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THE METALLURGY OF INDUSTRIAL Al-Li ALLOYS

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Abstract

All industrial unrecrystallized Al-Li-X alloys exhibit a strong tendency to intergranular failure leading to a slaty fracture of the specimens, associating transgranular shear and intergranular delamination in L-LT planes.

The level of alkali impurities measured in industrial metal quality cannot explain this fracture behaviour.

In the opposite, the studies conducted on bending specimens indicate that the slip heterogeneity within the grains induce the delamination through a severe and detrimental shear localization at grain boundaries. This behaviour is still possible in strongly aged tempers because of the heterogeneity of the precipitation.

In order to avoid or reduce the stress localization at grain boundaries, increasing the homogeneity of deformation has become the guideline for the metallurgy of industrial Al-Li-X alloys, illustrated by the following three examples.

An efficient means to reduce or even suppress slatiness consists in underaging alloys like 2091 or CP 276 which offer a high potential of matrix coprecipitation. These underaged tempers maximize short transverse properties as well as the resistance against exfoliation corrosion. In addition, toughness and crack growth resistance reach higher levels.

Recrystallization is another way to induce better microscopic slip homogenization. In that case, the slaty fracture is no more possible. The best combination of properties for damage tolerant alloys (2024 T3 replacement) is obtained when associating the recrystallized structure and underaging. This applies well to 2091 sheets where the choice of the T8X temper ensures a high and stable level of toughness.

On profiles, which are characterized by an unrecrystallized structure, the sensitivity to texture of underaged or peakaged tempers may be too high for some alloys. A slight overaging may reduce the microscopic tendency to slip localization. Though the delamination is not totally suppressed, this kind of treatment applied to 8090 extrusions leads to more homogeneous properties which can be combined with improved resistance towards exfoliation corrosion for medium strength alloys replacement (2214 T6 - 7XXX T73).

Introduction

AI-Li alloys offer attractive combination of properties (Ref ; 1) for an extensive use in aircraft structures : reducing the density while increasing other properties like stiffness or fatigue resistance could already lead to structural weight reduction of up to 15 % on commercial airliners (Ref : 2).

However, the coherency of the Al3Li precipitate, which is present in all industrial Al-Li alloys, induces a strong tendency towards strain localization (Ref : 3). In a first step, we shall observe that this behaviour induces the macroscopic fracture mode and infer different possible ways to improve the properties of Al-Li X alloys. These means will be then illustrated by industrial achievements obtained by Pechiney in order to overcome the previously mentioned weakpoints and meet the demand for the replacement of existing 2XXX and 7XXX series alloys.
Fracture mechanisms in industrial AI-Li alloys

Fracture aspect of AI-Li-X alloys
Binary Al-Li alloys suffer from extreme brittleness when submitted to artificial ageing and their fracture is intergranular. The beneficial effect of additional precipitation through copper and magnesium alloying and the interest of zirconium as dispersoid have been established (Ref : 3). The following industrial Al-Li-Cu-(Mg)-Zr alloys have been developed on that basis : 8090 - 2091 - CP 276 - 809i and 2090 (Ref : 4, 5, 6).

However, even in these alloys, the fracture of unrecrystallized specimens may be characterized by an unusual slaty surface associating transgranular shearing and intergranular fracture along the L - LT plane (fig 1). The following studies have been conducted at the Voreppe R and D Center in order to better understand the reasons of this intergranular delamination and hence find ways to improve the properties of the Al-Li-X alloys.

Effect of alkali impurities
Two series of measurements have been made in order to evaluate the effect of these elements which were suspected to be of specific importance for Al-Li-X alloys.

- Surface measurements (Ref : 7)
  30 mm thick flat extruded bars made out of 8090 with the following bulk composition :
  Al 2.5 Li, 1.3 Cu, 1.0 Mg, 0.04 Fe, 0.03 Si, 0.11 Zr (wt%), and 5 µg/g Na, heat treated 12 H at 190°C, have been tested in short transverse direction. Tensile specimens have been pulled in vacuum (10^-9 - 10^-10 Torr), characterized by Auger Analysis and further Scanning Electron Micrography. The fracture surface is intergranular. Auger Energy Scanning (from 0 to 600 and 900 to 1400 ev) detected no peak due to Na or K : The concentrations of these alkali elements within the two extreme atomic layers are respectively below 1 and 0.25 atomic %. This low level of alkali contamination at grain boundaries suggests that Al-Li-X alloys of industrial quality do not exhibit any specific problem of sodium or potassium embrittlement.

- Sodium Distribution at grain boundaries.
  A highly Sodium contaminated Al-Li alloy was cast in the 150 kg unit of Voreppe into a 200 mm diameter billet, homogenized, extruded into a 100 x 13 mm² rectangular flat bar, solution heat treated at 530°C, cold water quenched, stretched (2%) and aged 20 H at 190°C. The bulk composition is the following : Al 2.8 Li, 1.5 Cu, 0.9 Mg, 0.10 Zr (wt %), 40 µg/g Na. The precise study of the grain boundary segregation reveals sodium rich nodules, but no evidence of any continuous film (Ref : 8).

The low level of alkali impurities and their tendency to form nodules (rather than continuous layers at the grain boundaries) indicate that they should not be suspected to cause the intergranular fracture of industrial alloys.

Fracture mechanisms (Ref : 9)
Further experiments have been conducted on 8090 13 mm thick extrusions. The composition is the following : Al 2.6 Li, 1.35 Cu, 1.0 Mg, 0.11 Zr, 0.04 Fe, 0.03 Si (wt%) and the grain structure is unrecrystallized. Rectangular test bars (L : 70 X LT : 13 X ST 13 mm³) are machined in the L direction. One L-ST plane was submitted to polishing and chromic etching, and observed during the testing. The specimens was bent as indicated in fig. 2 .

Though the shearing calculated by the elastic linear fracture mechanics is close to zero in the L-LT plane including the crack tip, the crack propagates as indicated in fig. 2 by transgranular shear and intergranular delamination. This Mode II delamination starts at the crack tip and propagates in the L - LT plane.

The explanation of such a paradox, confirmed on 4 points bending (where oxy = 0 in the central part) is given in fig. 3. Different stages of the bending are observed on the L-ST polished plane. Streaks were made in order to reveal the local strains during the bending. Fig. 3a shows the surface without any applied stress.
In fig. 3b the bending has started and the strain localizes first within a limited number of crystallographic planes (usually 1, sometimes 2 per grain). Intense shear accumulates at the grain boundaries. After further bending this limited number of transgranular slip systems is no more able to accomodate the strains at the grain boundaries and a secondary non crystallographic slip system appears in some grains. Sigmoidal steps are characteristic of this second slip system which is already well developed in 2 of the grains in Fig. 3b (indicated with arrows). After additional bending, a macroscopic shear crack starts from the surface of the specimen while the strains develop within the crystallographic and the secondary slip systems. When this latter system extends to the grain boundary, the local shear may become too important, especially in the vicinity of the macroscopic crack tip. The delamination may initiate there and extend along the grain boundaries in order to release the shear stresses. The steps generated by the released slip systems appear on the intergranular fracture surface.

In Fig. 3c arrows indicate the stress release due to delamination on the L-ST plane. As observed ahead of some delaminations, direct measurements of local shear (made on the streaks) indicate that levels as high as 50 % can be reached before failure in the vicinity of the grain boundaries. Fig 3d summarizes the different steps of the fracture.

These observations induce the following conclusions:
- There is no evidence of any intrinsic brittleness of the grain boundary which may be sheared up to 50 % before fracture.
- The substructure does not seem to interfere with the fracture process.
- The slaty fracture results from microscopic stress localization at the grain boundary induced by heterogeneous transgranular slips.
- This heterogeneous deformation which induces high local stress concentration is probably a limit to the use of the whole hardening potential of the alloy.
- Hence, increasing the homogeneity of the deformation should be a key to optimizing the properties.

Further understanding through grain size and ageing variations
Two alloys A and B were chilled cast into 43 mm diameter billets with the following compositions in major elements : Li 2.60 ± 0.05 %, Cu 1.30 ± 0.05 %, Mg 0.95 ± 0.05 %. The grain sizes are respectively 150 µm and 50 µm for alloys A and B. They have been homogenized 24 H 533°C, extruded into rectangular flat bars (19 X 2.7 mm² section) at 430°C, solution heat treated 2 H 540°C, cold water quenched, stretched 2 % and aged. The fracture aspect is the following:

<table>
<thead>
<tr>
<th>Ageing</th>
<th>12 H 190°C peak ageing</th>
<th>48 H 220°C severe overageing</th>
</tr>
</thead>
<tbody>
<tr>
<td>A</td>
<td>Transgranular + delamination</td>
<td>Transgranular + delamination</td>
</tr>
<tr>
<td>B</td>
<td>Transgranular + delamination</td>
<td>Transgranular</td>
</tr>
</tbody>
</table>

The peakaged alloy A exhibits intense shear at grain boundaries (Fig. 4a) associated with large intergranular fracture surfaces. Even with the reduced grain size leading to limited dislocations pile up, the slip localization is strong enough to cause slaty failure on alloy B aged in the same conditions. In the opposite case, the overaged alloy B leads to a fully transgranular failure, though the grain boundary is weakened by the extreme ageing.

This behaviour, associated with the lack of slip bands in the matrix and with the stress release in the wide P.F.Z. of the overaged alloy (Fig. 4b) demonstrates that the slip behaviour is more important than the grain boundary itself during the slaty fracture of fibered Al-Li-X alloys.
The other interesting feature is the beneficial effect of finer grain size, which helps to reduce the stresses at the tip of the dislocation pile ups (U.A and P.A tempers) or the stresses localized within the PFZ (O.A. tempers : comparison between alloys A and B).

Hence grain size must be well controlled. However, this parameter is not sufficient to master the fracture of Al-Li-X alloys and the severe overaged tempers previously mentioned are too brittle in S.T. direction.

Other efficient solutions had to be found to solve that problem.

**Ways to improve the homogeneity of deformation**

**Alloy composition-underageing**

This first means consists in improving the efficiency of the coprecipitation introduced in addition to $\delta'$(Al3Li) as a barrier against slip localization. Maximizing the volume fraction of potentially favorable precipitates has been a guideline for alloy selection at Pechiney leading to the following alloys, with increased hardening potential: (Ref : 10)

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Intergranular precipitation</th>
<th>Matrix precipitation</th>
</tr>
</thead>
<tbody>
<tr>
<td>8090</td>
<td>$\delta$, S, $T_2$</td>
<td>$\delta'$, GPB, S', S</td>
</tr>
<tr>
<td>2091</td>
<td>S, $T_2$</td>
<td>$\delta'$, GPB, S', S, ($T_2$)</td>
</tr>
<tr>
<td>CP 276</td>
<td>$T_1$, $T_2$</td>
<td>$\delta'$, $T_1$, $T_2$, GPB, S', S</td>
</tr>
</tbody>
</table>

with: $\delta$ (Al-Li) $T_2$ (Al3Cu(LiMg)3) $T_1$ (Al2CuLi) $\delta'$ (AlLi) GPB S' S (Al3CuMg) $T^*$ (AlCuLi-Al2Cu)

The underlined precipitates are the major ones.

Fig. 5 shows that the increased coprecipitation leads to better toughness-strength combinations when going from 8090 to 2091 and CP 276 alloys.

Bending tests have also been performed on the 3 alloys.

The results are reported in Fig. 6 for underaged, peakaged and overaged tempers.

The key points are the following:

- Only strongly hardenable alloys with a slight reduction of $\delta'$ volume fraction limit or even suppress delamination. (2091 and CP 276)
- The optimum behaviour is obtained for these alloys with underaged tempers.

To understand why peakaged or overaged tempers are less effective to homogenize deformation and suppress delamination, we observed a 2091 extrusion (solution heat treated, cold water quenched, stretched 2 %, aged 12 h 220°C) in its L-ST plane after polishing and chromic etching and further bending (Fig. 7). This etching reveals the heterogeneous hardening due to the preferential precipitation of S' on the slip bands generated by the stretch performed after quenching. The grain consists of alternate stronger and less hardened areas. The weaker areas may act as a guide for further heterogeneous deformation as observed in Fig. 7: intense slips generated by the bending are concentrated in the less hardened areas (arrows) after bending.

Instead of fully homogenizing the deformation through increased coprecipitation, it looks as if peak-and overaged tempers keep some slip heterogeneity because of the heterogeneity of their precipitation itself.

Underaged tempers which do not develop this hardening heterogeneity to such an extend may be more efficient in avoiding stress localization and delamination.

As a matter of fact underaged tempers apply well to industrial productions.
- 2091 has been designed to meet the 2024 damage tolerant replacement goal. Its underaged T8X temper (12 H 135°C) is characterized by a δ' (Al3Li) + GPB matrix precipitation hardening associated with a fine distribution of the S(Al2CuMg) phase at the grain boundary (Fig. 8). With a strength level close to that of 2024-T 351, it maximizes short transverse properties and resistance towards exfoliation corrosion (Table 1) as well as the toughness and the crack propagation resistance. This latter result is partially related to a strong crack closure effect and the gap between 2091 and 2024 is reduced when increasing the R ratio. This tendency is close to that of other Al-Li Alloys (Ref : 11 and 12) for both crack propagation resistance and S.T. properties.

Table 1 : Effect of ageing on 2091 38 mm thick plates

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Density</th>
<th>Aging</th>
<th>L YS</th>
<th>UTS</th>
<th>E1</th>
<th>LT YS</th>
<th>UTS</th>
<th>E1</th>
<th>ST YS</th>
<th>UTS</th>
<th>E1</th>
<th>Exco Rating</th>
</tr>
</thead>
<tbody>
<tr>
<td>2091</td>
<td>2.57</td>
<td>CA, 12 H, 235°C</td>
<td>377</td>
<td>443</td>
<td>7.0</td>
<td>372</td>
<td>430</td>
<td>6.5</td>
<td>353</td>
<td>409</td>
<td>3.3</td>
<td>P-EA</td>
</tr>
<tr>
<td></td>
<td></td>
<td>PA, 12 H, 190°C</td>
<td>473</td>
<td>523</td>
<td>8.5</td>
<td>430</td>
<td>495</td>
<td>8.0</td>
<td>383</td>
<td>466</td>
<td>3.5</td>
<td>EB-EC</td>
</tr>
<tr>
<td></td>
<td></td>
<td>UA, 12 H, 135°C</td>
<td>394</td>
<td>470</td>
<td>12.5</td>
<td>331</td>
<td>443</td>
<td>15.0</td>
<td>266</td>
<td>421</td>
<td>7.0</td>
<td>P-EA</td>
</tr>
<tr>
<td>2024</td>
<td>2.78</td>
<td>T351</td>
<td>385</td>
<td>500</td>
<td>16</td>
<td>345</td>
<td>475</td>
<td>16</td>
<td>320</td>
<td>435</td>
<td>8.0</td>
<td>EB-ED</td>
</tr>
</tbody>
</table>

YS and UTS = MPa
EI = %

- CP 276 offers another example of the association between underaging and strongly hardenable alloy.
CP 276 reaches optimum properties with an underaged temper consisting essentially of δ' (Al3Li) + T' (AlCuLi) + T1 (Al2CuLi) matrix hardening with T1 (Al2CuLi) at grain boundaries. Good toughness and resistance towards exfoliation corrosion result from that structure with a strength level comparable to that of 7XXX T6 - T7X series. (Table 2)

Table 2 : Properties measured on CP 276 T 851 extrusions

<table>
<thead>
<tr>
<th>ALLOY</th>
<th>TEMPER</th>
<th>CP 276</th>
<th>7075</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>T 851</td>
<td></td>
<td>T 651</td>
</tr>
<tr>
<td>Density</td>
<td>2.58</td>
<td></td>
<td>2.80</td>
</tr>
<tr>
<td>Modulus</td>
<td>80 G Pa</td>
<td></td>
<td>72 G Pa</td>
</tr>
<tr>
<td>Tensile (L direction)</td>
<td>YS [MPa]</td>
<td>UTS [MPa]</td>
<td>EI [%]</td>
</tr>
<tr>
<td></td>
<td>525-600</td>
<td>600-655</td>
<td>5-10</td>
</tr>
<tr>
<td>Toughness (P direction)</td>
<td>38 - 43</td>
<td></td>
<td></td>
</tr>
<tr>
<td>Exco rating</td>
<td>Surface</td>
<td>Core</td>
<td></td>
</tr>
<tr>
<td></td>
<td>N - EA</td>
<td>EA - EB</td>
<td></td>
</tr>
</tbody>
</table>

Texture control
An other means to induce better homogeneity of deformation consists in recrystallizing the material. In that case, the strong texture of the pancake fibered structure (e.g. Brass T. in the core of rolled semi-products) is suppressed and hence more slip systems can be activated within each grain (Ref : 13). The slaty surface no longer exists and the isotropy of properties is increased. Optimal properties are again obtained for underaged tempers and ageing temperature is adjusted to ensure the stability of properties during long term usage.
Damage tolerant sheets made of the 2091 alloy illustrate such kind of choice. In that case, the selection of the ageing treatment is governed by the need for a stable toughness. The evolution of toughness versus ageing time is given in (fig. 10). The time axis takes into account the equivalence established for strength: \(2X\) hours at 135°C \(\Rightarrow X\) hours at 150°C. The superior behaviour of the lower ageing temperature (135°C) designed the heat treatment 12 H 135°C for the 2024 replacement goal.

**Overaged tempers**

The tight control of texture previously mentioned for damage tolerant sheets is not always possible. A typical example is given by profiles which are unrecrystallized in order to take advantage of the "press-effekt". In this case the texture varies strongly from point to point even within a given section. Underaged or even peakaged alloys like 8090 may suffer from excessive properties variation due to their sensitivity to texture. Overaging, mentioned by some authors on sheets, (Ref: 14) can be applied in that case in order to spur the dense precipitation of \(S\) and/or \(S'\) throughout the whole matrix.

Though some tendency to strain localization remains, this kind of overaging applied to 8090 provides reduced properties variations, and increased resistance towards exfoliation as compared to the peak ageing (Fig. 11). It applies well to the medium strength replacement goal (e.g. 2214 T 651).

**Conclusion**

- No intrinsic brittleness of the grain boundaries of Al-Li X alloys has been observed.
- The particular fracture mode of Al-Li X alloys is essentially related to microscopic slip localization within the grains.
- The search for improved homogeneity of deformation through composition, grain size control, texture adjustment and ageing succeeded in reaching high properties which compare well with the ones of the existing alloys used on aircraft structures.

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Ref : 12 C.J. Peel, D. Mc Darmaid, B. Evans, Considerations of critical factors for the design of aerospace structures using current and future Al-Li alloys; Westec 87 to be published by A.S.M.


Ref : 14 P.J. Gregson and H.M. Flower; Microstructural Control of Toughness in Al-Li alloys; Acta. Met., Vol 33, No3, 1985
Fig. 1: 8090 peakaged extrusion: fracture surface of a T-L CT specimen. S.E.M.

\[ \sigma_{xy} = \frac{3P (y^2 - z^2)}{4b c^3} \]
\[ \sigma_{xx} = \frac{3P (1 - x)y}{2 b c^3} \]

\( b = \text{LT width} \quad \text{and} \quad 2c = \text{residual ST thickness} \)

Fig. 2: Bending test and fracture mode of Al-Li-X alloys

1. Localization within limited slip systems
2. Transgranular shear.
   - Increased strain within slip systems.
   - Secondary non-cristallophic shear system.
3. Transgranular shear.
   - Delamination: shear release at Grain Boundaries.

3d: Schematic fracture mode

Fig. 3: Different stages of the bending test.
8090 peakaged extrusions L-ST plane chromic etching.

4a: peakaged
4b: overaged

Fig. 4: Bending test 8090 fine grain extrusion L-ST plane. S.E.M.
Fig. 5: Yield strength-toughness combination of Al-Li-X alloys under production at Pechiney (L direction) 100 X 13 mm² section extrusions.

Fig. 6: Bending test of 8090, 2091 and CP 276 alloys at different heat treatments (T 351 + aging) 13 mm thick plates.

Fig. 7: Surface of a specimen made out of 2091 overaged at 220°C. L-ST plane. Chromic etching. The specimen has been submitted to bending after attack.

Fig. 8: Precipitation within 2091 (T8X temper)

Fig. 9: Precipitation within CP 276 (T8S1 temper)

Fig. 10: Dependence of toughness on aging conditions: 2091 recrystallized sheets (1.6 mm thick).

Fig. 11: Influence of aging on anisotropy of 8090 extrusions.