Fracture 1977, Volume 2, ICF4, Waterloo, Canada, June 19 - 24, 1977

INFLUENCE OF PRIOR TREATMENT ON HIGH STRAIN FATIGUE LIFE OF ALUMINIUM ALLOY RR58

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INTRODUCTION

Achievement of high fatigue resistance in alloys in general is of considerable practical significance. Alloys of Al in particular have attracted attention in this respect because of their great utility in the aerospace industry. Calabrese and Laird [1-3] have noted the relevance of their observations on high strain fatigue fracture mechanisms in an Al-4 pct Cu alloy to a wide range of industrial alloys. In regard to the thermomechanical processing (TMP) conflicting results have been reported on attempts to modify through TMP high cycle fatigue strength of several Al alloys. While Krause and Laird [4] (Al-4% Cu), Broek and Bowles [5] (2024), Ostermann [6] (7075) find a favourable influence, Lyst [7] (2024) reported no effect and Russo et al. [8] (7075 and 7039) a detrimental effect of TMP on high cycle fatigue resistance. In the present work Al alloy RR58 has been investigated in high strain low cycle fatigue at room temperature with focus on the effect of thermal and thermomechanical treatment on life. Uniaxial tensile properties and tear resistance have also been evaluated and the fatigue data are examined in the light of both monotonic and cyclic mechanical behaviour.

EXPERIMENTAL

The experimental alloy RR58 had the nominal composition, by wt.%, 2.67 Cu, 1.59 Mg, 1.11 Fe, 1.03 Ni, 0.22 Si, balance Al, and is based upon the pseudo-binary Al-Al_2CuMg (S-phase) system. Following solution annealing at 808±3 K, samples were quenched in water and treated as described below:

I. Thermal Treatments:

1) T-UA : 463 K/2.5 h (underageing)
2) T-PA : 463 K/15 h (peakageing)
3) T-OA : 463 K/72 h (overageing)

II. Thermomechanical Treatments:

1) T-NH4A : RT/15 h + rolling reduction at RT by 4% in thickness + 463 K/15 h
2) T-NH25A : RT/15 h + rolling reduction at RT by 25% in thickness + 463 K/15 h
3) T-AH4A : 413 K/4 h + rolling reduction at RT by 4% in thickness + 463 K/15 h

Characteristic features of the microstructures of the treated samples, as revealed by transmission electron microscopy, are GPB zones in T-UA, GPB zones and intermediate S' precipitates in T-PA, uniformly distributed coarse equilibrium S precipitate in T-OA samples. The relatively high dislocation density introduced through intermediate cold reduction re-

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mained even after the final ageing treatment. Cold reduction, however, breaks down the S precipitate to a large extent. For the same pre- and final ageing treatments the degree of fineness of the precipitate enhances with increasing amount of reduction. Replacement of natural preageing with artificial preageing has a similar effect. In addition, the structure in all the tempers consisted of undissolved $FeNiAl_9$ particles and possibly [9] Si particles. There is no significant effect of pre-strain on precipitate distribution. The hardnesses of thermomechanically treated samples, viz. 144 VPN (T-NH4A), 149 VPN (T-NH25A) and 157 VPN (T-AH4A), were close to that (148 VPN) of T-PA temper and lay on the slightly overaged side of the respective ageing curves.

Strain controlled fatigue tests were performed at room temperature (RT) on samples machined to the geometry shown in Figure 1 using a Floor Model Instron at a constant frequency of 0.033 Hz. Tests were carried out to fracture as a function of extension range. The plastic strain range ($\Delta\varepsilon_p$) was determined at midlife from the mechanical hysteresis loops. Fatigue hardening/softening characteristics were derived from the load response curves recorded throughout the test.

For the Kahn-type tear tests, employed for evaluation of tear resistance, specimens of 0.012 m gauge length, 0.02 m width and 0.0025 m thickness with a sharp notch of 60° extending up to the axis of the specimen were loaded statically on the Instron until a crack developed at the root of the notch and travelled across the width of the specimen. Unit Propagation Energy (UPE), the primary measure of toughness, was determined [10] from the load-displacement curve. For the RT tensile tests on the Instron, standard Hounsfield 12 specimen geometry was used.

RESULTS AND DISCUSSION

The effect of plastic strain amplitude on cyclic hardening and softening rate was examined by measuring [11] the slopes of the stress vs $\Sigma\,\Delta\varepsilon_p$ curves at a cumulative plastic strain of 0.2 and the results are presented in Figure 1. The alloy in tempers involving only thermal treatments is characterised by cyclic hardening in a manner analogous to annealed metals. Rapid hardening during the first few cycles followed by saturation, exhibited by the alloy in T-OA temper, may be understood on the basis of interaction between shuttling dislocations and the geometrically stored ones as explained by Calabrese and Laird [2]. The alloy in tempers involving intermediate working, irrespective of pre- and final ageing treatments, is found to soften during fatigue after initial rapid hardening, quite similarly to cold worked metals.

From Figure 1 it is evident that the cyclic hardening rate, regardless of the treatment, increases with an increase in $\Delta\varepsilon_p$. On the other hand, the rate of softening increases with increase in $\Delta\varepsilon_p$ and starts falling beyond a plastic strain range of about 2.5%. The rate of hardening is greater for the T-UA than it is for the T-PA temper. For the same pre- and final ageing treatments, greater intermediate working leads to an increased rate of softening. It is noteworthy that the T-NH4A temper displays a degree of stress stability which is only inferior to that of T-OA condition. Following Landgraf [12] in a qualitative manner, these cyclic hardening, stability and softening observations can be broadly rationalised in terms of the monotonic work hardening exponent, n (Table 1). Thermal treatments resulting in cyclic hardening (T-UA and T-PA) as well as a saturation stress (T-OA) have n values higher than 0.05 as compared to the thermomechanical treatments (T-NH4A, TNH25A and T-AH4A), which cause fatigue softening, for which n is nearly 0.04.

Fractured surfaces, examined in a Cambridge Stereoscan, in all cases, principally show regions covered by nearly equiaxed dimples (Figure 2). Striations are not frequent and cleavage type fracture of brittle particles is the other noteworthy feature. The particles within the dimples were analysed and shown to contain Al, Fe, Ni, Cu and Mg. The sites analysed thus can be presumed to be a superposition of $FeNiAl_9$ particles and S phase precipitate (Al_2CuMg). These fractographic observations are in line with those made earlier [13,14] on strong commercial Al alloys.

The dependence of fatigue life expressed as number of cycles to failure (N_f) on $\Delta\varepsilon_p$ is shown in semilog plots (Figure 3). Clearly there is considerable impact of prior treatment on N_f. Maximum N_f values for the T-OA (overaged) temper, at any $\Delta\varepsilon_p$, is nearly 2.5 times that for T-UA (underaged) treatment which results in the lowest N_f values. Among the thermomechanical treatments, T-NH4A leads to the highest N_f in this group which is greater than that obtained for the T-PA (peakaged) condition and is only smaller than N_f for the T-OA temper. The sequence of N_f values at any $\Delta\varepsilon_p$ for the different treatments in increasing order is listed in Table 1.

The differences in fatigue life may first be examined in relation to monotonic mechanical test results (Table 1). For the controlled plastic strain situation it has been emphasized in the past [15] that a larger monotonic fracture ductility ($\varepsilon_f = \ln A_o/A_f$) is indicative of a larger fatigue life (N_f). This is true for the present results only if T-UA results are deleted from consideration. Correlations of the type suggested in the literature between monotonic work hardening rate [16] and plain strain fracture toughness [17,18] on the other hand and crack growth rates and thus fatigue life on the other are not also borne out by the results in Table 1.

The cyclic stress-strain behaviour, however, provides striking explanations for variations in N_f observed here in terms of (1) degree of hardening/softening (Figure 1 and Table 1) and (2) cyclic strain hardening exponent n' ($\Delta\sigma = K\,\Delta\varepsilon_p^{n'}$ where $\Delta\sigma$ is the stress range at midlife [19] and K is the cyclic strength coefficient).

Alloy in the overaged condition (T-OA), which is characterised by saturation in its cyclic stress response, shows highest fatigue life. The microstructure, in this instance, contains impenetrable particles and is analogous to the situation of coarse θ' particles in the Al-4% Cu alloy of Calabrese and Laird [2,3]. At the other extreme, when the alloy contains penetrable zones (T-UA) precipitate disordering results from the cutting process of the to and fro dislocation movement which creates soft regions and consequently localisation of plastic strain [1]. The greater stability of the alloy in the T-NH4A condition, as compared to that of T-PA, is probably because of a smaller fraction of unstable GPB zones in the thermomechanically treated material. The increasing rates of softening in the two other thermomechanical treatments, involving either an artificial preageing step (T-AH4A) with the same amount of intermediate reduction as in the T-NH4A treatment or greater amount of intermediate working (T-NH25A), results in progressively decreasing fatigue life. Thus, with increasing deviation from the stable situation evidenced by either increasing cyclic hardening or softening rate, fatigue life is progressively impaired.

It is obvious from the data recorded in Table 1 that the rank order of n' values strictly corresponds to the rank order of N_f values. Although correlations of fatigue life with n' have earlier been illustrated, the present results seem to be an excellent pointer to n' being a very reliable index of the material microstructure for predicting fatigue life. As pointed out by Coffin [20], for n' approaching zero, the plastic strain concentrates at the crack tip in highly localised shear planes while the stress concentration is low. On the contrary, for high n', the plastic strain becomes much more diffuse around the crack tip and the stress concentration is high. And thus in strong alloys, such as the present one, capable of resisting, without fracture, high stress concentrations, increasing n' provides a significant advantage in rendering slip more homogeneous and thus in improving fatigue life. Tomkins' [21] equations, too, as is well known, predict greater fatigue lives for higher n'.

Another useful conclusion emerges from the data of Table 1. Mild deformation prior to final ageing (T-NH4A) significantly improves the tensile yield strength compared to treatments without mechanical deformation while retaining the advantage of a fairly stable cyclic stress rate, relatively higher n' and thus longer fatigue life, which high strain fatigue characteristics are inferior only to those of the T-OA condition. The treatment T-NH4A, among those investigated here, should be, on balance and an overall basis, the best for Al alloy RR58.

ACKNOWLEDGEMENTS

We are greatful to Professor T. R. Anantharaman for encouragement and provision of facilities, Dr. W. Blum and Dr. R. P. Wahi, members of Professor B. Ilschner's group, for interesting us in and supplying the Al alloy RR58, Dr. V. Kutumba Rao for obtaining the SEM picture used here, and Dr. Vakil Singh for useful discussions.

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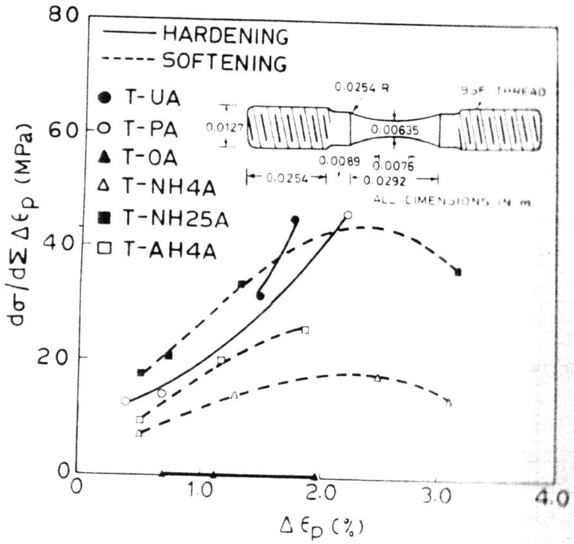

Figure 1 Cyclic Strain Hardening/Softening Rates at Room Temperature at a Cumulative Strain of 0.20 versus Plastic Strain Range. (Note also Fatigue Specimen Geometry)

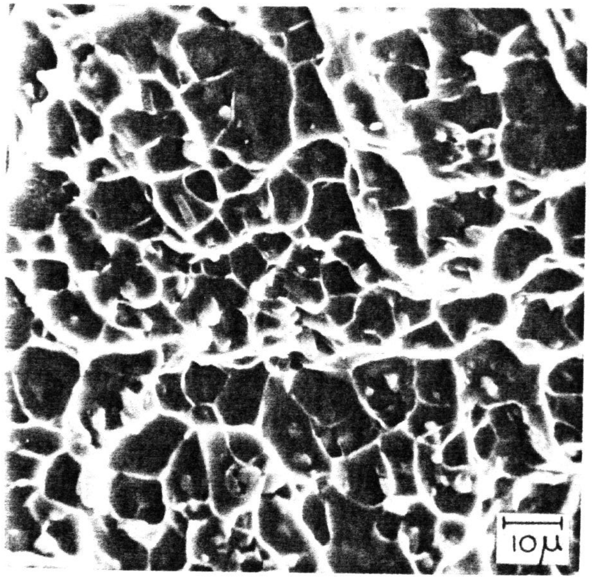

Figure 2 Typical Room Temperature Fatigue Fracture Surface of Al Alloy RR58 in T-NH4A Temper ($\Delta\varepsilon_p$ = 2.5%, N_f = 105)

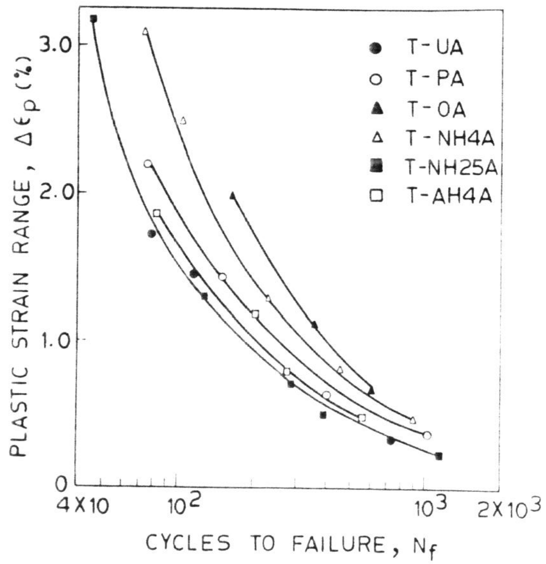

Figure 3 Plastic Strain Range versus the Number of Cycles to Failure for all Treatments

Table 1 Results of Fatigue, Tension and Tear Tests on Al Alloy RR58

N_f for any $\Delta\varepsilon_p$ (increasing rank order)	Degree* of Hardening (H) or Softening (S)	$\sigma_{0.2}$ MPa	ε_f	n	UPE kg.m/m^2	n'
T-UA	H-2	294	0.28	0.10_5	4320	0.04_9
T-NH25A	S-3	472	0.15_3	0.03_7	600	0.05
T-AH4A	S-2	475	0.15_7	0.04_2	970	0.05_7
T-PA	H-1	415	0.18	0.05_5	2070	0.06_3
T-NH4A	S-1	442	0.22	0.04	850	0.06_6
T-OA	ZERO	357	0.24	0.07	845	0.10_3

*Increasing Digits Following Leters H and S Indicate Increasing Rates of Hardening and Softening Respectively at any $\Delta\varepsilon_p$