, as could occur in either welded constructions or forgings.

Experimental

A 12.7 mm thick plate of X2095, produced by Reynolds Metals, was supplied in the -T8 temper. Its composition (weight per cent) was Cu 4.47, Li 1.25, Mg 0.36, Ag 0.49, Zr 0.13, Fe 0.07, Si 0.05, balance Al.

The plate was re-solution treated at 504°C for 0.75 hours and water quenched. Age hardening was carried out immediately at 125°C for up to 208 hours.

Tensile tests were carried out on 6 mm diameter specimens with a 25 mm gauge length with the tensile axis parallel to the rolling direction. The initial strain rate was $1 \times 10^{-3}\ s^{-1}$.

Compact tension testpieces, 12.5 mm thick with an L-T orientation, were used in the determination, according to BS 5477: 1977, of plane strain fracture toughness. The initial fatigue pre-cracking was done at a stress intensity range of 6.64 ± 5.46 MPa$\sqrt{m}$.

Transmission electron microscopy was carried out on a JEOL JEM-100CX microscope. Specimens were electropolished using a 30% nitric acid-methanol solution at -25°C. In order to remove silver deposits from the foils, it was necessary to clean them successively in ethanol, 50% nitric acid and distilled water, prior to examination in the microscope.

Results

Tensile properties and fracture toughness data are shown in Table II. These show an increasing tensile yield strength as the ageing time increases, but all show the high ductility and work-hardening rate of underaged structures. In spite of the low tensile yield strengths, fracture toughnesses remain low.

Table II Tensile Properties for Unstretched X2095 as a Function of Ageing Time at 125°C

Ageing time (hours)	Tensile Strength (MPa)	0.2% Proof Strength (MPa)	Ductility (%)	Fracture Toughness (L-T) (MPa$\sqrt{m}$)
14	537	328	22	25.4*
35	552	356	21	25.5
208	554	385	19	23.1
-T8 (control)	636	600	11	22.6 (9)

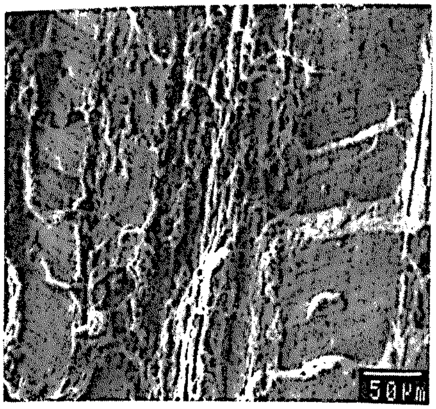
* This is a K_Q value but very close to K_{Ic}

Results for the alloy stretched by 5% and aged at 125°C are shown in Table III. It can be seen that the 35 hour specimen shows the same tensile characteristics as the underaged and unstretched specimens in Table II, while the 120 hour specimen shows the high proof strength, relative to the tensile strength, described by Blankenship and Starke (6).

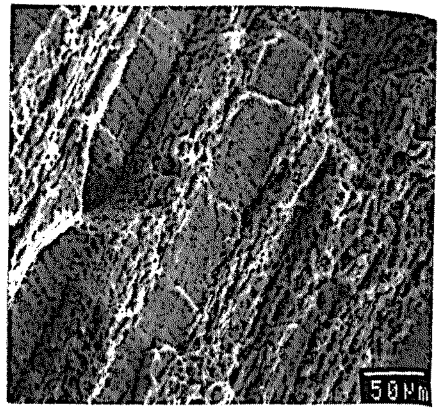
Table III Tensile Properties for Stretched X2095 as a Function of Ageing Time at 125°C

Ageing time (hours)	Tensile Strength (MPa)	0.2% Proof Strength (MPa)	Ductility (%)
35	500	332	20
120	584	517	17

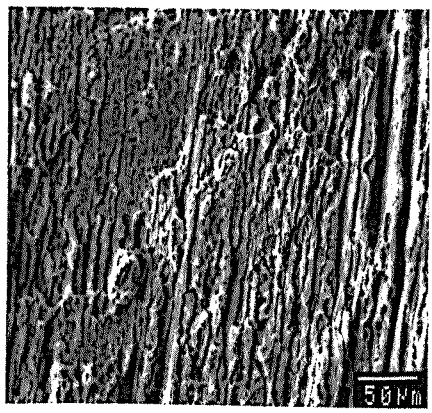
Fractographs (Fig.1), taken from the tensile testpieces, show for the underaged samples a pattern of ductile fracture separated by ledges. These are coarser for the 14 hour aged and unstretched specimen (a) than for the 35 hour aged and stretched specimen (b). With the 120 hour aged and stretched specimen the scale of the ledges is much reduced and the fracture surface appears more homogeneous.



a)



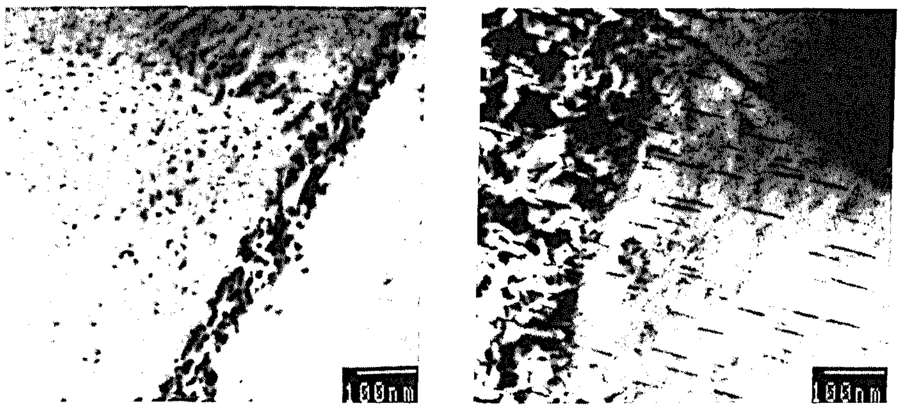
b)



c)

Figure 1 Fracture surfaces taken from tensile testpieces of a) 14 hours, 125°C age, unstretched, b) 35 hours, 125°C age, stretched 5% and c) 120 hours, 125°C age, stretched 5% specimens.

Transmission electron microscopy (Fig.2) shows the presence of extensive T_1 precipitation within the subgrains for the stretched sample, aged at 125°C for 120 hours, but there are no similar precipitates in the underaged unstretched sample.



a) b)

Figure 2 TEM microstructure of specimens aged at 125°C: a) 35 hours, unstretched and b) 120 hours, 5% stretch.

Discussion

For Al-Li-Cu alloys strengthened by T_1 precipitates artificial ageing is accelerated by stretching after solution treatment. Cassada, Shiflet and Starke (10) showed, in the case of an Al-2.45Li-2.45Cu-0.18Zr alloy, the importance of the density of dislocation jogs rather than the dislocation density and hence the great acceleration of ageing possible at high stretches, such as 6%. In the case of the high copper, silver-bearing Weldalite™ alloy the effect of stretching on the ageing kinetics was studied by Kumar, Brown and Pickens (11) and related to the work on the ageing kinetics of unstretched Weldalite™ carried out by Gayle, Heubaum and Pickens (12). For the unstretched material Gayle started with the naturally aged (> 1000 h) -T4 temper, strengthened by GP zones and δ' precipitates. During ageing at 180°C the precipitates changed, GP zones and δ' disappearing and being replaced by T_1, θ' (Al_2Cu) and S (Al_2CuMg) in the peak aged condition after 24 hours. Stretched material was aged at 160°C rather than 180°C and reached peak hardness in 24-48 hours. Comparing the -T3 and -T4 tempers, strength in the stretched condition is lower, and the GP zones are much less marked than those in the unstretched alloy. At peak hardness the same precipitates (T_1, θ' and S) were present as in the unstretched alloy, although on overageing only the T_1 remained, confirming earlier work by Langan and Pickens (13) on peak aged material. More recent work (14) has shown that the precipitates, even though T_1 predominates, of peak aged Al-Cu-Li-Mg-Ag-Zr alloys will be dependent on the exact Cu, Mg and Li contents of the alloy. S will be associated with higher lithium and magnesium content alloys, while θ' will be found where copper contents are high.

Considering the alloys aged at 125°C, Blankenship and Starke (5) showed that, after 120 hours for stretched samples, the structure has T_1 precipitates within the subgrains. With no stretch ageing will be slower, and alloys aged for up to 208 hours are underaged; indeed it is likely that it will take up to 1000 hours to achieve a fully aged structure at such a low ageing temperature as 125°C. The tensile test results (Table II) confirm this with all tests showing a high work hardening rate, high tensile ductility and a low proof strength. The tensile strength of 554 MPa for 208 hours ageing is much closer to that produced, either by the standard -T8 heat treatment (636 MPa) or the 120 hour at 125°C ageing of stretched material (584 MPa), than is the proof strength (385 MPa). Similarly underaged (35 hours at 125°C) stretched samples (Table III) show similar characteristics of relatively high tensile strength (500 MPa) and low proof strength (332 MPa), as do the corresponding unstretched alloys (Table II). In comparison with the results of Blankenship and Starke (6), the tensile and proof strengths in this work are lower with respect to the -T8 properties. This may indicate that the variations in the heat treatment (amount of stress, ageing time and time of natural ageing after stretching) at 125°C will affect the mechanical properties more than would such variations at higher ageing temperatures.

The TEM microstructures (Fig.2) show how there is extensive within subgrain T_1 precipitation in the alloy given a stretch and aged for 120 hours. For the 35 hour unstretched alloy there was little such precipitation of T_1 and even for the 208 hour aged alloy, the unstretched alloy had little T_1 present within the subgrains. Since the work of Cassada, Shiflet and Starke (10) had shown the importance of dislocation jogs in the precipitation of T_1, it is likely that the absence of a stretch and the low ageing temperature would encourage precipitation, if it occurs, to be away from the subgrain interiors.

Fracture which involves grain boundaries splitting back from the fracture surface has been described by Miller et al (15) for an Al-Li-Cu-Mg-Zr alloy in both the solution treated and peak aged conditions. They ascribed this to a weak grain boundary or subgrain boundary giving an easy fracture path. In the present work it is suggested that the absence of large numbers of T_1 precipitates within the subgrains will increase the heterogeneous slip within the subgrains, and, together with the possible presence of grain boundary precipitates, will make deep grain boundary splitting more likely in the underaged specimens than in the stretched and 120 hour aged specimens.

Compared with other aluminium alloys of similar proof and tensile strengths, such as the Al-Cu-Mg alloy 2024-T4 with its fracture toughness of 55 MPa$\sqrt{}$m (16), the fracture toughnesses shown in Table II are low. If the toughness of a binary Al-3.2Li alloy (17) with its intensely planar slip is considered, a peak aged alloy with a proof strength of 263 MPa has a fracture toughness (measured with a pre-cracked slow bend Charpy testpiece) of 19.5 MPa $\sqrt{}$m. Similarly ternary Al-2Li-4Mg alloys (18), heat treated to a proof strength of 381 MPa and a tensile strength of 508 MPa, showed a fracture toughness of 24 MPa$\sqrt{}$m due to the presence of both planar slip and grain boundary precipitation of Al_2MgLi. These early alloys show inferior fracture toughnesses to those of modern alloys, which use combinations of ternary and quaternary alloying additions and stretching to ensure precipitation within the grains of ternary precipitates, such as S (Al_2CuMg) and T_1. Thus, in the underaged alloys of the present work, the shortage of T_1 precipitation within the subgrains may well contribute to low alloy fracture toughness.

Conclusions

a) In the unstretched condition Weldalite™ remains in an underaged state for ageing times of up to 208 hours at 125°C, and will require upwards of 1000 hours to achieve proof strengths similar to that of the stretched alloy, age hardened for 120 hours at 125°C.

b) In this underaged state fracture toughnesses, in the L-T orientation, are in the range of 25.4-23.1 MPa$\sqrt{m}$ for proof strengths of 325-385 MPa, these being much lower than for other aluminium alloys of the same strength.

c) Underaged alloys do not show extensive T_1 precipitation within the subgrains, and it is believd that the relatively low fracture toughness is associated with this relative lack of within subgrain T_1 precipitation.

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