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**DEVELOPMENT OF MICROSTRUCTURES WITH IMPROVED
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DEVELOPMENT OF MICROSTRUCTURES WITH IMPROVED CRYOGENIC TOUGHNESS THROUGH LOCAL VARIATIONS IN STRESS STATE: ALUMINUM-LITHIUM ALLOYS

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ABSTRACT

Microstructurally-induced changes in the local stress state (triaxial constraint) and their effect on fracture-toughness behavior are examined at ambient and cryogenic temperatures in an Al-Li-Cu-Zr alloy, processed in the form of 12.7 mm-thick "*naturally laminated*" plate containing aligned-weak interfaces and 1.6 mm-thin unlaminated sheet. It is shown that marked improvements in long-transverse (L-T) toughness can be achieved in the plate material at cryogenic temperatures by promoting through-thickness delamination along these interfaces, which relaxes local constraint and promotes a fracture-mode transition from *global* plane strain to *local* plane stress. Conversely, in thin sheet material, the absence of such interface delamination leads to a reduction in toughness with decrease in temperature, consistent with the greater degree of crack-tip constraint.

INTRODUCTION

In addition to intrinsic properties such as crystal structure, grain size, slip character and second-phase particle distribution, the fracture behavior of a material is greatly influenced by the local stress-state under which failure occurs [1-4]. For example, the load necessary to generate a given plastic strain in a structure can be higher in the presence of a notch compared to an unnotched sample, which leads to strain-controlled failures in ductile materials being associated with notch strengthening. Conversely, tensile stress-controlled cracking processes occur more readily at notches, because crack-opening stresses, σ_{yy} , are elevated by triaxial constraint [1]. Consequently, transgranular cleavage cracking in mild steels is greatly enhanced by triaxial stresses; similarly, boron-doped Ni₃Al intermetallic alloys exhibit a predominantly intergranular fracture in the presence of a notch, compared to transgranular tearing-type failures under uniaxial tensile loading conditions [5].

The state of stress within the deformation or fracture process zone in metallic materials can be strongly influenced by thickness of the specimen [1-4]. In thin samples where the extent of plasticity is comparable to specimen thickness, the spread of local yielding restricts constraint at the crack tip ($\sigma_{yy,max} \approx \sigma_y$, the yield strength, and through-thickness stresses, $\sigma_{zz} \approx 0$), thereby promoting primarily plane-stress conditions. Such effects are largely confined to the surface in very thick samples; maximum σ_{yy} crack-tip stresses in the interior may approach over $5\sigma_y$ in the presence of strain hardening [6,7]. Such plane-strain conditions represent a more severe state of stress in triaxial tension. Measured (crack-initiation) fracture-toughness values in plane strain (maximum constraint) are therefore generally significantly lower than those evaluated under plane-stress (minimal constraint) conditions [1,4].

This concept can be used to promote toughness microstructurally in thick-plate material where the degree of local constraint can be varied in specific orientations by laminating composite materials or simply by thermomechanical treatments to produce "natural laminates" in monolithic materials [8,9]. If the interfaces between individual laminates are sufficiently weak, delamination can occur due to the high crack-tip constraint, thereby enhancing toughness in crack-divider orientations by relaxing local constraint; toughening also can be induced in the crack-arrester orientation from marked crack deflection (Fig. 1). Examples of such toughening from aligned, weak interfaces include the design of laminated mild steels, Damascus swords and ceramic composites [8-11]. However, in the present work, the concept is used to promote exceptional cryogenic toughness in naturally laminated microstructures of an advanced Al-Li-Cu-Zr alloy 2090-T8.

MATERIALS AND EXPERIMENTAL PROCEDURES

The material studied was a hot rolled ingot-metallurgy Al-2.86Cu-2.06Li-0.12Zr (in wt.%) alloy 2090, processed as 12.7 mm-thick plate and 1.6 mm-thin sheet. Both forms were solution treated at 549°C, cold-water quenched, stretched 3-6% and aged at 163°C to peak-strength [12,13]; aging practices differ slightly for plate and sheet, and are designated as T81 and T83, respectively. Mechanical properties at 298 and 77 K are listed in Table I.

Grains structures in 2090-T81 plate were unrecrystallized, coarse and pan-cake shaped, with dimensions $\sim 50 \mu\text{m}$ thick, $\sim 500 \mu\text{m}$ wide and elongated several mm in the rolling direction (Fig. 1); in contrast, fine, equiaxed recrystallized subgrain structures were seen in T83 sheet [13]. Both microstructures were hardened by uniform matrix distributions of coherent, ordered δ' (Al_3Li) spheres, T_1 (Al_2CuLi) and Θ' -like (Al_2Cu) plates and β' (Al_3Zr) dispersoids; in addition, Fe- and Cu-rich intermetallics and constituent phases were present in

the matrix and particularly along high-angle grain boundaries. Limited heterogeneous precipitation along grain and subgrain boundaries was also seen, resulting in ~50 to 100 nm-wide δ' -precipitate free zones (PFZs) [12,13].

Fracture-toughness K_{Ic} tests on T81 plate were conducted at 298 K (ambient air), 196 K (dry ice/ethanol) and 77 K (liquid N_2) temperatures, using fatigue-precracked, 12 mm-thick, 4-point single-edged-notch bend (SEN(B)) and double-cantilever beam (DCB) specimens, in the long-transverse (L-T) and short-transverse (S-L) orientations, respectively [12]. Corresponding K_{Ic} behavior in 1.6 mm-thin sheet was assessed using 50 mm-wide compact C(T) specimens machined in the L-T orientation to measure $K_{R-\Delta a}$ resistance curves (R-curves); for simplicity, toughness values were defined at maximum load. Crack initiation and growth was monitored using d.c. electrical potential and unloading elastic-compliance methods. Additional tests were performed for thicknesses of 0.8, 0.34 and 0.24 mm, fabricated by machining away the surface layers of precracked C(T) samples of T83 sheet. Crack-path profiles and fracture surfaces were examined using optical and scanning electron microscopy.

RESULTS AND DISCUSSION

Behavior in Thick Plate

The fracture-toughness behavior of the 12.7 mm-thick plate is shown in Fig. 2a; a marked improvement in L-T toughness is apparent with decrease in temperature, concurrent with increases in longitudinal strength, tensile elongation and work-hardening exponent (Table I). Behavior, however, is dependent upon orientation; K_{Ic} values in the S-L orientation marginally decrease with temperature and are ~50% lower than in the L-T orientation at both 298 and 77 K. Fracture surfaces in the L-T orientation are similar at all temperatures and exhibit a coarse, transgranular-shear mechanism of failure with limited void coalescence around constituent phases. However, also apparent are significant through-thickness intergranular delaminations, the incidence and depth of which is considerably greater at 77 K (Figs. 2b,c). S-L fractures, conversely, are highly planar and involve lamellar-type splitting along grain boundaries.

The principal effect of increased delamination splitting in the L-T plate at cryogenic temperatures is to relieve through-thickness stresses (triaxial constraint), such that deformation conditions at the crack tip are transformed from global plane strain (fully constrained) to a series of plane-stress (unconstrained) ligaments, with a corresponding reduction in crack-tip tensile stresses and an increase in local ductility. Consequently, the nominally flat (plane-strain) mode of fracture at 298 K (Fig. 2b) becomes a series of individually slant (plane-stress) failures at 77 K (Fig. 2c), where the major spacings between delaminations are of the order of

the plastic-zone size (Table I). As plane-stress fracture-toughness values invariably exceed those in plane strain, such "natural" delamination promotes increased toughness, in this case concurrent with an elevation in strength and tensile ductility.

The through-thickness splitting in 2090-T81 plate results from poor short-transverse properties induced by thermomechanical processing; deformation textures sustained during rolling and stretching, the presence of brittle constituent phases along high-angle boundaries, segregation of lithium to grain boundaries, heterogeneous grain-boundary precipitation and consequent formation of δ' -PFZs are all salient contributing factors. In addition, because of the unrecrystallized, pan-cake shaped nature of the grain structure, many grain boundaries are oriented perpendicular to the rolling plane; failures along these planes occur by brittle, intergranular decohesion with less than 1% elongation. Such failure modes are inevitably stress-controlled, and are thus promoted at low temperatures in thick sections by higher local crack-tip stresses from increased flow stress and higher strain-hardening rate.

Behavior in Thin Sheet

The 1.6 mm 2090-T83 sheet, conversely, which also displays higher strength, ductility and strain hardening at 77 K (Table I), shows no such splitting and fails at both 77 and 298 K by ductile intergranular void-coalescence. Consequently, despite having similar composition and hardening precipitates as in the plate material, toughness now *decreases* with temperature. The absence of splitting is attributed to the lack of through-thickness stresses and the more equiaxed, *recrystallized* grain structures in the thin sections. Here, a plane-stress (slant) mode of fracture at 298 K becomes mixed (slant plus flat) mode at 77 K, as diminished plasticity at cryogenic temperatures is insufficient to fully relax constraint at the crack tip (plastic-zone sizes at both 298 and 77 K are again comparable with the width of the slant-fracture section - Table I, Fig. 3b).

The transition to a (partly) plane-strain mode of fracture at low temperatures can be suppressed simply by reducing specimen thickness to preserve plane-stress conditions (Fig. 3b). Thus, for T83 sheet mechanically thinned to thicknesses below 0.34 mm, the fracture mode remains 45° slant even at 77 K, such that toughness now *increases* with decreasing temperature (Fig. 3a), presumably due to the enhanced strength, ductility, modulus (E) and strain hardening.

It should be noted, however, that concurrently room-temperature toughness decreases with thickness (B), consistent with Cottrell's model for plane-stress fracture [1,14], where K_{Ic} is proportional to $\sqrt{E\sigma_y B}$. This is apparent in Fig. 4 by replotting the sheet toughness data as a function of B and 77 and 298 K; it is clear that, for a given temperature, there is a maximum in toughness corresponding to the largest thickness that yields a fully unconstrained (slant) mode

of failure at that temperature. Above this critical thickness, fully plane-stress conditions no longer prevail at the crack tip (except at the surface), and toughness decreases with increasing B until it approaches K_{Ic} ; below this thickness, deformation conditions remain totally unconstrained and toughness decreases with $\sqrt{B}$. This implies that for crack-divider delamination toughening in thick plate, excessive through-thickness splitting or fabricating laminates of ply thicknesses below the critical dimension can be counterproductive; conversely, limited splitting may also be ineffective if it is insufficient to relax local constraint.

The marked improvement of the toughness of the 2090 alloy with decreasing temperature is also aided by several other factors. In fact, alternative explanations for improved cryogenic toughness of Al-Li plate alloys have related the behavior to the grain-boundary solidification of low-melting point impurities [15], to increased homogeneity of plastic deformation from increased strain hardening [16] and reduced strain localization in closer and more widely spaced slip bands [17]. While these mechanisms may provide contributions to enhanced toughness at any given stress state, they fail to account for the observed differences in the temperature-dependence of toughness between sheet and plate material. Moreover, they fail to explain why toughening is specific to orientations in the rolling plane (L-T, T-L and L+45°) and not to short-transverse (S-L, S-T) orientations. In fact, increased toughness from the freezing of grain-boundary liquid phases at low temperatures would be expected to be more pronounced in the latter orientations, where failures are totally intergranular.

CONCLUSIONS

It has been shown that processing peak-aged Al-Li-Cu-Zr 2090-T8 alloy in the form of a naturally-laminated thick plate, containing planar weak interfaces (or as thin (<0.34 mm thick) sheet) can lead to remarkable improvements in L-T fracture toughness at cryogenic temperatures; such microstructures result in unconstrained (plane-stress) modes of failure even at low temperatures. Although for a given stress state, the temperature-dependence of toughness is a function of the variation in strength, slip homogeneity and work-hardening rate, the delamination toughening approach, however, does have limitations; there is an optimum delamination spacing, and akin to behavior in advanced composites, properties in laminated microstructures can be highly anisotropic.

ACKNOWLEDGEMENTS

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Table I. Mechanical Properties of Al-Li-Cu-Zr Alloy 2090-T8 at 298 and 77 K†

Yield Strength σ_y (MPa)		U.T.S. (MPa)		% Elongation (on 25 mm)		Work-Hardening Exponent n		Fracture Toughness (MPa√m)		Plastic Zone Size* (mm)	
298K	77K	298K	77K	298K	77K	298K	77K	298K	77K	298K	77K
<u>12.7 mm T81 plate</u>											
552	587	589	642	11	14	0.06	0.15	36 •17	51 (L-T) 15 (S-L)	0.7	1.2
<u>1.6 mm T83 sheet</u>											
505	568	549	674	6.8	8.0	0.05	0.08	43 34 30 22	30 (1.6 mm) 35 (0.74 mm) 40 (0.34 mm) 42 (0.24 mm)	0.8 0.7 0.5 0.3	0.4 0.6 0.8 0.9

† Tensile properties in the longitudinal direction.

* Computed using $(1/2\pi)(K_c/\sigma_y)^2$

XBB 8912-10552

Fig. 1: Micrograph of the grain structure in 2090-T81 plate, showing the orientation of test samples in the L-T (crack divider, T-S (crack arrester) and S-T (crack delamination) orientations.

XBL 8710-4220

Fig. 2: (a) Temperature dependence of K_{Ic} in 12.7 mm thick 2090-T81 plate, and optical micrographs of the fracture modes of (b) 298 K and (c) 77 K.

XBB 892-1269

Fig. 2(b,c)

XBL 899-3219A

Fig. 3: (a) Variation in (L-T) toughness of sheets with temperature, showing a strong dependence with specimen thickness (crack-tip constraint), and (b) corresponding changes in fracture mode.

XBB 8912-10557

Fig. 3(b)

XBL 8912-4380

Fig. 4: Influence of specimen thickness on fracture toughness at 298 and 77 K, for near peak-aged 2090-T83 sheet (L-T orientation).

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