LOW CYCLE FATIGUE BEHAVIOUR OF
AN UNDERAGED Al-Li-Cu-Mg ALLOY

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ABSTRACT
Fully reversed, total axial strain controlled low cycle fatigue tests were conducted on aluminium-lithium alloy plate of AA8090 composition in underaged (UA, T3) condition at strain amplitudes ranging from 0.0045 to 0.015. The properties studied include cyclic stress response behaviour, cyclic stress-strain data and fatigue life variation with plastic strain amplitude ($\Delta e_p/2$), average stress amplitude ($\Delta \sigma/2$) or average plastic strain energy per cycle ($\Delta W_p$). The fatigue data obtained for the underaged alloy have been compared with those of a peakaged (PA, T8E51) temper alloy plate of similar composition. The alloys in the two ageing conditions exhibit similar cyclic stress response behaviour, which varies with applied strain amplitude. Initial cyclic hardening was followed by cyclic stability till fracture at lower strain amplitudes, while cyclic softening followed initial hardening at higher strain amplitudes. A comparison of the fatigue life data of the two alloys with reference to $\Delta e_p/2$, $\Delta \sigma/2$ or $\Delta W_p$ reveals that the underaged alloy possesses lower fatigue resistance than the peakaged alloy. The UA alloy exhibits bilinear fatigue life power-law relationships with power-law constants at lower strain amplitudes being higher than those at higher strain amplitudes, in a manner similar to that reported earlier for the PA condition. The observed transition in the fatigue life power-law relationships is attributable to changes in the deformation and/or deformation-assisted fracture modes.

1. INTRODUCTION
Lithium containing aluminium alloys have become attractive candidate materials to replace the traditionally used 2XXX- and 7XXX series aluminium alloys for aircraft structural applications because of their superior properties such as reduced density, improved elastic moduli, higher specific strength, enhanced resistance to fatigue crack propagation and superior strength-toughness combinations at cryogenic temperatures $^{1-3}$. However, progress in the realization of their potential has been inhibited by their limited ductility, inferior low cycle fatigue resistance and low fracture toughness $^{3-6}$. Attempts have been made to improve fatigue and fracture resistance of these alloys by resorting to lower strength levels. One such option is to choose an underaged temper, in which condition these alloys possess higher fracture resistance as compared to that of the peakaged condition $^6$ and, hence, a strength-toughness combination comparable to the traditionally used aluminium alloys. However, Eswara Prasad and coworkers $^{7-9}$ have demonstrated that though the underaged Al-Li 8090 alloy possesses higher mode I fracture toughness, its resistance to fracture under shear loading (either in mode II (in-plane shear) or mode III (anti-plane shear) or even the combined tensile and shear loading, i.e., mixed-mode I/III) is significantly inferior to that of the peakaged alloy. Such effects of aging on the low cycle fatigue properties of Al-Li alloys are not yet understood. The low cycle fatigue studies conducted on these alloys till date are limited to the effects of alloying addition, degree of recrystallisation and
directionality in properties. In the present paper, we report the low cycle fatigue (LCF) properties of a nearly recrystallised, quaternary Al-Li-Cu-Mg alloy plate in the underaged (T3) condition. LCF properties in the underaged condition are compared with those of a nearly unrecrystallised, peakaged (T8E51) alloy of similar composition. Detailed results of the LCF behaviour of peakaged (T8E51) alloy have been reported earlier.

2. EXPERIMENTAL

The quaternary Al-Li-Cu-Mg alloy, used in the present study, was received in the form of 20 mm thick plate in the UA (T3) temper condition. The nominal composition of the alloy (in wt.%) is Al-2.2Li-1.1Cu-0.7Mg. The alloy contained minor addition of Zr, to impart grain refinement as well as for inhibiting recrystallisation. The chemical composition of the peakaged (PA, T8E51) alloy plate of 12.5 mm thickness, considered here for the purpose of comparison, is the same as that of underaged alloy plate, except for the Zr content. The PA alloy plate contained higher levels of Zr, i.e., 0.12 wt.% and possessed nearly (80%) unrecrystallised grain structure. On the other hand, the UA alloy plate contained 0.08 wt.% Zr and the resulting grain structure was nearly recrystallised (80%). Three dimensional optical micrographs shown in Fig.1 demonstrate such variation in grain structure of the two alloy plates. The details of the microstructural features of the two alloy plates are summarized in Table 1.

Tensile properties of the two alloy plates, obtained in the longitudinal (L, specimen axis oriented parallel to the rolling direction) direction, are given in Table 2. The UA alloy shows higher work hardening exponent (n = 0.072) as compared to that of the PA alloy (n = 0.053). The two alloy plates in the L direction exhibit uniform elongation and total elongations of the same magnitude. Further, these values are equal to or lower than the work hardening exponent, which implies that the alloys failed before the onset of necking.

Fully reversed, total axial strain controlled LCF tests were conducted in laboratory air atmosphere at room temperature. The tests were conducted on a computer controlled, servo-hydraulic MTS 880 test system at a frequency of 0.1 Hz. Cylindrical specimens of 15 mm gauge length and 6.25 mm gauge diameter were used. The test specimens were mechanically polished, followed by chemical polishing at 343 K using a solution containing, by volume percent, 70 H3PO4, 25 H2SO4 and 5 HNO3. The surface finish thus obtained was better than 0.1 μm. The specimen axis was oriented parallel to the rolling direction (L, longitudinal direction). The data recorded for each test include tensile peak stress (σT), compressive peak stress (σC), stress range (Δσ = σT + σC) and plastic strain amplitude (Δε/2) as a function of elapsed cycles. The failure limit (Nf) at any strain amplitude corresponds to either complete separation of the specimen or 20% drop in the peak stress range. The fractured specimens were examined under...
**Table 1**

MICROSTRUCTURAL FEATURES OF THE Al-Li 8090 ALLOYS

<table>
<thead>
<tr>
<th>Description</th>
<th>Underaged (T3) alloy plate</th>
<th>Peakaged (T8E51) alloy plate</th>
</tr>
</thead>
<tbody>
<tr>
<td>1. Degree of recrystallisation, %</td>
<td>80 - 90</td>
<td>20 - 25</td>
</tr>
<tr>
<td>2. Grain width, mm</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(a) Longitudinal</td>
<td>10 - 30</td>
<td>5 - 9</td>
</tr>
<tr>
<td>(b) Long-transverse</td>
<td>4 - 10</td>
<td>1 - 2</td>
</tr>
<tr>
<td>(c) Short-transverse</td>
<td>6 - 15</td>
<td>2 - 3</td>
</tr>
<tr>
<td>3. Grain aspect ratio</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(a) Longitudinal</td>
<td>1 - 2</td>
<td>1 - 4</td>
</tr>
<tr>
<td>(b) Long-transverse</td>
<td>6 - 10</td>
<td>40 - 45</td>
</tr>
<tr>
<td>(c) Short-transverse</td>
<td>3 - 8</td>
<td>20 - 25</td>
</tr>
<tr>
<td>4. Intragranular precipitates</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(a) Al\textsubscript{3}Li (d\textsuperscript{'} )</td>
<td>High density (5-10 nm in diameter)</td>
<td>High density (25 - 30 nm in diameter)</td>
</tr>
<tr>
<td>(b) Al\textsubscript{2}CuMg (S\textsuperscript{'} )</td>
<td>Low density</td>
<td>High density</td>
</tr>
<tr>
<td>(c) Al\textsubscript{2}CuLi (T\textsubscript{1} )</td>
<td>Low density</td>
<td>Low density</td>
</tr>
<tr>
<td>5. Intergranular precipitates</td>
<td></td>
<td></td>
</tr>
<tr>
<td>(a) Al\textsubscript{6}CuLi\textsubscript{3} (T\textsubscript{2} )</td>
<td>Fine, low density</td>
<td>Coarse, high density</td>
</tr>
<tr>
<td>(b) Coarse T\textsubscript{1}</td>
<td>Low density</td>
<td>Low density</td>
</tr>
<tr>
<td>6. Width of \textdelta\textsuperscript{'} precipitate free zones, \mu m</td>
<td>0.02 - 0.04</td>
<td>0.1 - 0.12</td>
</tr>
</tbody>
</table>

**Table 2**

TENSILE PROPERTIES OF THE Al-Li ALLOY PLATES IN LONGITUDINAL DIRECTION

<table>
<thead>
<tr>
<th>Property</th>
<th>Underaged (T3) alloy plate</th>
<th>Peakaged (T8E51) alloy plate</th>
</tr>
</thead>
<tbody>
<tr>
<td>0.2% Yield Strength, MPa</td>
<td>410</td>
<td>490</td>
</tr>
<tr>
<td>Ultimate Tensile Strength, MPa</td>
<td>474</td>
<td>567</td>
</tr>
<tr>
<td>Uniform Elongation, %</td>
<td>6.3</td>
<td>5.4</td>
</tr>
<tr>
<td>Total Elongation or Strain to Fracture, %</td>
<td>6.3</td>
<td>5.4</td>
</tr>
<tr>
<td>Work Hardening Exponent (n)</td>
<td>0.072</td>
<td>0.053</td>
</tr>
</tbody>
</table>
Jeol JSM 840 scanning electron microscope to record fracture features under cyclic loading.

3. RESULTS AND DISCUSSION

3.1 Cyclic Stress Response (CSR) Behaviour

The cyclic stress response (CSR) curves of the underaged alloy showing the variation of average stress amplitude ($\Delta\sigma/2$) with elapsed fatigue cycles (N) as well as normalized fatigue cycles ($N/N_f$) are shown respectively in Figs. 2a and 2b. Though the tests were conducted at close intervals of strain, for the sake of clarity, CSR curves obtained at a few select strain amplitudes have been included. Data in Fig.2 show that the underaged alloy exhibits significant cyclic hardening in the initial few cycles (up to $N/N_f = 0.1$). The extent of hardening was found to increase with increasing strain amplitude and reaches up to 8% of the corresponding average stress amplitude ($\Delta\sigma/2$) of the first cycle. Following initial hardening, the alloy exhibits cyclic stability at lower strain amplitudes ($\Delta\varepsilon_r/2 < 7.5 \times 10^{-3}$) and cyclic softening at higher strain amplitudes ($\Delta\varepsilon_r/2 > 7.5 \times 10^{-3}$).

The nature of the cyclic stress response of the UA alloy is similar to that of the PA alloy. The PA alloy (Fig. 3) also exhibits initial cyclic hardening followed by cyclic stability at lower strain amplitudes and cyclic softening at higher strain amplitudes. However, there are two noticeable differences in the CSR behaviour of the alloy as a result of the change in the aging condition. Firstly, the extent of initial hardening is considerably higher in UA condition (4-8% of the corresponding $\Delta\sigma/2$ of the first cycle) as compared to that of the PA alloy (1-4% of $\Delta\sigma/2$ of the first cycle). Secondly, the plastic strain amplitude corresponding to the occurrence of transition in CSR behaviour, i.e., the change from cyclic stability to cyclic softening, occurs at a higher value of the strain amplitude in case of the UA alloy ($\Delta\varepsilon_r/2 = 7.5 \times 10^{-3}$ or the corresponding $\Delta\varepsilon_p/2 = 2.3 \times 10^{-3}$) as compared to that in the PA alloy ($\Delta\varepsilon_r/2 = 6 \times 10^{-3}$ or the corresponding $\Delta\varepsilon_p/2 = 7 \times 10^{-4}$).

The initial cyclic hardening observed in the first few cycles, at all strain amplitudes, might be attributed to the increase in dislocation density and to the increased degree of dislocation-dislocation interactions as well as interactions with the dispersoid particles and strengthening precipitates.\(^{18, 23}\) The increase in dislocation density as well as the extent of dislocation interactions increases with increase in strain amplitude resulting in higher extent of cyclic hardening. Higher extent of initial cyclic hardening in case of the UA plate as compared to the PA plate is attributable to higher work hardening capability of the alloy plate in the underaged condition as revealed by the uniaxial tensile properties (see data in Table 2).
The cyclic stability followed by the initial cyclic hardening, occurring mostly at lower strain amplitudes, is attributable to stable dislocation structure, especially a constant dislocation density in the form of nearly the same dislocation cell wall size and their thickness with elapsed fatigue cycles. The cyclic softening observed in the case of alloys AA 8090 composition in both UA and PA conditions can be attributed to the shearing of \( \delta' \) precipitates, the major strengthening phase of the alloy. The \( \delta' \) precipitates are highly coherent with the Al matrix and are ordered. The shearing of \( \delta' \) precipitates, predominates after a critical shear strain (or the corresponding shear stress) is attained. The extent of softening increases with higher degree of precipitate shearing, which occurs with increasing strain amplitude. The \( \delta' \) precipitates are much finer in the underaged condition and are prone to shearing at lower stress levels as compared to the peakaged alloy.

### 3.2 Cyclic Stress-Strain (CSS) Behaviour

The cyclic stress-strain (CSS) data, corresponding to half-life, of the alloy in the UA and PA tempers are shown in Figs. 4a and b, respectively. For the purpose of comparison, the tensile stress and strain data of the alloys are also included in Figs. 4a and b. The relative position of cyclic stress-strain curve with that of the monotonic tensile stress-strain curve is traditionally used to bring out cyclic hardening or cyclic softening that a material exhibits during fatigue loading. The data in Figs. 4a and b indicate that the alloy plates in the two temper conditions exhibit distinctly different behaviour. In case of the UA alloy, the cyclic stress-strain curve lies above the tensile stress-strain curve. This indicates cyclic hardening. On the other hand, the PA alloy exhibits an average stress amplitude that is lower than the corresponding tensile stress value (Fig. 4b; except for a small region at lower strain amplitudes), indicating that the CSS curve lies below the tensile stress-strain curve. This shows that the PA alloy exhibits cyclic softening. Hence, the CSS behaviour of the UA alloy is quite opposite to that of the PA alloy. This is despite the fact that both the alloy plates exhibit similar nature of cyclic stress response (CSR) behaviour, i.e., initial cyclic hardening followed by cyclic stability at lower strain amplitudes.
and cyclic softening at higher strain amplitudes. The reason for this opposite CSS behaviour of the alloys in two temper conditions could be traced to the extent of cyclic hardening that occurs in the first few fatigue cycles ($N/N_f < 0.1$) and subsequent cyclic stability at lower strain amplitudes and the extent of cyclic softening at higher strain amplitudes. From the results obtained, it is evident that in case of the UA alloy, the extent of initial cyclic hardening predominates and compensates more than the extent of subsequent cyclic softening that occurs at the higher strain amplitude region. The net cyclic hardening effect is also true at the lower strain amplitude region as the alloy exhibits cyclic stability after initial hardening. This leads to net cyclic hardening in the UA alloy plate, as noted in Fig. 4a. On the other hand, in case of the PA alloy the cyclic softening that occurs at higher strain amplitudes ($\Delta \varepsilon_p/2 = 6 \times 10^{-3}$ or $\Delta \varepsilon_p/2 = 7 \times 10^{-4}$) predominates over the initial cyclic hardening. The degree of cyclic softening is of the order of 4-8% of the peak or the saturated stress amplitude, while the degree of hardening is of the order of 1-4% of the average stress amplitude of the first cycle 18. Hence, the peakaged alloy in the longitudinal direction experiences net cyclic softening, especially at higher strain amplitudes. The initial hardening followed by cyclic stability at lower strain amplitudes results in a small degree of net cyclic hardening.

A comparison of the CSS constants ($K'$ and $n'$ values) with the corresponding values for monotonic tensile loading ($K$ and $n$) is also an indication whether an alloy exhibits cyclic hardening or cyclic softening 24. The situation $K' > K$ and $n' < n$ indicates cyclic hardening, while the opposite trend indicates cyclic softening. The CSS data of plastic strain amplitude and average stress amplitude at half-life of the alloy in the UA and PA conditions are given in Fig.5. The data are analysed in terms of the power-law relationship of the form $\Delta \sigma/2 = K'(\Delta \varepsilon_p/2)^{n'}$. A power-law fit of the data corresponding to the UA condition could be described by a single power-law equation and yields cyclic strength coefficient, $K' = 594$ MPa and cyclic work hardening exponent, $n' = 0.054$. The value of $n'$ is significantly lower than the value of $n$ (0.072), indicating cyclic hardening. The CSS data of the PA alloy, given
along with the data of the UA alloy in Fig. 5, show a distinctly different behaviour. Unlike the UA alloy, the PA alloy shows bilinear CSS power-law relationship with two sets of power-law constants ($K'_v = 821$ MPa and $n'_v = 0.088$ at lower strain amplitudes and $K_v = 716$ MPa and $n_v = 0.077$ at higher strain amplitudes) and a transition at $\Delta \epsilon / 2 = 6 \times 10^{-4}$. The cyclic power-law constants, $K'_v$ and $n'_v$ are higher at lower strain amplitudes. The values of $n'_v$ in the two regions of lower and higher strain amplitudes are significantly higher than the monotonic tensile n value (0.053), indicating net cyclic softening. Hence, the comparison of the power-law constants with those corresponding to the monotonic tensile data indicates similar trends in CSS behaviour as that evinced by the relative positions of the stress-strain curves obtained under cyclic and monotonic tensile loading.

### 3.3 Fatigue Life Power-Law Relationships

The variation of fatigue life has been analysed using different power-law relationships, viz., plastic strain amplitude at half-life, $\Delta \epsilon_p / 2$ (known as Coffin-Manson power-law relationship), average stress amplitude at half-life, $\Delta \sigma / 2$ (known as Basquin-like power-law relationship), average plastic strain energy per cycle, $W_p$ (known as Halford-Morrow power-law relationship) as:

$$\Delta \epsilon_p / 2 = \epsilon'_f (2N_f)^c$$

$$\Delta \sigma / 2 = \sigma'_f (2N_f)^b$$

$$\Delta W_p = W'_f (2N_f)^\beta$$

where $\epsilon'_f$ and $\sigma'_f$ are the fatigue ductility and strength coefficients, respectively; $c$ and $b$ are the fatigue ductility and strength exponents, respectively. $W'_f$ and $\beta$ are constants and $b$ is correlated to $b$ and $c$ as $\beta = b + c$. The value of $\Delta W_p$ is derived from the fatigue parameters using the relationship:

$$\Delta W_p = \Delta \epsilon_p \Delta \sigma (1 - n')/ (1 + n')$$

The variation of fatigue life (in terms of number of reversals to failure, $2N_f$) with $\Delta \epsilon_p / 2$, $\Delta \sigma / 2$ and $\Delta W_p$ as per the power-law relationships (1) to (3) for the underaged alloy is shown in Figs. 6a-c. The cyclic parameters used in Fig. 6 correspond to the values obtained at half-life. For the sake of comparison, data pertaining to the peakaged alloy plate are included in Fig. 6. All the three power-law relationships are found to be bilinear in nature for both UA and PA alloy plates. The power-law constants derived are listed in Table 3. The values of $\beta$ derived using eqn.3 and as per the relation $\beta = b + c$ match well for UA and PA alloys. The data in Table 3 reveal that the fatigue constants corresponding to the lower strain amplitudes or higher fatigue life region (referred to as hypo-transition region) are significantly higher than those pertaining

<table>
<thead>
<tr>
<th>Power-law Constant</th>
<th>Underaged alloy</th>
<th>Peakaged alloy</th>
</tr>
</thead>
<tbody>
<tr>
<td>Fatigue Ductility Coefficient, $\epsilon'_f$</td>
<td>4.07</td>
<td>5.50</td>
</tr>
<tr>
<td>Fatigue Ductility Exponent, $c$</td>
<td>-1.71</td>
<td>-1.15</td>
</tr>
<tr>
<td>Fatigue Strength Coefficient, $\sigma'_f$, MPa</td>
<td>631</td>
<td>887</td>
</tr>
<tr>
<td>Fatigue Strength Exponent, $\beta$</td>
<td>-0.06</td>
<td>-0.093</td>
</tr>
<tr>
<td>$W'_f$, MJ/m$^3$</td>
<td>$2.9 \times 10^4$</td>
<td>$9.3 \times 10^3$</td>
</tr>
<tr>
<td>$\beta$ (from Eq.3)</td>
<td>-1.4</td>
<td>-1.17</td>
</tr>
<tr>
<td>$\beta$ (derived as $\beta + c$)</td>
<td>-1.23</td>
<td>-1.24</td>
</tr>
</tbody>
</table>

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3.4. Comparison of Low Cycle Fatigue Resistance

In order to bring out the difference in the LCF resistance of the two alloys quantitatively, the fatigue lives at two select values (one corresponding to the hypo-transition region and the other to the hyper-transition region) of the three fatigue parameters ($\Delta \varepsilon_p/2$, $\Delta \sigma/2$ or $\Delta W_p$) are obtained from Fig. 6 and are listed in Table 4. The fatigue life of the UA alloy at any particular value of the chosen fatigue parameter (either $\Delta \varepsilon_p/2$ or $\Delta \sigma/2$ or $\Delta W_p$) is more than 40% lower than the corresponding fatigue life of the PA alloy. A comparison of fatigue resistance on the basis of plastic strain amplitude ($\Delta \varepsilon_p/2$) reveals that the UA alloy is inferior to that of the PA alloy. At the two strain levels considered for comparison (Table 4), fatigue lives of the UA alloy are nearly half of those of the PA alloy. The comparison of the fatigue resistance based on $\Delta W_p$ of the two alloy plates is similar. On the other hand, fatigue life comparison on the basis of stress amplitude ($\Delta \sigma/2$) shows unusually large disparity. This is especially so at hyper-transition region, where the UA alloy exhibits a fatigue life that is more than an order of magnitude lower than that of the PA alloy. Such behaviour is attributable to the variation in strength levels in the two temper conditions. The UA alloy by virtue of its temper condition and near-recrystallised grain structure possesses nearly 20% lower strength (yield as well as ultimate tensile strength; see data in Table 2), which the strain based analysis does not take into account. Similar observation in the fatigue life analysis was made in an earlier study where the PA alloy showed wide variation in fatigue life as well as strength level with test direction. The above observation reveals that all the three power-law relationships based on the fatigue parameters $\Delta \varepsilon_p/2$, $\Delta \sigma/2$ or $\Delta W_p$ point to similar trend in the LCF resistance. Based on any of these three parameters, the PA alloy plate exhibits higher fatigue resistance consistently. However, the magnitude of the variation in the fatigue resistance is distinctly different for the three parameters. While the parameters $\Delta \varepsilon_p/2$ and $\Delta W_p$ show lesser predominance of the aging condition, the LCF resistance looks vastly different if the comparison is based on the stress amplitude ($\Delta \sigma/2$). It is, however, more appropriate to employ Eq. 3 as this relationship is based on the average plastic strain energy per
cycle ($\Delta W_p$) that unifies the effect of plastic strain and cyclic stress. With $\Delta W_p$ as the basis, the UA alloy plate possesses fatigue life that is 42% lower in the hypo-transition region and 60% lower in the hyper-transition region as compared to the PA alloy plate.

The low fatigue resistance of the underaged alloy in comparison with the peakaged alloy can primarily be attributed to three factors. Firstly, the nature of the grain structure. The UA alloy is nearly recrystallised while the PA alloy is near unrecrystallised. In an earlier study conducted on a high copper and low lithium containing Al-Li alloy AA2020, Srivatsan et al.\(^\text{12}\) have demonstrated that the alloy with higher degree of unrecrystallised grain structure exhibits considerably higher (200-300%) fatigue lives than the recrystallised microstructure. Secondly, the PA alloy contains higher amount of Al$_2$CuMg (S$'$) precipitates as compared to the sparse amounts of S$'$ precipitates in the UA alloy\(^\text{8}\) (see Table 1). The higher content of S$'$ precipitates in the PA alloy is a result of 2-3% tensile prestrain that the alloy has been subjected to in the solution treated condition before it was artificially aged to peak strength. The incorporation of such a thermomechanical treatment introduces sites in the form of dislocation networks for the heterogeneous nucleation of S$'$ precipitates. These precipitates having an altogether different crystal structure (orthorhombic) as compared to the matrix aluminium (face centered cubic), promote cross slip, hence hindering slip localization\(^\text{30, 31}\). On the other hand, the UA alloy has not been subjected to cold work and was not aged to peak strength, making the sluggish reaction of S$'$ precipitation incomplete. Hence, the higher content of S$'$ precipitates of the peakaged alloy too contributes to the improved low cycle fatigue resistance. In addition to these two factors, the S$'$ precipitates in the UA condition are finer and particle shearing occurs more rapidly and readily. The above three factors together explain the lower fatigue resistance of the UA alloy.

### 3.5. Fatigue Analysis Based on Total Plastic Strain Energy till Fracture ($W_f$)

Alloys subjected to fatigue dissipate most of the irrecoverable plastic strain energy as heat, with other minor forms of energy release being vibration and acoustic emission. Hence, only a small fraction of the total plastic strain energy imparted to the material actually contributes to the increase in the extent of immobile line defects and surface damage. The total plastic strain energy till fracture ($W_f$), given by the sum of areas of all the hysteresis loops before failure, is widely used to quantify the materials’ resistance to fatigue damage. The value of $W_f$ is computed as:

$$W_f = \Delta W_p N_f$$

<table>
<thead>
<tr>
<th>Select value of chosen fatigue parameter</th>
<th>Number of reversals to failure ($2N_f$)</th>
<th>Underaged plate</th>
<th>Peakaged plate</th>
<th>($2N_f$)$<em>{UA}$ / ($2N_f$)$</em>{PA}$</th>
</tr>
</thead>
<tbody>
<tr>
<td>$\Delta\varepsilon_p/2 = 5 \times 10^{-4}$ (hypo-transition)</td>
<td>2236</td>
<td>3756</td>
<td>0.60</td>
<td></td>
</tr>
<tr>
<td>$\Delta\varepsilon_p/2 = 5 \times 10^{-3}$ (hyper-transition)</td>
<td>128</td>
<td>292</td>
<td>0.44</td>
<td></td>
</tr>
<tr>
<td>$\Delta\sigma/2 = 375$ MPa (hypo-transition)</td>
<td>3924</td>
<td>8659</td>
<td>0.45</td>
<td></td>
</tr>
<tr>
<td>$\Delta\sigma/2 = 465$ MPa (hyper-transition)</td>
<td>30</td>
<td>501</td>
<td>0.06</td>
<td></td>
</tr>
<tr>
<td>$\Delta W_p = 0.5$ MJ/m$^3$ (hypo-transition)</td>
<td>2467</td>
<td>4241</td>
<td>0.58</td>
<td></td>
</tr>
<tr>
<td>$\Delta W_p = 0.5$ MJ/m$^3$ (hyper-transition)</td>
<td>83</td>
<td>206</td>
<td>0.40</td>
<td></td>
</tr>
</tbody>
</table>
where $\Delta W_p$ is average plastic strain energy per cycle (as per the Eq. 4). Substituting in the equation 5, the cyclic parameters $\Delta \epsilon_p$ and $\Delta \sigma$ as per the fatigue power-law relationships of Eqs. 1 and 2 and $n' = b/c$, one can obtain:

$$W_f = 2\epsilon' \sigma_f' \left[ (c - b)/(c + b) \right]^{2N_f} (b + c)$$

(6)

The above equation is a power-law relationship with an exponent of $1 + b + c$. The variation of $W_f$ with the number of reversals to failure for the UA alloy is shown in Fig. 7. The data reported earlier for the peakaged alloy are also included in Fig. 7. The variation of $W_f$ with $2N_f$ follows the same trend for both the alloys, with nearly matching values of the power-law exponents in the two regions of lower (hyper – transitional region) as well as higher (hypo – transition region) fatigue life. The alloys exhibit a peak in $W_f$ at a fatigue life ($N_p$) of 500 cycles.

In an earlier analysis 17, 18, the authors proposed that the occurrence of peak in the $W_f - 2N_f$ relationship indicates fatigue degradation at both the lower and the higher strain amplitude regions. In an ideal situation, the alloys would have shown no transition in the fatigue power-law relationships of Eqs. 1 and 2, if the alloys exhibited constant $W_f$ at all the strain amplitudes. A constant $W_f$ at all strain amplitudes implies $b' + c' = -1$, where $b'$ and $c'$ denote fatigue life exponents corresponding to this ideal situation. This is schematically shown in Fig. 8.

The fatigue degradation at lower fatigue lives or higher strain amplitudes (corresponding to the hyper-transition region of the Coffin-Manson plots) is attributable to $\delta'$ precipitate shearing and the resulting planar slip and strain localization. It is generally agreed in case of Al-Li alloys that ductility under

![Image](trans-indian-inst-met-vol-57-no-2-april-2004_p190.png)

**Fig. 7**: Variation of total plastic strain energy till fracture ($W_f$) of the underaged and peakaged alloy plates. The data clearly show that the peakaged alloy plate exhibits superior LCF resistance and that both alloy plates exhibit bilinearity in the power-law relationship.

**Fig. 8**: (a) Schematic showing the variation in $W_f$ with fatigue life with respect to the observed transition in the power-law relationships of $\Delta \epsilon_p/2$ (b) and $\Delta \sigma/2$ (c). The ideal behaviour is represented by the solid lines, while the actual behaviour is shown by the dashed lines.
monotonic tensile ($\varepsilon_0$) and cyclic ($\varepsilon'_0$) loading conditions are lower due to planar slip and the related strain localization \(^4\). On the other hand, the fatigue degradation at higher fatigue lives, or lower strain amplitudes (corresponding to hypo-transition region of Coffin-Manson plots), is possibly due to environmental effects. The longer test duration at lower strain amplitudes and the nature of the test environment, i.e. laboratory air with ~30% humidity, could cause reduction in low cycle fatigue resistance of the alloy. Srivatsan et al. \(^{12}\) have shown in an earlier study on a low Li (1.21 wt.%) and high Cu (4.45 wt.%) alloy AA2020 that environmental degradation in LCF resistance occurs at low strain amplitudes, but only at elevated temperatures (433K). However, the present Al-Li AA8090 alloys contain higher amounts of lithium (2.2 wt.%) and, hence, could possibly be prone to environmental degradation even at ambient temperatures.

### 3.6. Transition in Fatigue Life Power-Law Relationship

Several studies, including those on Al-Li alloys, have identified that bilinearity in fatigue life power-law relationships occurs due to (i) change in the mode of deformation, as observed by transmission electron microscopy \(^{11,12,18,22,32-35}\) or as reflected by a change in the $n'$ values \(^{14,18,22,36,37}\), (ii) change in the fracture mode \(^{9,14,15,29,38,39}\) and (iii) extent of environmental degradation, especially at elevated temperatures \(^{17,40,41}\). The transition observed in the fatigue life power-law relationships in case of the UA alloy plate has been studied in light of the above micromechanisms, which have been summarized in Ref. \(^{42}\).

Fractographic studies conducted on the fractured specimens of the UA plate tested under different LCF loading conditions have revealed that there is a gradual change in the fracture morphology with

![Composite picture showing a gradual change in the fracture mode with applied strain amplitude. The values of $\Delta \varepsilon'/2$ corresponding to the individual fractographs (a-h) are indicated in the central Coffin-Manson plot. The fractographs depict (a-c) transgranular shear fracture at lower strain amplitudes, (d) mixed fracture mode comprising transgranular shear and quasi-cleavage fracture at intermediate strain amplitudes and (e-h) quasi-cleavage fracture with microdimples at higher strain amplitudes.](image)
increasing strain amplitude. Figure 9 is the composite picture showing the fracture features of the specimens failed at different strain amplitudes as indicated in the Coffin-Manson plot. The alloy fails by transgranular shear fracture with distinct fatigue striations (Figs. 9a-c) at strain amplitudes below the transition. The crack extension at these strain amplitudes occurs with plastic intrusions and extrusions leading to predominant fatigue striations. With increase in strain amplitude, the fracture mode gradually changes to a relatively lower energy mixed-mode fracture morphology comprising quasi-cleavage or faceted fracture with isolated regions of microdimples (Figs. 9e-h). Hence, in case of the UA alloy, the fracture at both lower and higher strain amplitudes is predominantly governed by the intragranular microstructural features of the alloy (especially the major and minor strengthening precipitates). Fracture mode at intermediate strain amplitudes (near-transition) is mixed comprising both transgranular shear fracture and quasi-cleavage/faceted fracture with transgranular microdimples (Fig. 9d). In view of the above observations, the transition in the fatigue life power-law relationships is attributable to the change in fracture morphology with applied strain amplitude.

The peakaged alloy plate too exhibited a gradual change in the deformation mode as well as deformation-assisted fracture morphology with applied strain amplitude $\Delta \varepsilon_p/2 = 1.1 \times 10^{-4}$ in the three different in-plane directions (L, L+45° and LT), the alloy failed in transgranular shear fracture mode at lower strain amplitudes (Fig. 10a). The fracture mode gradually changed to a lower energy ductile intergranular fracture with increase in strain amplitude. Figs. 10 b and c depict the fracture mode at higher strain amplitudes and show the grain boundary fracture as well as the network of voids at the high angle grain boundaries. The nature of fracture mode change was found to be same for all the three test directions. In the absence of apparent strain localization at lower strain amplitudes (the alloy exhibits cyclic stability), the locally originated microcracks coalesce to form major cracks, resulting in higher energy transgranular shear fracture. On the other hand, at higher strain amplitudes intense shearing of major strengthening precipitates leads to strain localization (the alloy exhibits cyclic softening) and, also, concentration of strain along the soft, d’ precipitate-free, high angle grain boundaries. This has led to void nucleation and their coalescence at coarse, grain boundary $\text{Al}_3\text{CuLi}_2(T_2)$ precipitates, resulting in lower energy ductile intergranular fracture.

Fig. 10: Fractographic features of LCF tested specimens of peakaged alloy plate showing (a) transgranular shear fracture at lower strain amplitudes ($\Delta \varepsilon_p/2 = 1.1 \times 10^{-4}$) and (b,c) ductile intergranular fracture at higher strain amplitudes ($\Delta \varepsilon_p/2 = 6.6 \times 10^{-3}$). The change in fracture mode with applied strain amplitude is gradual with the alloy plate showing mixed-mode of fracture at the intermediate strain amplitudes $\Delta \varepsilon_p/2 = 1.1 \times 10^{-4}$.
What has been brought out above is that the aging condition and the associated microstructural features strongly influence the nature of change in the deformation and the deformation-assisted fracture morphology. Studies on Al-Li alloys have revealed that the transition in the fatigue life power-law relationships, especially the Coffin-Manson power-law relationship, occurs with change in the deformation and/or fracture morphology.

4. CONCLUSIONS

(i) The Al-Li alloys of AA 8090 composition in the underaged (UA, T3) and peakaged (PA, T8E51) temper conditions show similar cyclic stress response behaviour, which varies with applied strain amplitude. Cyclic hardening in the initial few cycles is followed by cyclic stability at lower strain amplitudes and cyclic softening at higher strain amplitudes.

(ii) The alloys exhibit bilinearity in the fatigue power-law relationships with a clear transition. The underaged alloy is characterised by a single CSS power-law relationship, while the peakaged alloy showed distinct change in the cyclic work hardening behaviour with higher n' value at lower strain amplitudes as compared to that at higher strain amplitudes.

(iii) The underaged alloy showed inferior low cycle fatigue resistance as compared to the peakaged alloy. The low LCF resistance of the underaged alloy is attributable to the combined effects of lower degree of unrecrystallised grain structure, lower content of S precipitates and finer size of d' precipitates.

(iv) The transition in the fatigue power-law relationships is attributable to the changes in the fracture morphology. The alloy in the underaged temper exhibits higher energy transgranular shear fracture with fatigue striations at lower strain amplitudes. With increase in strain amplitude, the fracture mode changes to a relatively lower energy quasi-cleavage/faceted fracture with microdimples.

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