

Fatigue of aluminium–lithium alloys

K. T. Venkateswara Rao and R. O. Ritchie

Aluminium–lithium alloys are a class of low density, high strength, high stiffness monolithic metallic materials that have been identified as prime candidates for replacing 2000 and 7000 series aluminium alloys currently used in commercial and military aircraft. In this review, the cyclic fatigue strength and fatigue crack propagation characteristics of aluminium–lithium alloys are reviewed in detail with emphasis on the underlying micromechanisms associated with crack advance and their implications to damage tolerant design and lifetime computations. Compared with traditional aerospace aluminium alloys, results on the fatigue of binary Al–Li, experimental Al–Li–Cu, and near commercial Al–Li–Cu–Zr and Al–Li–Cu–Mg–Zr systems indicate that alloying with Li degrades the low-cycle fatigue resistance, though high-cycle fatigue behaviour remains comparable. The alloys, however, display superior (long crack) fatigue crack growth properties, resulting from a prominent role of crack tip shielding, principally due to deflected and tortuous crack path morphologies, induced by the shearable nature of coherent δ' precipitates, crystallographic texture, and anisotropic grain structures. Environmental fatigue resistance is comparable with 2000 series alloys and better than 7075-type alloys. The accelerated growth of small fatigue cracks, strong anisotropy, poor short-transverse properties, and a sensitivity to compression overloads are the principal disadvantages of Al–Li alloys.

IMR/237

© 1992 The Institute of Materials and ASM International. Dr Venkateswara Rao and Professor Ritchie are with the Center for Advanced Materials, Lawrence Berkeley Laboratory, and the Department of Materials Science and Mineral Engineering, University of California, Berkeley, CA, USA

Introduction

Designers of modern commercial and military aerospace vehicles and space launch systems are constantly in search of new materials with lower density, and higher strength, stiffness, and stability at elevated or cryogenic temperatures. To meet these challenges, much effort has been directed toward developing intermetallics, ceramics and composites as structural and engine materials for future applications.^{1–6} However, for structural airframes, age hardened 2000 and 7000 series aluminium alloys have long been preferred for civil and military aircraft by virtue of their high strength/weight ratio, though the use of composite materials, particularly for secondary structures, is rapidly increasing. Nearly 75% of the structural

weight of the Boeing 757-200 aeroplane is comprised of plates, sheets, extrusions, and forgings of 2024, 2224, 2324, 7075, and 7150 aluminium alloys;⁷ typical components include, body frames, fuselage (frames, skins, and substructures), wings (stiffeners, ribs, stringers, and skins), and horizontal stabilisers.

Today, conventional aluminium alloys face strong competition from emerging composite technologies, particularly in the structural aerospace market; hybrid materials based on organic and metal matrices with fibre, whisker, or particulate ceramic reinforcements offer impressive combinations of strength, stiffness, and high temperature resistance.^{2,8–12} In addition, aramid polymer-reinforced aluminium alloy (ARALL) laminates,^{13–15} fabricated by resin bonding aramid fibres sandwiched between thin aluminium alloy sheets, show exceptional promise as fatigue resistant materials. In the light of these advances, the aluminium industry has recently introduced a new generation of aluminium–lithium alloys,* by additionally incorporating ultra-low density lithium into traditional aluminium alloys; these alloys represent a new class of light weight, high modulus, high strength, monolithic structural materials, which are cost effective compared with the more expensive composites.^{16–28}

Despite possible limitations regarding specific stiffness and high temperature stability, Al–Li alloys enjoy several advantages over composite materials. Economically, Al–Li alloys are typically only three times as expensive as conventional aluminium alloys, whereas competing hybrid materials can be up to 10–30 times more expensive.²⁸ Secondly, Al–Li alloy fabrication technology is generally compatible with existing manufacturing methods such as extrusion, sheet forming, and forging to obtain finished products. Finally, they offer considerably higher ductility and fracture toughness properties compared with most metal matrix composites.^{9–12}

Historically, efforts to design lithium containing aluminium alloys date from the early 1920s, though concerns over their poor ductility and toughness halted production and use in the 1960s; Refs. 26, 27, 29, and 30 provide an excellent review of these developments. Escalation in fuel costs during the 1970s rekindled research on the Al–Li system with prospects of building more fuel efficient aircraft; design tradeoff studies had shown that reducing density was the optimal route.^{30–32} Each weight per cent of lithium added to aluminium lowers the density about 3% and improves the elastic modulus by nearly 6% for Li additions of up to 4%.^{21,33} Accordingly, direct substitution of Al–Li alloys (con-

* Although this class of alloys contain alloying elements such as copper, magnesium, and zirconium in addition to aluminium and lithium, the term aluminium–lithium (or Al–Li) alloys is generally used to encompass all aluminium alloys containing greater than 0.5 wt-% lithium.

Table 1 Chemical composition limits for major alloying elements in advanced aluminium–lithium alloys and traditional aluminium alloys, wt-%

Alloy*	Li	Cu	Mg	Zn	Zr	Mn	Ot	Other	Al
IM 2090	1.9–2.6	2.4–3.0	0.0–0.25	...	0.08–0.15	Bal.
IM 2091	1.7–2.3	1.8–2.5	1.1–1.9	0.25	0.04–0.16	Bal.
IM 8090	2.2–2.7	1.0–1.6	0.6–1.3	...	0.04–0.16	Bal.
IM 8091	2.4–2.8	1.6–2.2	0.5–1.2	...	0.08–0.16	Bal.
IM W049 (Weldalite)	1.3	5.0	0.4	...	0.1	Bal.
IM 2020	1.1	0.5	...	0.2Cd	Bal.
MA IN-905XL	1.5	...	4.0	0.8	1.1C	Bal.
RSP 644-B	3.0–3.2	0.8–1.1	0.4–0.6	...	0.4–0.6	Bal.
IM 2124	...	4.5	1.5	0.25	Bal.
IM 7150	...	2.1	2.2	6.16	0.13	Bal.
IM 7075	...	2.0	2.6	5.8	Bal.
PM 7091	...	1.7	2.5	6.2	0.35	0.4Co	Bal.

*IM ingot metallurgy; MA mechanical alloying; PM powder metallurgy; RSP rapid solidification processing.
†As oxide.

taining ~3%Li)* for traditional aluminium alloys can yield ~11% weight savings; complete redesign of aircraft utilising the full potential of Al–Li alloys could save up to ~17% by weight.³⁰

Early studies by Sanders *et al.*^{21,34} on binary Al–Li alloys showed that problems of low ductility and toughness could be traced to inhomogeneous slip and strain localisation, resulting from coherent δ' (Al_3Li) particle hardening in the matrix, nucleation and growth of grain boundary δ (AlLi) precipitates, and the formation of δ' -precipitate free zones (PFZs). Subsequent developments in alloy design and processing have attempted to reduce this inhomogeneous deformation mode by incorporating additional alloying elements and modifying thermomechanical treatments. As a result, advanced commercial Al–Li–X–Y–Z alloys today exhibit strength–toughness combinations which are comparable, and often superior, to traditional aluminium alloys 2024, 2124, 7075, 7150 and 7475, particularly at cryogenic temperatures.^{35–42} Details of the principal alloys of commercial interest, their designations, and nominal chemical composition limits are summarised in Table 1.

Since advanced Al–Li alloys are principally intended for safety critical structural aerospace applications, their durability and damage tolerance performance are of considerable importance. Over the past decade, a number of investigations have focused on various aspects of fatigue crack initiation and crack propagation in Al–Li alloys. It is therefore the objective of the present paper to review these cyclic properties with emphasis on the underlying micro-mechanisms; reviews on the crack growth mechanisms under monotonic loading may be found in Refs. 43 and 44. Crack initiation and crack growth behaviour is examined as a function of microstructural, mechanical, and environmental factors. It is concluded that in general the fatigue behaviour of Al–Li alloys is not significantly different from traditional aluminium alloys, though crack propagation kinetics of long cracks are retarded by more pronounced crack tip shielding† from such mechanisms as crack deflection, crack path tortuosity, and consequent roughness

induced, crack closure mechanisms. In fact, the observed variation in fatigue resistance of Al–Li alloys with respect to microstructure, crack size, loading sequence, and product form are largely a manifestation of the degree to which shielding is promoted or restricted in the different microstructures under various loading conditions.

Microstructural features

Strengthening in Al–Li alloys is predominantly from hardening by the nucleation and growth of one or more second phase particle distributions, precipitated from a supersaturated solid solution. This is achieved by aging naturally (at room temperature) or artificially (at an elevated temperature below the metastable solvus line) following solution treatment and quenching the alloy from the single phase field. By controlling the aging time and temperature, microstructures are optimised for targeted mechanical properties in the various tempers. Generally, these temper designations are referred to as (a) underaged (T3), synonymous with low strength and fine uniform distributions of coherent hardening precipitates in the matrix, (b) peak aged (T6, T8), with maximum strength and partially coherent precipitates, and (c) overaged (T7), with slightly reduced strength and coarse incoherent precipitates. Many published papers^{16–20,47–56} have reviewed extensively the solid state phase transformations and precipitation sequences in Al–Li alloys; salient microstructural features are summarised below.

Binary Al–Li alloys

Binary alloys of aluminium containing less than ~1.7%Li are essentially random solid solutions of aluminium and lithium. The elevated flow stress compared with pure aluminium is due to short range order and dislocation interactions between solute atoms and associated strain fields. For lithium contents greater than 1.7%, the alloys are hardened by long range ordering through the precipitation of

* All compositions are in weight percent unless otherwise stated.
† Crack tip shielding mechanisms act to impede crack advance by lowering the local stress intensity actually experienced at the crack tip.^{45,46} Such mechanisms, which act principally in the crack wake,

include transformation and microcrack toughening in ceramics, crack bridging in composites, and crack closure during fatigue crack growth. Closure mechanisms are described in detail in the section on 'Fatigue crack propagation', below.

Table 2 Summary of chemical compositions, heat treatments, microstructures and mechanical properties of selected binary aluminium–lithium alloys (Ref. 57)

Alloy, wt-%	Heat treatment*	Microstructural features	Tensile properties†		
			σ_y , MN m ⁻²	σ_u , MN m ⁻²	Elongation, %
Al–0.7Li	SHT + CWQ naturally aged	Li in solid solution with short range order	45	65	26
Al–2.5Li underaged	SHT + CWQ naturally aged	δ' (Al ₃ Li) 2–5 nm in diameter	67	157	33
Al–2.5Li peak aged	SHT + CWQ aged 1 h at 473 K	δ' (Al ₃ Li) 15–20 nm in diameter, δ (AlLi)	185	220	2.6

*SHT solution heat treatment; CWQ cold water quench.

† σ_y and σ_u are the yield and ultimate tensile strengths, respectively.

ordered ($L1_2$ structure), metastable, and spherical δ' particles, in α (fcc) Al–Li solid solution matrix. Owing to the low particle/matrix misfit strains ($\sim -0.12\%$), the precipitates tend to remain coherent with the matrix and retain their spherical morphology, even for particle diameters as large as 300 nm.⁵⁴ In the naturally aged condition (or underaged tempers), microstructures show fine homogeneous distributions of δ' particles (2–5 nm in diameter) in the matrix and at grain boundaries. With increased aging time and/or temperature, the matrix δ' precipitates coarsen, and equilibrium δ particles nucleate heterogeneously along grain boundaries, thereby resulting in the formation of Li-depleted δ' -PFZs surrounding δ . Improvements in strength in these alloys are primarily due to the resistance of δ' particles to dislocation motion (from order hardening, elastic moduli differences, and misfit coherency strains between the matrix and precipitates); the extent of hardening is proportional to both the volume fraction and size of δ' precipitates. However, deformation in binary alloys is highly localised along narrow slip bands and within soft PFZs, which can promote premature crack nucleation, and contribute to poor ductility and toughness of Al–Li alloys.³⁴ Typical chemical compositions, heat treatments, and mechanical properties of selected binary alloys are summarised in Table 2.

Ternary Al–Li–Cu and Al–Li–Mg alloys

To reduce the inhomogeneous mode of deformation, ternary elements have been incorporated in Al–Li

alloys to enable the nucleation of additional hardening precipitates and to modify the coherency strains between δ' particles and the matrix. For example, Al–Li alloys containing 2–3%Cu are strengthened by semicoherent T_1 (Al₂CuLi) and θ' -like (Al₂Cu) plates in the matrix. Precipitation of the equilibrium T_2 (Al₆CuLi₃) phase is sometimes observed along grain and subgrain boundaries particularly in peak and overaged tempers. Moreover, in addition to the commonly observed spherical morphology, the ordered δ' -phase also forms surrounding the θ' plates. In Al–Li–Mg alloys containing less than 2%Mg, δ' is the only hardening phase; however, magnesium reduces the solid solubility of lithium in aluminium and hence increases the volume fraction of δ' precipitates. At longer aging times and for alloys with Mg contents greater than 2%, Al₂MgLi precipitates in the matrix and along grain boundaries.

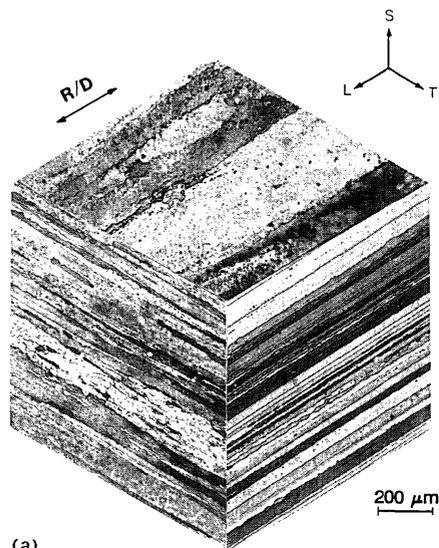
Commercial alloys

Commercial Al–Li alloys are alloyed with Cu, Mg, and dispersoid forming elements such as Zr, Mn, and Cr that form a fine dispersion of intermetallic phases to reduce the tendency for localised deformation. Zirconium is generally preferred because it results in fine, as cast grain structures and better corrosion properties. In addition, to reduce the deleterious effects of δ' -PFZs on ductility and toughness, wrought Al–Li sheet and plate products are given a 2–6% permanent stretch before aging; this promotes uniform intragranular precipitation and thus suppresses

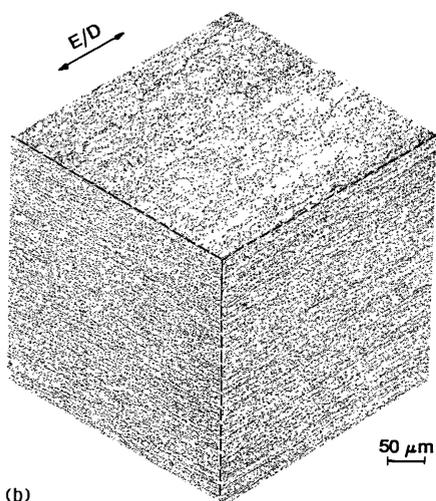
Table 3 Experimental thermomechanical treatments utilised to obtain various microstructures in commercial aluminium–lithium and aluminium alloys (Refs. 16–20, 58, 59, 68)

Alloy	Condition	Heat treatment*
2090–T81	Peak aged	SH, CWQ, 6% stretch, aged 24 h at 436 K
2091–T351	Underaged	SH, CWQ, 2% stretch, naturally aged
2091–T8	Peak aged	SH, CWQ, 2% stretch, aged 10 h at 408 K
8090–T351	Underaged	SH, CWQ, 3% stretch, naturally aged
8090–T8	Peak aged	SH, CWQ, 3% stretch, aged 16 h at 463 K
8091–T351	Underaged	SH, CWQ, 3% stretch, naturally aged
8091–T8	Peak aged	SH, CWQ, 3% stretch, aged 16 h at 463 K
W049–T3	Underaged	SH, CWQ, 3% stretch, naturally aged
W049–T8	Peak aged	T3 condition + aged 24 h at 433 K
2020–T651	Peak aged	SH, CWQ, 1.5% stretch, aged 24 h at 422 K
IN–905XL	Peak aged	SH, CWQ, aged 24 h at 443 K
644–B	Near peak aged	SH, HWQ, aged 16 h at 405 K
2124–T351	Underaged	SH, CWQ, 2% stretch, naturally aged
7150–T651	Peak aged	SH, CWQ, 2% stretch, aged 100 h at 394 K
7075–T651	Peak aged	SH, CWQ, 2% stretch, aged 24 h at 394 K
7091–T7E69	Overaged	SH, CWQ, aged 24 h at 394 K + 4 h at 436 K

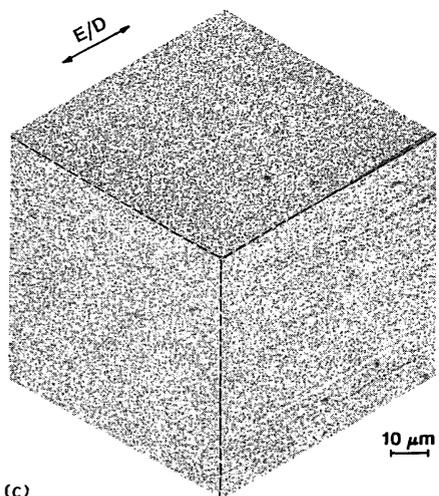
*ST solution treat; CWQ cold water quench at 298 K; HWQ hot water quench at 333 K.



(a)



(b)



(c)

a ingot metallurgy; b rapid solidification; c mechanical alloying; R/D and E/D refer to the rolling and extrusion directions

1 Three dimensional optical micrographs of grain structures in commercial Al-Li alloys (Ref. 58)

heterogeneous grain boundary precipitation phenomena.

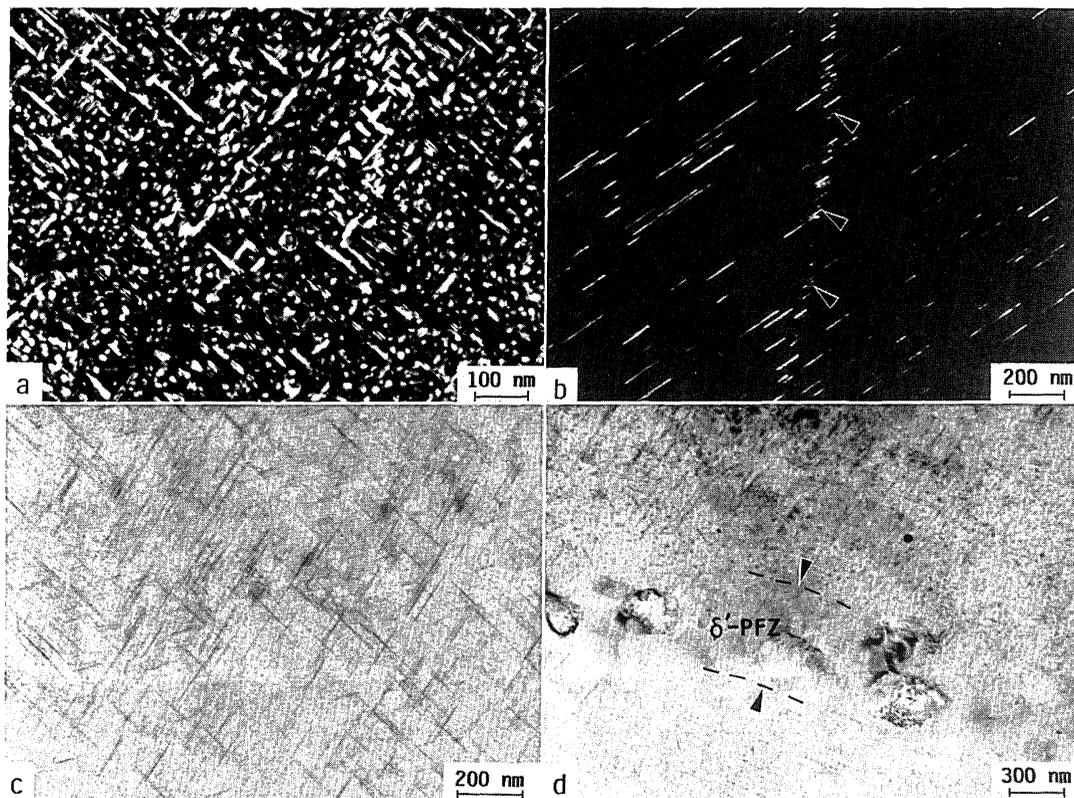
The resulting grain structures for most ingot metallurgy (IM)-based, high strength aluminium alloys

following such thermomechanical processing are disc shaped and elongated in the rolling direction; Al-Li alloy sheet and plate, however, show a greater degree of anisotropy because of the small Zr additions which can effectively suppress recrystallisation and retard grain growth. As shown in Fig. 1a, grain sizes in IM alloys are fairly coarse, typically 500 μm wide, 50 μm thick, and several mm in length; the exception is the 8091 alloy where grain sizes are finer.⁵⁹ Microstructures in powder metallurgy (PM) processed Al-Li alloys are more refined; grain sizes are typically between 0.5 and 5 μm and the structures are less anisotropic (Fig. 1b,c). Strong deformation textures are common in IM alloys owing to prior thermomechanical treatments. These textures are found to be predominantly of the 'brass' type ($\{110\}\langle 112 \rangle$), with evidence of weaker 'S' ($\{123\}\langle 634 \rangle$) and 'copper' ($\{112\}\langle 111 \rangle$) types.⁶⁰⁻⁶⁷ However, the various components of texture vary strongly across the thickness of the plate with the intensity of each component showing a maximum at the centre;⁶⁰ moreover, $\{001\}\langle 110 \rangle$ type recrystallisation textures are observed in the near surface regions.⁶³ Chemical compositions and thermomechanical treatments for the principal commercial Al-Li alloys are listed, respectively, in Tables 2 and 3.

In underaged tempers, commercial Al-Li alloys are strengthened by metastable δ' spheres, β' (Al_3Zr , also referred to as α') dispersoids, and composite precipitates of δ' surrounding β' (Fig. 2a). Al_3Zr particles are spherical, coherent, and ordered with an $L1_2$ superlattice structure. In addition, aging to peak strength (T8 condition) in Al-Li-Cu-Zr (2090) systems causes matrix precipitation of T_1 and θ' plates, coarsening of matrix δ' , the formation of relatively narrow (~ 100 nm wide) δ' -PFZs at high angle grain boundaries, and fine precipitation of T_1 plates along subgrain boundaries (Fig. 2a,b). In Mg-containing alloys, T_1 and θ' plates are replaced by S (Al_2CuMg) laths in the matrix (Fig. 2c); precipitation of T_1 occurs in competition with S for available nucleation sites and copper atoms. For example, precipitation of T_1 plates has been reported for 8090 type alloys but not for 8091 compositions.⁴⁷ Despite the pre-aging deformation, grain boundary effects are prevalent, even in peak aged tempers. In 8090-T8, small amounts of grain boundary precipitation result in ~ 500 nm wide δ' -PFZs (Fig. 2d); in 8091-T8, heterogeneous precipitation is more extensive and PFZs are correspondingly wider (~ 1 μm).

With longer aging times (overaged tempers), coarse equilibrium Cu- and/or Mg-rich phases and T_1 plates or T_2 precipitates decorate grain and subgrain boundaries, respectively, concurrently leading to the formation of solute denuded PFZs. In addition, undissolved impurities during processing are also present as coarse Fe- and Cu-rich intermetallic phases along the elongated high angle grain boundaries. Microstructural details and room temperature mechanical properties of the principal IM and PM Al-Li alloys are summarised in Tables 4 and 5, respectively.

Specific Al-Li alloys can be $\sim 30\%$ higher in strength with similar elongation and toughness properties when compared with traditional aluminium alloys (Table 5). However, these properties are typical



a δ' (Al_3Li) spheres and θ' -like (Al_2Cu) plates; b T_1 (Al_2CuLi) plates seen in Al-Li-Cu-Zr systems (2090-T81); c S (Al_2CuMg) laths in Mg-bearing alloys (peak aged 8090 and 8091); d grain boundary precipitation with associated δ' -PFZs (8090-T8)

2 TEM micrographs showing predominant microstructural features and strengthening precipitates in commercial Al-Li alloys. Spherical β' (Al_3Zr) dispersoids surrounded by δ' , and δ' wetting the θ' plates may also be noted in a; imaging in a, b was done under dark field conditions using δ' and T_1 superlattice reflections, respectively (Ref. 41)

of thermomechanically processed wrought product forms (namely prestretched and artificially aged sheet and plate) in the longitudinal orientation; generally short-transverse properties are considerably less attractive. The unrecrystallised elongated grain structures and deformation-texture variations ensure that

properties are strongly dependent on orientation, both in the rolling plane and through the plate thickness.^{41,65-67,70} Moreover, the excellent strength and toughness properties, characteristic of sheet and plate products, are often compromised for forgings, extrusions, and thick section plates, where uniform

Table 4 Summary of grain size measurements and strengthening precipitates in various advanced and conventional aerospace aluminium alloys (Refs. 16-20, 58, 59, 68)

Alloy	Typical grain size*			Principal hardening precipitates†
	L, mm	T, μm	S, μm	
2090-T81	2-3	500	50	δ' (Al_3Li), T_1 (Al_2CuLi), θ' (Al_2Cu) β' (Al_3Zr)
2091-T351	1-2	600	50	δ' , β'
2091-T8	1-2	350	40	δ' , β'
8090-T351	0.3	65	25	δ' , β'
8090-T8	δ' , T_1 , S (Al_2CuMg), β'
8091-T351	δ' , β'
8091-T8	δ' , S, β'
W049-T3	G-P zones, δ'
W049-T8	δ' , T_1 , θ'
2020-T651	0.8	75	35	δ' , T_1 , θ' , T_B
IN-905XL	0.0015	0.4	0.3	Al_2O_3 , Al_4C_3
644-B	0.05-0.1	1-2	1-2	δ' , β' , composite δ' - β'
2124-T351	0.7	350	50	G-P zones
7150-T651	2-3	750	30	η' (MgZn_2 - $\text{Mg}(\text{CuAl})_2$)
7075-T651	0.125	16	10	G-P zones, η' , $\text{Al}_2\text{Mg}_2\text{Cr}$
7091-T7E69	2-5	1-2	1-2	G-P zones, η' , Co_2Al_9 , Al_2CuMg , oxides

*L, T, and S refer to measurements in the longitudinal, long-transverse, and short-transverse directions, respectively.
†Refer to text for details on morphology and structure of the different phases.

quenching rates and pre-aging deformation are not feasible.^{65–67,70–75}

Cyclic stress–strain and fatigue crack initiation behaviour

The application of cyclic loads to metallic materials at stress levels much below the yield or tensile strengths can significantly alter their constitutive behaviour. Depending on the initial wrought product form, heat treatment, and applied plastic-strain amplitude ($\Delta\epsilon_p/2$), the material may undergo cyclic softening, cyclic hardening, show mixed (hardening and softening) response, or have no effect at all.⁷⁶ Following transient hardening or softening, materials generally exhibit saturation or stability at some equilibrium stress or strain value, before eventually softening due to incipient damage or crack nucleation. High strength aluminium alloys are good examples of cyclic hardening materials, whereas softening is observed in quenched and tempered steels; low carbon steels, high strength low alloy (HSLA) steels, and titanium alloys are examples of materials which initially soften before hardening.⁷⁶ The cyclic stress–strain properties are traditionally evaluated either in low cycle fatigue (LCF), where the strains are predominantly plastic, or in high cycle fatigue (HCF), where the macroscopic strains are predominantly elastic. Typically, lifetimes under LCF conditions are below $\sim 10^3$ cycles, whereas HCF pertains to lifetimes exceeding $\sim 10^3$ cycles.

Low cycle fatigue behaviour

As with most high strength Al alloys, Al–Li alloys in general initially harden under cyclic loading, particularly at low plastic-strain amplitudes (typically $\Delta\epsilon_p/2$ below $\sim 10^{-3}$), followed by saturation before softening to final fracture.^{57,77–79} At higher plastic strains ($\Delta\epsilon_p/2$ above $\sim 10^{-3}$), however, the alloys continuously soften to failure with little cyclic stability.^{57,79–81} Similar trends are apparent for most experimental and commercial Al–Li alloys, though

the extent of hardening, saturation, or softening and resultant fatigue lifetimes are a strong function of alloy composition, microstructure, temperature, and environment.^{34,57,77–94} Typical variations in the cyclic stress amplitude with applied strain, under fully reversed, strain controlled loading for binary Al–Li alloys and commercial 8090–T6 alloy, are illustrated in Fig. 3; corresponding monotonic and stabilised cyclic stress–strain curves are compared in Fig. 4.

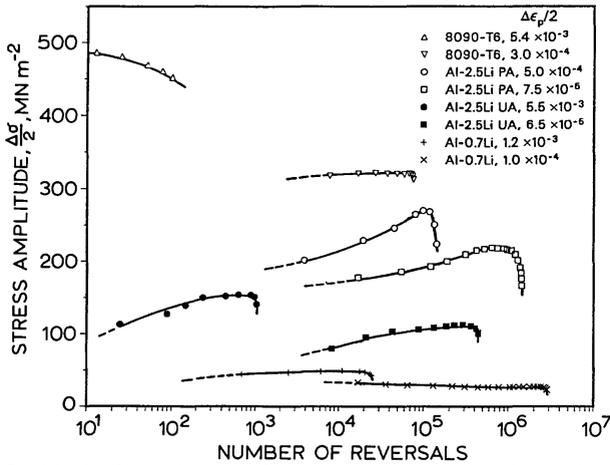
Binary Al–Li alloys

Strengthening by lithium in solid solution ($\sim 1.7\%$ Li) yields microstructures which initially cyclically harden at plastic strain amplitudes less than $\sim 10^{-3}$, before softening and eventually attaining cyclic stability until fracture.^{57,78,84,85} Hardening can be permanent for alternating strains greater than $\sim 10^{-3}$ with no saturation before softening⁵⁷ (Fig. 3); corresponding cyclic stress–strain curves show continuous hardening to failure without any plateau region (Fig. 4a). The increased flow stress during fatigue can be attributed to an increase in the dislocation density, dislocation–dislocation interactions, and dislocations interacting with solute-atom strain fields.⁷⁷ Transmission electron microscopy (TEM) observations reveal the deformation to be largely homogeneous.^{34,57} Cellular dislocation structures, similar to those observed in single phase, high stacking-fault energy fcc metals, are common particularly at high plastic-strain amplitudes (Fig. 5a), though they are slightly more elongated and less polygonised compared with pure aluminium (wavy slip); no veins or ladder-like persistent slip bands (PSBs) are seen. At low plastic-strain amplitudes, slip is planar.³⁴ With increased deformation, softening is associated with the nucleation of microcracks along slip band intrusions emerging on the surface for $\Delta\epsilon_p/2 < 10^{-3}$, and along grain boundaries or where slip bands impinge at grain boundaries for $\Delta\epsilon_p/2 > 10^{-3}$. Propagation of the fatal crack occurs by non-crystallographic, ductile striation type, crack growth mechanisms (Fig. 5b) that involve alternating shear or successive blunting and resharping of the crack tip.^{34,57,78}

Table 5 Room temperature mechanical properties of selected experimental and commercial aluminium–lithium and aluminium alloys*

Alloy	Young's modulus E , GN m ⁻²	Yield strength σ_y , MN m ⁻²	UTS σ_u , MN m ⁻²	Elongation (on 25 mm gauge), %	Strain hardening coefficient n	Fracture toughness K_{Ic} , MN m ^{-3/2}
2090–T8	76	552	589	11	0.06	36
2091–T351	75	369	451	10	0.12	33
2091–T8	75	425	481	8	0.10	46
8090–T351	77	226	352	17	0.19	27
8090–T8	77	482	534	6	0.08	36
8091–T351	...	309	417	11	0.16	38
8091–T8	...	537	581	6	0.07	20
W049–T3	79	379	496	15	...	37
W049–T8	79	683	703	5
2020–T651	77	534	567	5	...	21
IN–905XL	...	559	596	2.3	...	13
644–B	...	422	539	8	0.19	24
2124–T351	73	360	488	18	...	24
7150–T651	73	404	480	6	...	21
7075–T651	73	503	572	11	...	28
7091–T7E69	...	545	593	11	...	46

*Uniaxial tensile properties are obtained in the longitudinal (L) direction and fracture toughness in the long-transverse (L-T) orientation. Values taken from Refs. 16–20, 58, 59, 68, 69.

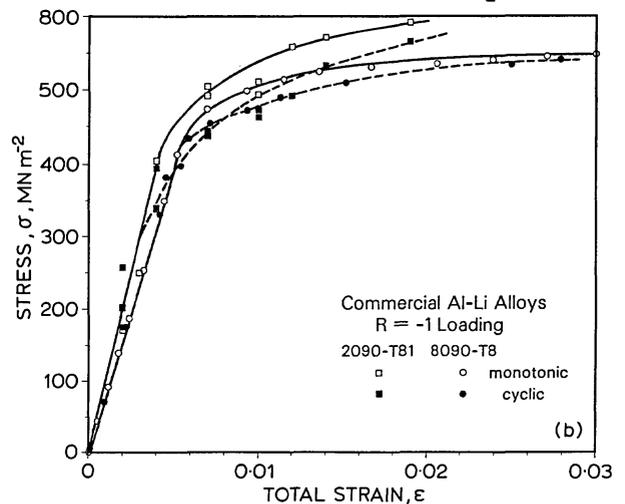
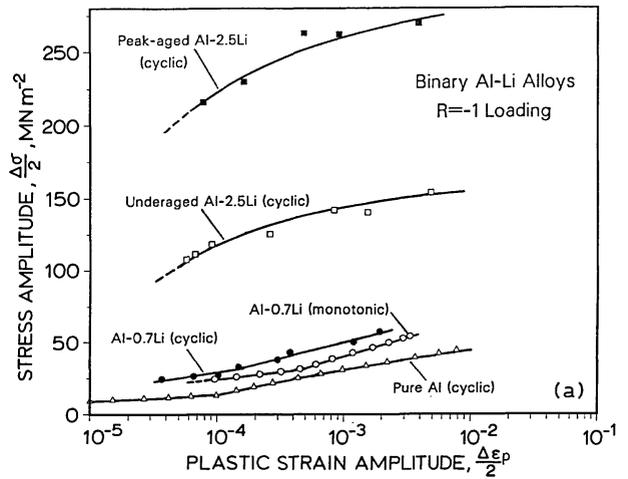


3 Variation in cyclic stress amplitude with number of reversals at various plastic-strain amplitudes in a Al–0.7Li solid solution, underaged (UA) and peak aged (PA) Al–2.5Li alloy (precipitation hardened), and commercial 8090–T6 alloy (Refs. 57, 79)

In contrast, aged microstructures with lithium contents exceeding $\sim 1.7\%$ that are precipitation hardened by δ' particles (e.g. Al–2.5Li alloy), show significant cyclic hardening (Figs. 3, 4a). Cyclic stress–strain curves show an initial increase in stress amplitude due to dislocation interactions, with small saturation periods (which diminish with artificial aging), followed by softening to failure from δ' shearing and strain localisation.³⁴ Unlike dislocation cell structures in solid solution hardened microstructures, fatigue deformation is inhomogeneous and slip is essentially planar for all plastic strain amplitudes;^{34,57} intense slip bands are developed parallel to $\{111\}$ planes (Fig. 5c).

However, the strong cyclic hardening behaviour of Al–Li alloys with higher Li content and higher δ' volume fractions is offset by marked reductions in LCF resistance, as shown by the Coffin–Manson curves in Fig. 6a. Although solid solution hardened and underaged microstructures experience only limited losses in fatigue strength compared with pure aluminium (an exception is underaged alloys at low plastic strains where slip is planar), fatigue lives are reduced by as much as four orders of magnitude in artificially peak aged and overaged binary Al–Li alloys.^{34,57,78,80–86}

The detrimental effects of Li on the LCF resistance of Al–Li alloys is principally due to the presence of large volume fractions of shearable, ordered δ' precipitates; small Li additions which go into solid solution have almost no effect.^{34,57} During deformation, initial dislocation motion shears the δ' precipitates and reduces order by creating antiphase domain boundaries within the particles; following dislocations preferentially move on the same glide plane at a lower resolved stress to restore order, thus resulting in pair-dislocation motion and a tendency toward planar slip. Additionally, planar slip is promoted by the reduced effective diameter of sheared δ' particles and reductions in the stacking fault energy of aluminium due to Li additions.³⁴ So deformation is localised along narrow PSBs parallel to the $\{111\}$ glide



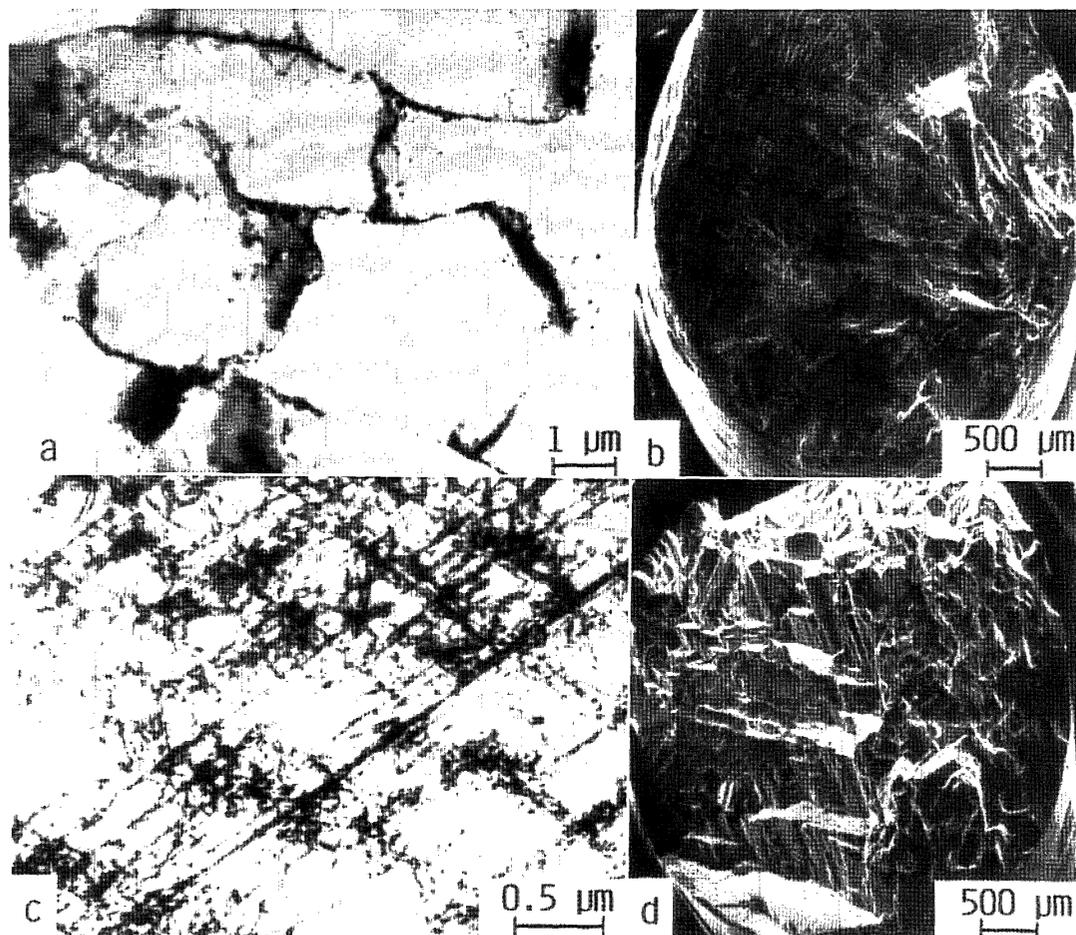
a binary microstructures; b commercial 2090–T81 and 8090–T8 alloys

4 Monotonic and stabilised cyclic stress–strain curves for various Al–Li alloys (longitudinal direction, $R = -1$) (Refs. 57, 70, 79)

planes causing stress concentrations at their intersection with grain boundaries. These effects become more pronounced in the peak and overaged tempers because of an increase in size and amount of δ' particles;^{34,57} equilibrium δ precipitates further accentuate this effect by aggravating the degree of strain localisation in the soft δ' -PFZs at grain boundaries. Microcrack nucleation at dislocation pile-ups is common for all plastic-strain amplitudes; coalescence of these microcracks to form a fatal flaw may occur crystallographically along slip bands or at weakened grain boundaries, depending on the alloy temper. As shown in Fig. 5d, fatigue fracture in underaged Al–2.5Li is transgranular as microcracks link up by slip band cracking.^{34,57} Conversely, intergranular cracking is prevalent in the peak and overaged microstructures; few secondary cracks initiate, and the first intergranular crack to nucleate propagates catastrophically to failure.

Commercial Al–Li alloys

In commercial Al–Li alloys, cyclic and monotonic flow stresses are substantially higher, and fatigue lives correspondingly lower, compared with binary Al–Li alloys because of additional strengthening from T_1



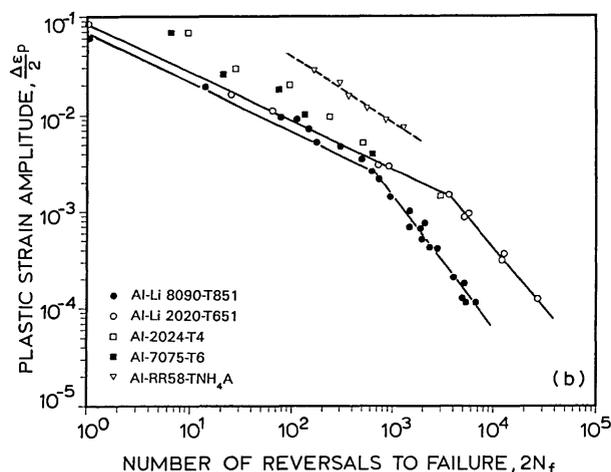
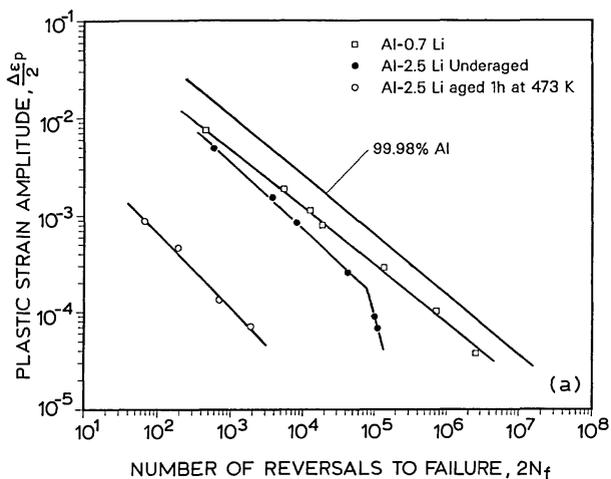
5 TEM micrographs of dislocation substructures generated during cyclic deformation *a,c*, and corresponding SEM images of fatigue fracture surfaces *b,d*, in binary Al-Li alloys hardened by: *a,b* Li in solid solution (underaged Al-0.7Li alloy), and *c,d* ordered δ' (Al_3Li) precipitates (peak aged Al-2.5Li). Failures in solid solution occur by ductile striated type mechanism corresponding to homogeneous, cellular deformation structures, whereas δ' -hardened structures fracture by slip band cracking due to localised deformation along close packed $\{111\}$ planes (Refs. 34, 57)

and S precipitates (Figs. 4*b*, 6*b*). T_1 and S precipitates are thought to favour a dispersion of slip and promote plastic strain homogenisation at the subgrain level through the activation of multiple slip systems within the grain. However, inhomogeneous fatigue deformation still occurs along PSBs or embrittled grain boundaries. In fact, the partially coherent S precipitates have been observed to shear during fatigue at high plastic strains, presumably due to stress concentrations at the matrix/ S interfaces that favour dislocation nucleation and precipitate cleavage.⁸⁸ Moreover, the highly textured grain structures in commercial alloys limit the number of slip systems by reducing the effectiveness of grain boundaries as barriers to dislocation motion due to the low degree of misorientation.^{78,79,81,86-91}

Cyclic stress-strain behaviour in commercial alloys is similar to that in binary alloys strengthened by δ' particles (Figs. 3, 4*b*). Limited hardening and stability, followed by softening, is seen at low strain amplitudes due to crack nucleation at slip bands emanating from large ($\sim 1-10 \mu\text{m}$) cracked or uncracked constituent particles; at high strains, the alloys soften continuously to failure. Crack nucleation occurs in coarse slip bands, possibly at intrusions or extrusions on the

specimen surface; hardening and softening are generally associated with dislocation-particle interactions and δ' shear, respectively.^{89,90} However, Coffin-Manson curves for commercial alloys show a transition at low strain amplitudes below $\sim 10^{-3}$, which approximately corresponds to strains below which the cyclic stress response is stable (harden) and above which the alloys soften (Fig. 6*b*). Most studies associate this transition with changes in deformation mechanism, or the degree of slip homogeneity.^{79,89,90} Such changes have also been noted in 8090-T8 and 2020-T651 Al-Li alloys; ductile intergranular fractures, characteristic of monotonic failure, are prevalent at high strain amplitudes above the transition strain, whereas transgranular shear (slip band cracking) failures are observed below the transition strain, suggesting marked effects of slip planarity at low strain ranges.^{78,79,81,86-88}

Microstructural factors that enhance slip homogeneity in Al-Li alloys are generally found to improve LCF resistance. For example, stretching (before aging) to increase the number density of the partially coherent T_1 and S phases, the addition of dispersoid forming elements, and fine recrystallised grain structures, are all considered beneficial to LCF.^{78,80,89-91} In con-



a binary alloys; b commercial 8090 and 2020 alloys

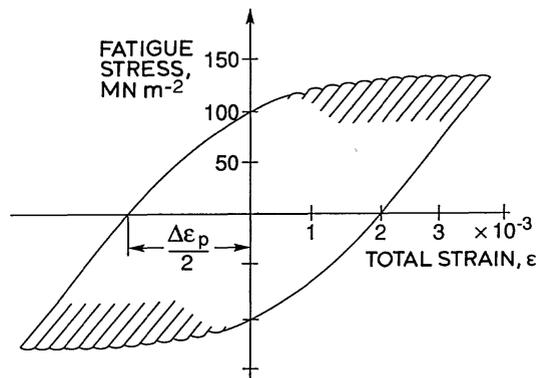
6 Comparison of Coffin–Manson relationships for different binary and commercial Al–Li alloys, compared with pure aluminium and traditional 2000 and 7000 series Al alloys in the longitudinal orientation. Note that Li present as δ' particles, is detrimental to LCF resistance. Data taken from Refs. 34, 57, 79

trast, results in 2020 suggest that unrecrystallised microstructures may exhibit better fatigue resistance than recrystallised structures;⁸⁰ this effect, however, seems to be associated with the suppression of grain boundary precipitation by the pre-aging deformation.

Finally, compared with traditional high strength alloys, commercial Al–Li alloys possess comparable fatigue crack initiation properties (Fig 6b). In the high strain region of the Coffin–Manson curve, however, crack initiation resistance may be somewhat lower in Al–Li alloys because of higher elastic cyclic stresses for a given cyclic strain, resulting from their increased Young's modulus.⁷⁰

Stress instabilities during cyclic hardening

Recent studies^{92–94} have shown that following cyclic hardening, stabilised cyclic stress–strain curves (or hysteresis loops) for binary Al–Li alloys often exhibit stress instabilities, characterised by successive drops in the fatigue stress (Fig. 7), depending on the plastic-strain amplitude, alloy temper, test frequency, and temperature. For example, the instabilities are only

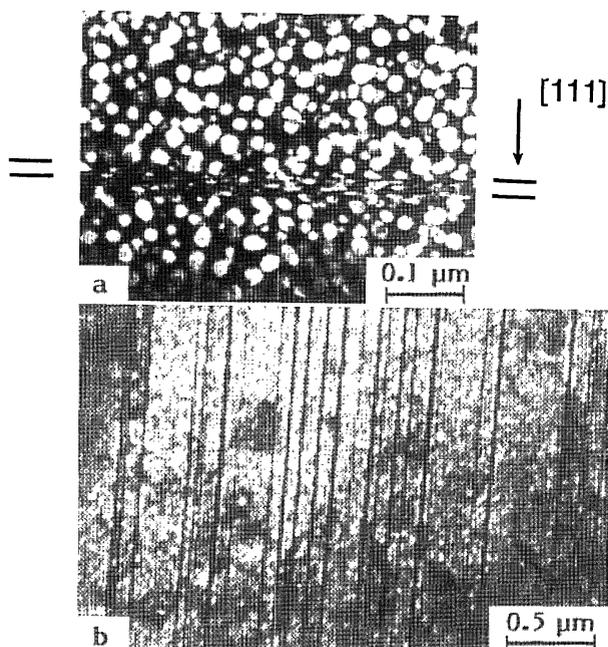


7 Illustration of fatigue stress instabilities observed on stabilised cyclic stress–strain hysteresis loop of Al–2.5Li alloy, aged for 5 h at 423 K. Results are for plastic strain amplitude $\Delta\epsilon_p/2 = 2 \times 10^{-3}$ (0.01 Hz) at room temperature (Ref. 92)

seen at specific temperatures above 233 K for Al–Li microstructures hardened by δ' (as opposed to solid solution strengthening), when accumulated plastic strains ($2N \times \Delta\epsilon_p/2$, where N is the number of cycles) exceed ~ 0.1 (independent of alloy composition and aging temper), and at frequencies below 0.2 Hz. Moreover, increased cyclic flow strengths have been measured by interrupting the test and room temperature aging the material.

Gentzbitel and Fougères⁹² suggest that this behaviour is influenced by dynamic strain aging or the Portevin–Le Chatelier phenomenon; however, as the stress instabilities are not seen in Al–Li solid solutions (i.e. in Al–0.7Li alloy), this implies that the effect is associated with δ' precipitates rather than dislocation–lithium (solute) atom interactions. It is more probable that the effects originate from the instability of small δ' particles ($\gtrsim 1.5$ nm) to localised shearing by moving dislocations; mechanical instabilities in the fatigue stress are then a result of successive reversion and reprecipitation of δ' in the matrix. In fact, TEM studies by Brechet *et al.*⁹³ have confirmed this observation. In the early stages of deformation, spherical δ' particles are sheared by dislocations leading to the formation of antiphase domain boundaries (Fig. 8a). On repeated shearing, the δ' precipitates become unstable from the antiphase boundary energy and dissolve in the matrix. Since deformation is concentrated along narrow bands parallel to the {111} slip planes, the resulting microstructure (Fig. 8b) consists of precipitate free bands enriched in lithium, where dislocation motion is relatively unimpeded compared with microstructure outside the bands. Finally, assisted by fatigue induced vacancies, lithium diffuses to the side of the band and reprecipitates as δ', thus enhancing the local coarsening kinetics and volume fraction of δ' particles (snow plough effect). Simple models proposed to explain the redissolution of δ' within the bands, band width, and band separation, based on an increase in the free energy from shear induced, antiphase boundaries, are in reasonable agreement with experimental results.^{93,94}

Although the precise relevance of such microscopic fatigue damage processes to macroscopic crack



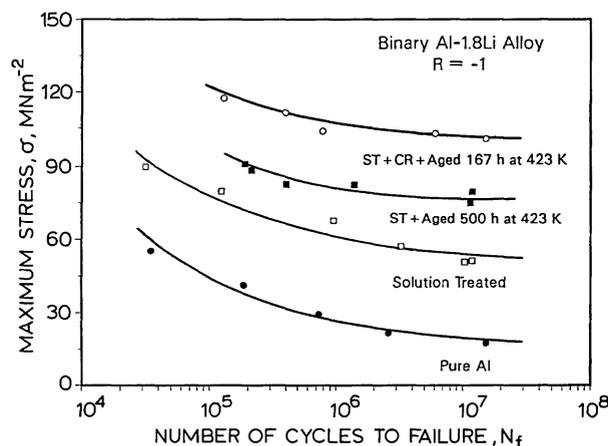
8 Dark field TEM micrographs showing a shearing of ordered δ' particles during early stages of cyclic deformation (as seen in peak aged 8090 Al-Li alloy, $\Delta\epsilon_p/2=0.5\%$, from Ref. 88) and b fatigue induced, δ' precipitate free bands parallel to $\{111\}$ slip bands in binary Al-Li alloys (Al-2.5Li, aged for 8 h at 373 K; mean diameter of δ' particles ~ 2 nm, from Ref. 93). Imaging performed using δ' -superlattice reflections

advance mechanisms is as yet unclear, these observations do confirm that cyclic deformation in Al-Li alloys proceeds by dislocation shearing of the ordered δ' precipitates. Similar studies on the LCF behaviour of 8090 and 2020 Al-Li alloys have also shown that, at high plastic-strain levels, spherical δ' , lath-like S precipitates and even Al_3Zr dispersoids are shearable and may contribute to low fatigue lives.⁸⁷⁻⁸⁹

High cycle or $S-N$ fatigue

In contrast to LCF behaviour where Li in the form of δ' precipitates has an adverse effect on fatigue life, the high cycle fatigue properties of Al-Li alloys are excellent.⁹⁵⁻⁹⁸ Compared with pure aluminium, the fatigue resistance of binary Al-Li alloys increases with Li content and aging time (Fig. 9), primarily due to strengthening by Li in solid solution, and the growth of ordered δ' precipitates; cold work before artificial aging further enhances this effect. Similar results have been reported for binary Mg-Li alloys,⁹⁶ where increases in fatigue strength and tensile strength are again related.

Although the global strains that cause HCF failures are nominally elastic, plastic strains are experienced locally at microstructural heterogeneities. Metallographic observations on as quenched Al-1.8Li alloy show very fine surface slip bands distributed homogeneously within the grains during the early stages of cycling at high stress levels; these coarsen and assist microcrack nucleation at sharp slip bands or grain boundaries before finally propagating to failure.⁹⁵



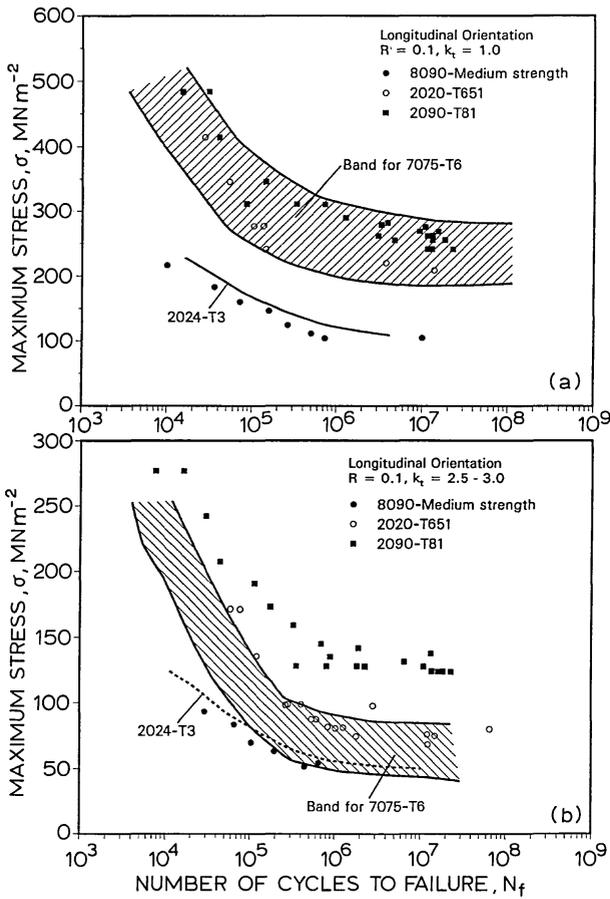
9 High cycle fatigue curves for a binary Al-1.8Li alloy, following various thermomechanical treatments compared with pure aluminium. Results shown are for smooth, 2.3 mm thin-sheet specimens, tested under fully reversed ($R=-1$) bending fatigue parallel to rolling direction, taken from Ref. 95. ST and CR refer to solution treatment and 67% cold rolling, respectively

Artificial aging further increases strength and induces greater inhomogeneity in fatigue deformation with increasing δ' -particle size and distribution. Planar dislocation arrays are seen in TEM studies along $\langle 110 \rangle$ in the solution treated condition; similar features are evident in artificially aged tempers at low cyclic stresses, and dislocation cell structures are evident at high stress levels. After thermomechanical treatments, Al-Li alloys show greater improvements in fatigue strength because of higher monotonic strength, coupled with a cell structure that alleviates slip localisation and retards crack initiation.⁹⁵

Smooth- and notched-specimen stress-lifetime ($S-N$) fatigue data (at $R=0.1$) for commercial Al-Li alloys 2020-T651, 2090-T81, 2091-T8, and 8090 are comparable with results for 7075-T6 and 2124-T351⁹⁸⁻¹⁰¹ (Fig. 10). Typical endurance strengths of the commercial alloys are roughly 40-50% of the tensile strength; stretching before aging can lead to further improvements.^{95,98} However, results on 2091-T8 alloy⁹⁹ suggest that undissolved intermetallic constituent particles (rich in Cu, Fe, and Mg) can also influence HCF as they are preferential nucleation sites for microcracks. Moreover, the HCF properties of commercial Al-Li alloys are highly anisotropic because of marked deformation textures and disc shaped grain structure that result in strength variations with specimen orientation; in 2090-T83 Al-Li sheet, for example, $S-N$ fatigue lives for the L+45° orientation are significantly lower than for the longitudinal (L) or long-transverse (L-T) orientations, particularly in smooth as opposed to notched samples.⁷⁰

Effect of surface treatment

Surface treatments using shot peening, ion implantation, nitriding, and laser glazing techniques are commonly employed to improve fatigue crack initiation resistance through the introduction of compressive residual stresses at the surface; work



10 Comparison of a smooth unnotched (stress concentration factor, $k_t=1$), and b notched ($k_t=3$) fatigue crack initiation resistance under axial loading conditions ($R=0.1$) for various commercial Al-Li and conventional Al alloys. HCF resistance of Al-Li alloys is comparable with 2000 and 7000 series Al alloys. Results compiled from Refs. 100, 101

oxide films is presumed to be the principal cause for this behaviour.¹⁰²

Fatigue crack propagation

In recent years, damage tolerant design procedures have become increasingly important in predicting life and ensuring the durability of many safety critical structures; the concept presumes that defect populations pre-exist in components and that structural integrity is a function of the number of cycles to propagate the largest undetected flaw to failure.¹⁰⁵⁻¹⁰⁷ Lifetimes are conservatively predicted by integration of a relationship describing the crack growth rate (da/dN) as a function of the mechanical crack size. For nominally linear elastic conditions, where the driving force can be defined in terms of the applied stress intensity range, $\Delta K = K_{max} - K_{min}$ (K_{max} and K_{min} are the maximum and minimum stress intensities), these relationships are generally based on the Paris power-law relation, namely

$$da/dN = C \Delta K^m \dots \dots \dots (1)$$

where C and m are experimentally determined scaling constants. Although equation (1) provides a reasonable description for crack growth rates in a range typically between $\sim 10^{-9}$ and 10^{-6} m/cycle, it does not adequately characterise behaviour at ΔK levels approaching the fatigue threshold, ΔK_{TH} , below which long cracks are presumed dormant, or close to instability, where limit load or catastrophic failure (e.g. at $K_{max} = K_{Ic}$, the fracture toughness) occurs; in these instances, more complex equations are curve fitted to the crack growth data. Such crack propagation 'laws' are generally determined under constant amplitude loading conditions, using fracture mechanics type specimen geometries containing pre-existing, long (>5 mm) through thickness cracks, via manual or computer-controlled load shedding schemes which reduce ΔK levels at a fixed load ratio ($R = K_{min}/K_{max}$).

In many materials crack propagation behaviour is strongly affected by crack closure, particularly at near threshold levels;^{45,108-114} in fact, significant effects of microstructure, load ratio, and even environment have been attributed to variations in crack closure levels. The closure phenomenon is a consequence of the crack not being fully open during the entire loading cycle, even under tension-tension loading conditions, such that premature contact of the crack surfaces occurs at the closure stress intensity, K_{cl} , before minimum load is reached. Prominent mechanisms include closure caused by residual plastic deformation - plasticity induced crack closure, interference of fracture surface asperities between mating fracture surfaces in the crack wake - roughness induced crack closure, and wedging of the crack from corrosion or oxide deposits on the fracture surface - oxide induced crack closure.¹⁰⁸⁻¹¹² The salient mechanisms pertinent to high strength aluminium alloys and their influence on crack growth rates are illustrated in Fig. 11; measurement techniques are discussed in Refs. 113, 114.

Similar to other crack tip shielding mechanisms,^{45,46} the role of closure is to perturb the near-

hardening effects also play a minor role. Studies on peak aged 8090¹⁰²⁻¹⁰⁴ have confirmed that the HCF resistance of Al-Li alloys can be improved using such techniques. However, crack initiation lifetimes (defined as the number of cycles required to grow a 100 μ m crack) can be shorter in shot peened compared with untreated specimens, because of early crack nucleation at stress concentrations or fold like defects generated by the peening; subsequent crack propagation is retarded such that total fatigue lifetimes are improved.

Other treatments, such as acid pickling and anodising (used as precursors to adhesive bonding and for corrosion protection) can lead to reduced fatigue life due to premature crack nucleation from preferential grain boundary attack and the rupture of surface oxide films. Recent results^{102,104} indicate, however, that 8090-T8 is less susceptible to problems from chromic-sulphuric acid pickling than Al-Zn-Mg-Cu alloys, because the reduced grain boundary precipitation decreases the electrochemical driving force for dissolution and crack formation. Moreover, unlike other aluminium alloys, the fatigue strength can be restored by reanodising the pickled surfaces; an increased fracture strain of the lithium bearing surface

1. CRACK DEFLECTION AND MEANDERING

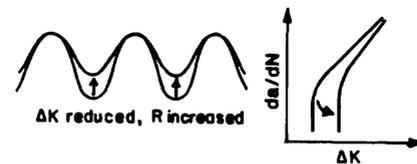
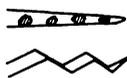


2. CONTACT SHIELDING

— wedging:

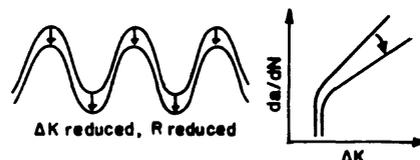
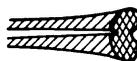
corrosion debris-induced crack closure

crack surface roughness-induced closure



3. COMBINED ZONE AND CONTACT SHIELDING

— plasticity-induced crack closure



11 Schematic illustration of crack tip shielding mechanisms influencing fatigue crack propagation in high strength Al alloys, showing primary mechanisms and their consequences on crack growth rate behaviour (Ref. 115)

tip stress or strain field by lowering the local driving force actually experienced at the crack tip. This driving force can be defined as⁴⁵

$$\Delta K_{\text{eff}} = K_{\text{max}} - K_{\text{cl}} \quad (K_{\text{cl}} > K_{\text{min}})$$

$$\Delta K_{\text{eff}} = \Delta K \quad (K_{\text{cl}} < K_{\text{min}}) \quad \dots \dots \dots (2)$$

Where growth rates are correlated in terms of ΔK_{eff} , the crack growth relationships have been termed intrinsic, i.e. devoid of mechanical crack-closure effects. Alternatively, the intrinsic fatigue crack propagation resistance can be directly assessed using constant K_{max} /increasing K_{min} (variable R) load shedding schemes which restrict closure effects.¹¹⁶

As noted previously, localised planar slip deformation in δ' -hardened Al–Li alloys has adverse effects on (LCF) crack initiation resistance; however, paradoxically, it can be beneficial to crack propagation resistance by inducing crystallographically tortuous crack paths. Such geometrical changes in crack path morphology promote crack tip shielding, in particular from the wedging of fracture surface asperities, and retard crack advance.

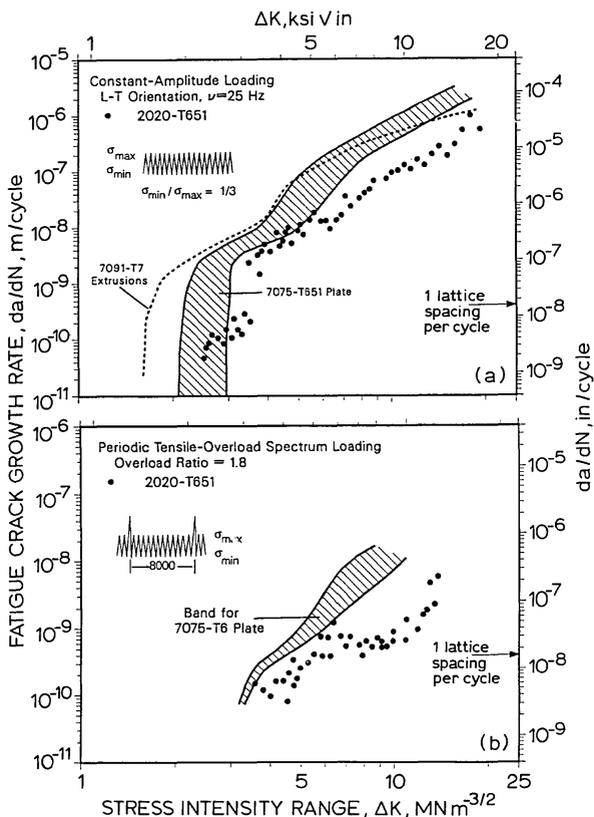
Behaviour in experimental Al–Li alloys and 2020

Early studies of fatigue–crack propagation in Al–Li alloys were performed on Al–3Li–1.25Mn systems⁷⁷ and peak aged 2020–T651¹⁰¹ in moist air, where it was found that crack growth rates, between 10^{-9} and 10^{-5} m/cycle, were significantly slower than in 7075–T6, particularly at low stress intensity levels. Such observations were attributed to the higher elastic modulus of Al–Li alloys, which results in lower crack tip opening displacement (CTOD) at fixed ΔK and consequently lower crack growth rates per cycle, though at higher ΔK levels approaching instability, growth rates were faster in the Al–Li alloys. Similar behaviour was noted for Al–Li–Mg alloys¹¹⁷ and

rapid solidification processed (RSP) Al–Li–Cu–Mn alloys¹¹⁸ and the improved fatigue properties were associated with slip band cracking; enhanced slip reversibility because of the planarity of slip from δ' -particle hardening was reasoned to diminish the crack tip damage (irreversible cyclic plastic strain) accumulated each cycle.¹¹⁸ Residual stresses developed during quenching after solution treatment were also found to have a significant influence on crack growth in RSP materials.

More recently, with the renewed interest in Al–Li systems, the fatigue behaviour of peak aged 2020 alloy has been re-examined over a wider range of growth rates between 10^{-11} and 10^{-6} m/cycle.¹¹⁹ Crack propagation data under constant amplitude (variable ΔK /constant $R=0.33$) and spectrum loading (single tensile overload every 8000 cycles with $K_{\text{ol}}/K_{\text{max}}=1.8$) are compared with corresponding results for 7075–T651 and 7091–T7E69 in Fig. 12 (for an ambient moist air environment). As reported in earlier studies,¹⁰¹ 2020 exhibits significantly slower crack growth rates, by more than two orders of magnitude at near-threshold levels; however, this cannot be explained purely on modulus differences since the improved crack growth properties of 2020 are apparent even after normalising growth rate data with Young's modulus.¹¹⁹ Of more importance here is the fact that crack advance in 2020 is highly crystallographic at all ΔK levels with substantial crack deflection and branching; consequently, fracture surfaces are extremely rough and faceted, with little evidence of ductile striations. In addition, oxide deposits have been detected on these surfaces, with excess oxide thicknesses estimated from secondary ion mass spectroscopy to be between ~ 5 and 10 nm.

Based on these observations, the superior fatigue crack propagation kinetics in Al–Li alloys, specifically 2020, are attributed to the highly non-linear and tortuous crack path morphologies. This follows from

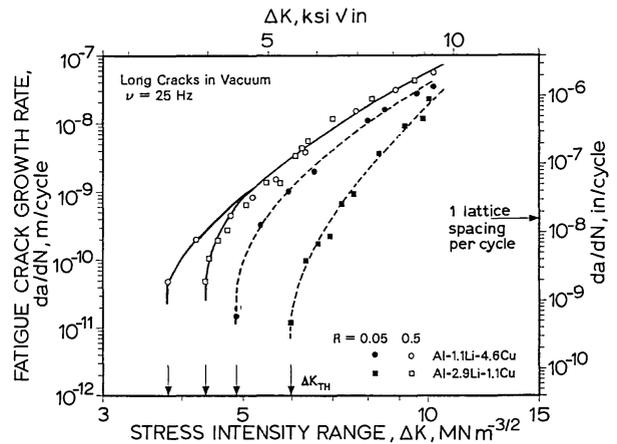


a constant amplitude; b periodic single-spike overload spectrum loading

- 12 Fatigue crack propagation behaviour in peak aged 2020-T651 alloy compared with IM 7075-T651 and PM 7091-T7E69 alloys. Note that growth rates in 2020 are significantly slower, particularly at near-threshold stress intensity levels and under tension dominated spectrum loading (Ref. 101)**

coherent particle (δ') hardening in the coarse grained and textured microstructures which results in marked planar slip deformation along close packed $\{111\}$ slip planes; this in turn promotes cracking along slip bands leading to a crystallographically faceted mode of crack extension.^{115,119-123} The resulting crack deflection and branching locally reduce the mode I stress intensity at the crack tip and further promote roughness induced crack closure via wedging of the fracture surface asperities (facets) in the crack wake. Despite the more heavily oxidised fracture surfaces in 2020 (compared with traditional 2000 series alloys) caused by the presence of lithium, additional contributions to crack closure from the wedging of oxide debris are relatively insignificant because the excess oxide thicknesses remain small compared with the CTOD.

Subsequent results on high purity Al-Li-Cu-Zr alloys, containing controlled variations of Cu and Li,^{35,124-127} provided further evidence that the superior crack propagation resistance of Al-Li alloys is associated with crack tip shielding mechanisms induced by inhomogeneous planar slip deformation. Increasing the Li/Cu ratio, and hence the amount of δ' , was found to enhance crack path tortuosity, which in turn resulted in lower crack extension rates and higher fatigue thresholds (Fig. 13). Conversely, δ'



- 13 Growth rate behaviour of long fatigue cracks in experimental Al-Li-Cu-Zr alloys under constant amplitude loading for $R=0.05$ and 0.5 in *vacuo* at room temperature. Note that increasing Li/Cu ratio leads to slower crack velocities, particularly at low R (Ref. 124)**

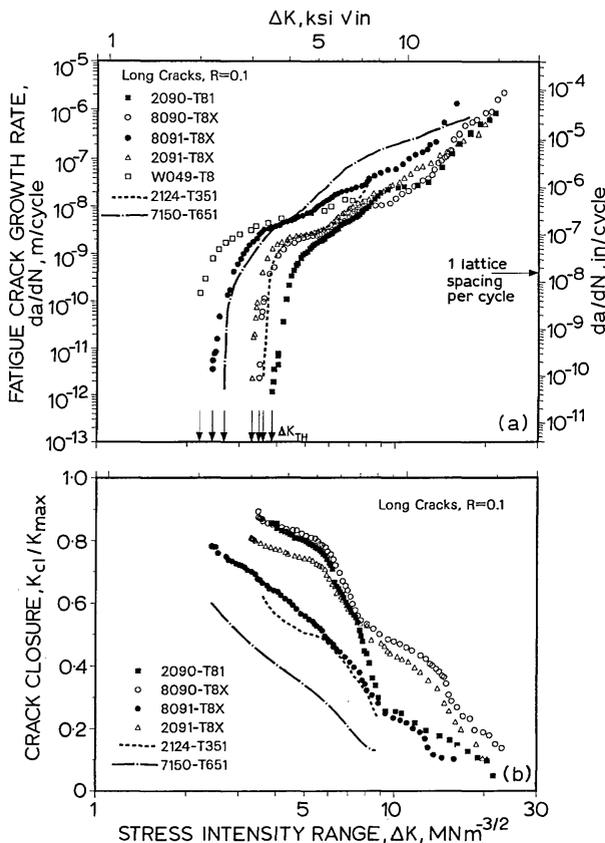
reversion and reprecipitation studies⁴⁴ indicate that T_1 plates inhibit planar slip induced crack deflection, thereby resulting in faster crack growth rates. Growth rate behaviour was also found to be sensitive to load ratio; alloys with higher Li content showed the strongest effect.¹²⁴

Behaviour in advanced commercial Al-Li alloys

The design of modern Al-Li alloys has principally focused on homogenising planar-slip deformation by adding solute elements that form incoherent precipitates and dispersoids; the rationale is to achieve acceptable ductility and toughness. The resulting microstructures are often unrecrystallised and highly textured with a laminated grain structure. Such microstructures can lead to exceptional fatigue crack propagation resistance in specific orientations, yet concurrently often result in poor short-transverse toughness and fatigue properties.^{68,70,115,123,128-138}

Crack growth rate results, as a function of ΔK , are plotted in Fig. 14a for commercial IM Al-Li alloys, namely peak aged 2090, 8090, 8091, 2091, and Weldalite (L-T orientation, $R=0.1$), and are compared with data on 2124-T351 and 7150-T651.^{68,123} Corresponding crack closure levels, measured using back-face strain compliance methods and plotted as K_{cl} values normalised by K_{\max} , are shown in Fig. 14b. As with experimental Al-Li-Cu-Zr alloys, the commercial alloys exhibit consistently slower crack velocities over the entire spectrum of growth rates (except compared with 2124-T351 at near-threshold levels). As shown in Fig. 14b, this can be attributed primarily to their higher crack closure levels which, unlike traditional 2124 and 7150 alloys, remain significant at high stress intensities, e.g. values of K_{cl} in 2090-T81 are 90% of K_{\max} ($K_{cl} \sim 2-3 \text{ MN m}^{-3/2}$) at the threshold ΔK_{TH} , and remain over 50% of K_{\max} ($K_{cl} \sim 3-4 \text{ MN m}^{-3/2}$) at ΔK levels as high as 7-8 $\text{MN m}^{-3/2}$ (Ref. 123).

Such high crack closure levels can be associated with highly tortuous and deflected crack paths,



14 Constant amplitude *a* fatigue crack growth rates and *b* corresponding crack closure levels for long cracks in peak aged commercial Al–Li and traditional Al alloys in plate form (L–T orientation, $R=0.1$, moist ambient–air environment, using 6.4 mm thick specimens, $t/2$ location). Growth rate behaviour in Al–Li alloys is superior to traditional Al alloys consistent with higher crack closure levels (Refs. 68, 123)

specifically for 2090, 8090, and 2091 Al–Li alloys (Fig. 15); fracture surface morphologies are therefore unusually rough and covered with transgranular facets.^{68,100,115,123,128–137} At intermediate to high ΔK levels exceeding $\sim 6\text{--}8 \text{ MNm}^{-3/2}$, these facets are associated with slip band cracking along $\{111\}$ planes; however, at low ΔK levels, ‘pseudocleavage’ type cracking along $\{100\}$ or $\{110\}$ planes has been observed.^{133–137} In fine grained 8091, however, which exhibits the lowest closure levels and consequently the fastest growth rates, crack paths are essentially linear; in 2090–T81 conversely, which displays the most deflected crack paths, closure levels are the highest and corresponding growth rates are the slowest.¹²³ This implies that tortuosity of the crack path provides a major contribution to crack tip shielding in Al–Li alloys; additional shielding may arise from closure induced by cyclic plasticity¹⁰⁸ and corrosion debris,^{111,120–122,124,125} though the latter mechanism is relatively insignificant.^{41,135}

Such shielding mechanisms are common to most coherent particle hardened aluminium alloys, though their influence is generally limited to near-threshold stress intensity levels.^{120–122} What is particularly striking about Al–Li alloys is that the predominance

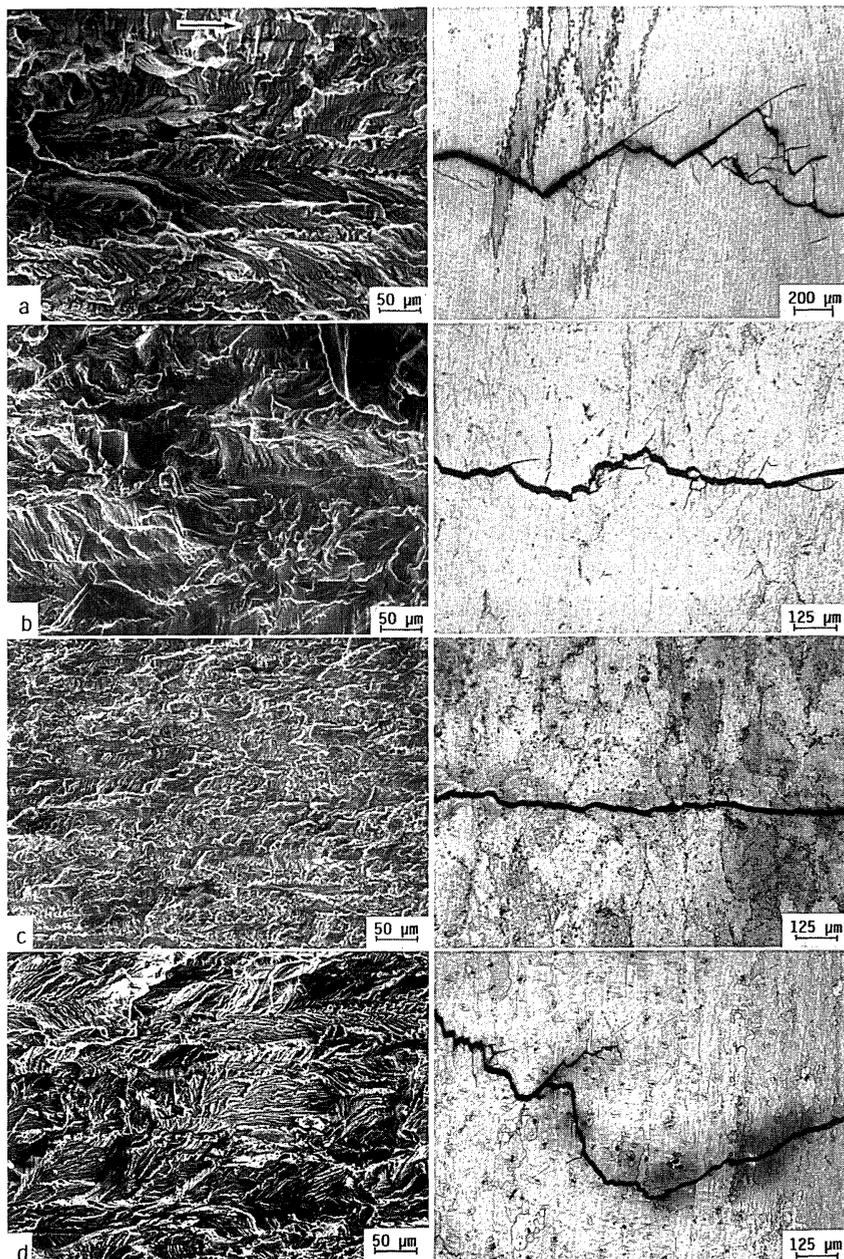
of $\{111\}$ slip band cracking preserves such high closure levels into the intermediate and high growth rate regimes;^{137,138} at near-threshold levels, where $\{100\}$ or $\{110\}$ pseudocleavage cracking occurs, behaviour in Al–Li alloys is comparable with conventional Al alloys.^{134,135} Moreover, since $\{111\}$ cracking is promoted by strong deformation texture, early vintage 2090 alloys, for example, tend to display the best fatigue crack growth properties.¹¹⁵ However, the onset and relative proportion of $\{111\}$ v. $\{100\}$ or $\{110\}$ type crystallographic fatigue cracking modes with respect to ΔK level can change in chloride environments, as discussed in the section ‘Behaviour in aqueous chloride environments’, below.^{134,136}

The effect of the higher elastic modulus, which reduces the CTOD, and increased slip reversibility, which reduces the crack tip damage per cycle, act in concert with mechanisms of crack tip shielding; however, in light of the high crack closure levels and observed effects of crack size, load ratio, and loading sequence on crack growth kinetics (as discussed below), the role of shielding appears to be more dominant. In essence, the tortuous crack paths in Al–Li alloys lead to slower (long crack) growth rates by increasing the path length of the crack, by reducing the local stress intensity levels due to deflection from the principal stress plane,¹³⁹ and by reducing the near-tip stress intensity range by enhancing roughness induced crack closure caused by asperity wedging^{109–112} (Fig. 16).

Crack path morphology

As noted above, an essential feature of the fatigue behaviour of Al–Li alloys is the shearable nature of δ' strengthening precipitates, which results in inhomogeneous (planar-slip) deformation and unusually tortuous crack path morphologies. The nature of the deflected crack paths takes several forms, as shown in Fig. 17: the crack may undergo macroscopic branching, where the entire crack deflects at some angle to the principal stress plane because of deformation texture; alternatively, crack growth may be faceted (‘zig-zagged’) from crystallographic deflection at boundaries because of the marked planarity of slip, or cracks may undergo (intergranular) short-transverse delamination because of the unrecrystallised, anisotropic ‘laminated’ grain structure.¹¹⁵

Additionally, pronounced deformation textures in Al–Li alloys promote deflection along the crack front. This is illustrated in Fig. 18 by a metallographic section of a fatigue fracture surface in 2090–T81, taken across the specimen thickness perpendicular to the crack plane and crack growth direction; the crack front is highly deflected and faceted, with apparent secondary crack branching beneath the surface. Texture analysis shows that the crystallographic facets (with an included angle of $\sim 60^\circ$) result from a change in slip plane orientation between two components of the $\{110\}\langle 112 \rangle$ deformation texture, resulting in cracking along two sets of intersecting $\{111\}$ slip bands.^{62,63} However, precise modelling of the $\{111\}$ slip band cracking mechanism occurring along planes with the highest resolved deviatoric stress is currently lacking. Combined with small mode III shear dis-



a 2090-T81; *b* 8090-T8; *c* 8091-T8; *d* 2091-T8

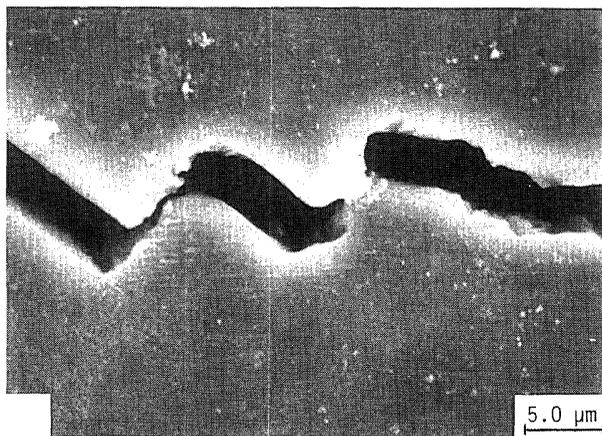
- 15** SEM (left) and optical (right) micrographs of fatigue fracture surfaces and crack path morphologies (L-T orientation, $R=0.1$, $t/2$ location) at ΔK levels between 6 and 8 $\text{MN m}^{-3/2}$. Note rough fracture surfaces and deflected crack paths in 2090, 8090, and 2091, compared with relatively flat surface and linear crack path in fine grained 8091. Arrow indicates general direction of crack growth (Ref. 123)

placements, this cracking morphology promotes roughness induced closure through the specimen thickness, in addition to mode II shear induced wedging of asperities along the crack growth direction. The crystallographic facets, however, become less sharp (the included angle approaches 180°) close to the specimen surface, consistent with the through thickness variations in textural components for wrought Al-Li alloys.⁶³

Microstructural effects

The influence of microstructure on fatigue crack growth in Al-Li alloys is essentially not unlike other

precipitation strengthened aluminium alloys. Resistance to crack growth is generally superior in under-aged structures, because the marked planarity of slip from coherent (shearable) precipitate distributions promotes deflected crack paths and hence high closure levels.^{35,68,115,123-128} With artificial aging, deformation becomes more homogeneous concurrent with loss of coherency of the precipitates; crack growth resistance is now diminished somewhat as the degree of slip reversibility, crack path tortuosity, and resulting crack closure levels all are reduced. In Al-Li alloys, however, δ' precipitates remain coherent up to large particle diameters; superior fatigue crack



16 Meandering fatigue crack paths and resultant wedging of fracture surface asperities (termed roughness induced crack closure) in the plane of loading for Al-Li alloys. Micrograph obtained from specimen surface of 2090-T81 in L-T orientation; horizontal arrow indicates general direction of crack advance (Ref. 123)

propagation resistance can be maintained in coarse grained microstructures even in peak aged tempers, provided grain boundary precipitation and associated PFZ formation are largely suppressed.^{68,115,123,127,128}

The close correlation between optimum crack growth resistance and high closure levels in Al-Li alloys, as illustrated by the T8 tempers of 2090, 8090, 8091 and 2091 in Figs. 14 and 15, implies that if (long crack) growth rates are compared as a function of ΔK_{eff} (after correcting for closure), behavioural differences with respect to microstructure and alloy composition will be less apparent (Fig. 19). This notion¹²³ is corroborated by a comparison of growth rates determined under constant K_{max} /variable R loading conditions* (Fig. 20). Al-Li-Cu alloys now exhibit similar growth rate behaviour to 2000 series aluminium alloys, though faster growth rates are observed for 7000 series alloys, presumably because of their increased environmental sensitivity.¹³⁵⁻¹³⁷

This should not be construed as an indication that microstructure and alloy composition are unimportant to fatigue crack propagation in Al-Li alloys since the microstructure clearly affects crack deflection, which in turn influences closure. In fact, coherent δ' -particle strengthening, strong deformation textures, coarse and unrecrystallised (elongated) grain structures, absence of heterogeneous grain boundary precipitation, and PFZs can all be identified as the prominent microstructural sources of crack deflection and branching.

* Constant K_{max} /increasing K_{min} procedures permit an estimation of the intrinsic crack growth behaviour by minimising closure effects caused by crack wedging. As such, they result in crack growth data similar to that characterised in terms of ΔK_{eff} . However, these techniques do not evaluate the influence of any other shielding mechanisms, such as crack deflection or bridging, and moreover, require prior knowledge of the variation in K_{cl} with ΔK level and R .

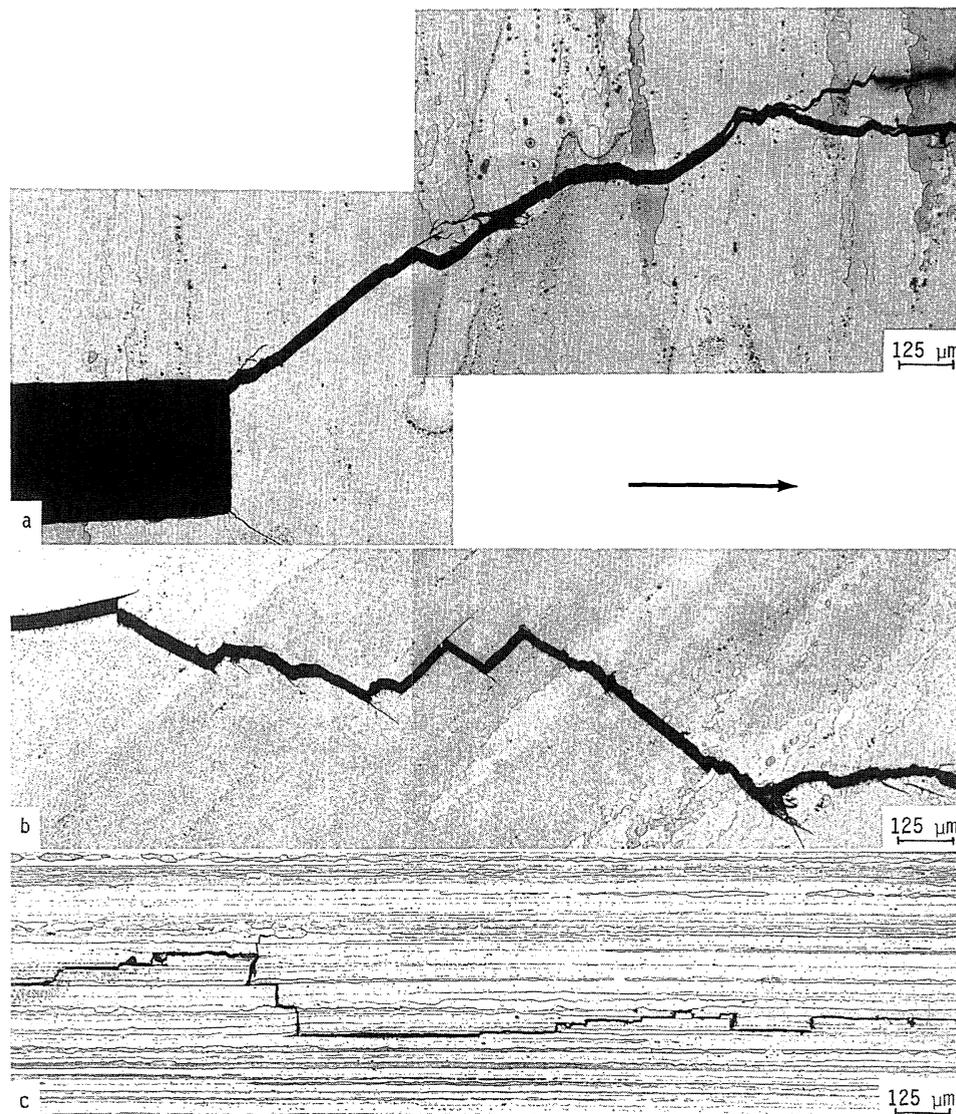
Load ratio effects

It is well known that increasing the load ratio at a specific stress intensity range can lead to faster crack propagation rates and lower fatigue thresholds. At near-threshold levels, the effect is primarily associated with a diminished role of crack closure induced by wedging mechanisms from the larger CODs at high mean stresses. In Al-Li alloys, growth rates (plotted as a function of ΔK) are particularly sensitive to the load ratio because closure levels are exceptionally high;^{68,115,123-128} however, by plotting as a function of ΔK_{eff} , after allowing for closure, the effect of load ratio becomes less apparent (Fig. 21). Moreover, because of this limited role of closure at high load ratios, the superior (long crack) fatigue crack growth properties of Al-Li alloys are less evident at high R .^{123,135-137} Similarly, as shown in Fig. 20, if growth rates in Al-Li alloys are measured under constant K_{max} /variable R conditions which minimise crack closure, behaviour is equivalent to that obtained for 2000 series alloys.¹³⁵⁻¹³⁷ The superiority of Al-Li alloys can also be compromised under completely reversed or tension-compression loading ($R < 0$) conditions, where the compressive cycles can similarly act to limit closure by crushing fracture surface asperities.¹³⁸

Plate orientation effects

As fatigue crack growth rates in Al-Li alloys are a strong function of crack path morphology, behaviour in commercially rolled sheet and plate products is strongly anisotropic,^{63,67,70,115,140} both in the rolling plane and through the thickness. Growth rates for long cracks (at fixed ΔK) in 13 mm thick 2090-T81 rolled plate can vary by up to four orders of magnitude between various orientations, concomitant with large differences in measured crack closure levels¹¹⁵ (Fig. 22). Specimen orientations in the rolling plane, namely L-T and T-L, develop the highest closure levels from highly deflected transgranular crack path morphologies (K_{cl} values approach 90% of K_{max}); these orientations correspondingly show the slowest growth rates and highest thresholds ($\sim 3-4 \text{ MN m}^{-3/2}$). Similarly, crack growth resistance is high for the T-S orientation where crack growth is often deflected through $\sim 90^\circ$ by delamination along weakened short-transverse grain boundaries.^{67,115} However, intrinsic crack growth rates for the L-T, T-L, and T-S orientations are all identical as the mechanism of crack propagation remains essentially transgranular;¹³³⁻¹³⁵ differences are associated with the variations in shielding. Conversely, crack paths for the S-L and S-T orientations are highly linear and involve delamination along grain boundaries, often weakened by the presence of Fe- and Cu-rich intermetallics, δ' -PFZs, and possible Li, Na, or K segregation; growth rates in these orientations are accordingly far faster than for the L-T, T-L, and T-S orientations.¹¹⁵

In addition to the short-transverse orientations, the $L+45^\circ$ (LT-TL) orientation also shows high growth rates^{67,70,115} and low thresholds ($\Delta K_{\text{TH}} = 2.4 \text{ MN m}^{-3/2}$), behaviour which appears to be associated with pronounced deformation texture in wrought products. Slip band cracking is favoured along close



a macroscopic crack branching; *b* crack meandering (crystallographic slip band cracking); *c* intergranular delamination (short-transverse) cracking

17 Various modes of non-linear cracking morphologies observed during fatigue crack propagation in commercial Al–Li alloys. Micrographs taken for 2090–T81 plate; arrow indicates crack growth direction (Ref. 115)

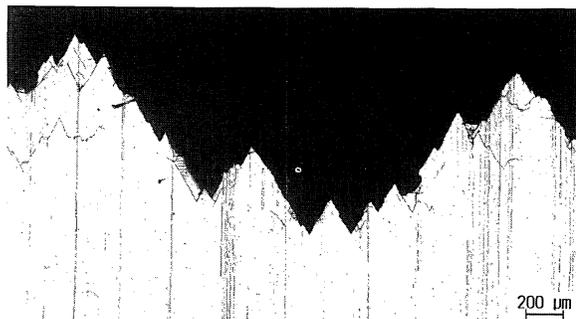
packed $\{111\}$ planes which are aligned closer to the principal stress plane in the $L+45^\circ$ configuration; consequently, crack paths show a lower tendency for deflection and meandering.¹¹⁵ Fatigue cracking mechanisms in the $L+45^\circ$ orientation have not been fully examined, though reduced modulus ($\sim 6\%$) and lower strength ($\sim 16\%$) in this configuration can additionally contribute to faster crack growth rates.

Fatigue crack growth rates in Al–Li alloys also tend to be substantially slower (at fixed ΔK) in the mid-thickness locations of wrought thick section plate compared with the surface regions,^{63,67} again associated with marked variations in grain structure and crystallographic texture. For example, in 2090–T81 plate, surface regions are often partially recrystallised and grains exhibit a weak $\{001\}\langle 110\rangle$ texture; conversely, at mid-thickness, the microstructure remains unrecrystallised with strong brass, copper, and weak S type texture components. Since the growth rate

kinetics are a strong function of crack path tortuosity, which is promoted by unrecrystallised and textured microstructures, growth rates are retarded at mid-thickness locations by the increased effect of crack closure.⁶³

Small crack effects

Current design methodologies often compute the lifetimes of structural components in terms of the cycles to propagate a pre-existing flaw to failure through integration of experimental crack growth relationships (e.g. equation (1)) derived from standard specimen geometries containing long cracks, typically in excess of 5–10 mm in size. Although presumed to be conservative, this approach can potentially overestimate lifetimes in the presence of small cracks, which are small compared with microstructural dimensions or the plastic-zone size, or are simply

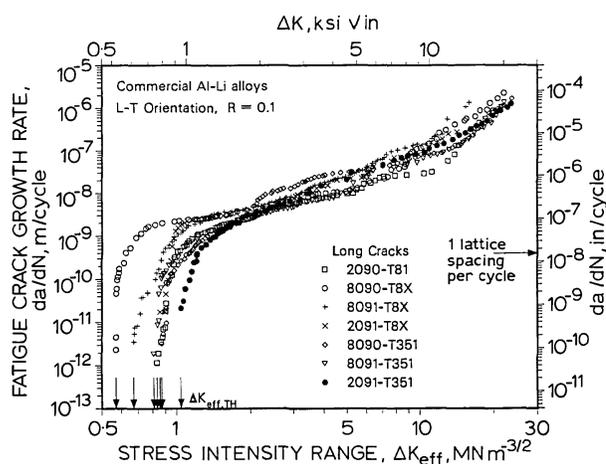


18 Crystallographic crack path tortuosity through specimen thickness observed during fatigue crack growth in Al-Li alloys induced by pronounced deformation texture. Sharp facets with an included angle of $\sim 60^\circ$ are a result of shearing along intersecting (111) planes across several high angle grain boundaries. Micrograph obtained for 2090-T81 plates (L-T orientation) at mid-section thickness normal to the crack growth direction and crack growth plane; $\Delta K = 8\text{--}10 \text{ MN m}^{-3/2}$; crack growth direction is into page

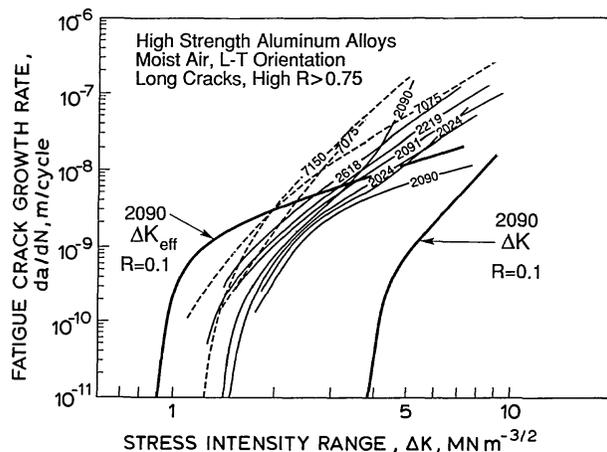
physically small (i.e. $< 1 \text{ mm}$);* such cracks are known to propagate faster than long cracks at equivalent stress intensity levels, in part owing to a restricted role of crack tip shielding (e.g. crack closure) with cracks of limited wake.¹⁴¹⁻¹⁴³ This effect has particular significance for Al-Li alloys because their superior (long crack) fatigue crack growth resistance arises primarily from (roughness induced) crack closure, a phenomenon which is strongly crack size dependent.

Small crack behaviour is generally characterised by monitoring the early growth of naturally occurring

* Mechanistically, a crack is considered 'small' when its size is less than, or comparable with, the size of the steady state shielding zone in the wake of the crack tip.

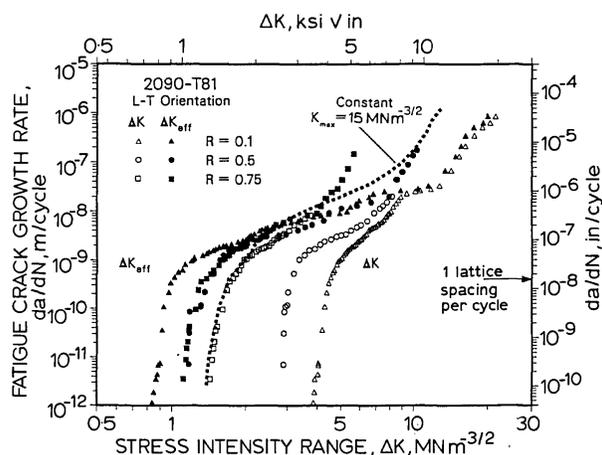


19 Intrinsic fatigue crack propagation resistance of various commercial Al-Li alloy microstructures in moist air environments obtained using variable ΔK /constant $R (=0.1)$ schemes and plotted in terms of ΔK_{eff} , after correcting for closure. Note that growth rate differences between alloys are reduced compared with results presented in Fig. 14 (Ref. 123)

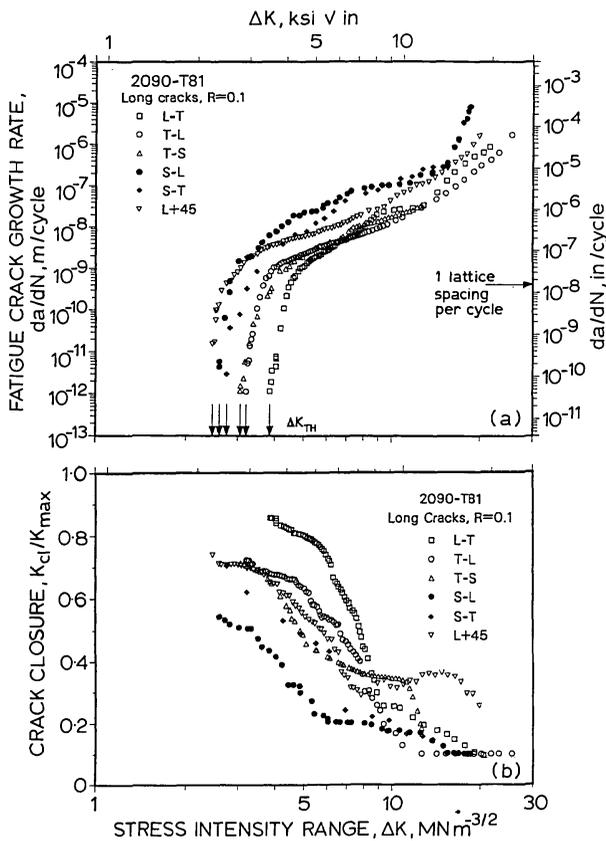


20 Intrinsic fatigue crack propagation rates in commercial Al-Li, and 2000 and 7000 series high strength Al alloys in moist air, obtained by using constant K_{max} /variable R loading techniques (R increases from 0.1 at the highest ΔK to about 0.9 at ΔK_{TH}) or by measuring crack growth rates at high R . All alloys show comparable fatigue resistance; 7000 series alloys are slightly more sensitive to environment (Refs. 135-137)

surface flaws (typically in the range $2\text{--}1000 \mu\text{m}$ in size) in unnotched bend or tension samples using optical, replication, or electrical potential techniques (see Ref. 143 for details). Using such procedures, studies in commercial Al-Li alloys 2090, 2091, 8090, and 8091 have indicated that small cracks initiate along slip bands, predominantly at cracked Fe- or Cu-rich inclusions or constituent particles on the surface, polishing pits, and uncracked matrix/particle interfaces.^{123,144-151} During early stages of small-crack growth, fracture surfaces exhibit large 'cleavage-like' features,¹⁴⁵ resembling those observed at near-threshold ΔK levels for long cracks. Subsequent crack



21 Influence of load ratio R on (long) crack extension rates in 2090-T81, as function of ΔK and ΔK_{eff} ; results obtained using constant K_{max} /variable R schemes ($K_{\text{max}} = 15 \text{ MN m}^{-3/2}$, $R = 0.1\text{--}0.9$) are also compared. Note that differences in growth rates with respect to R are reduced when data are characterised using ΔK_{eff} or determined using the constant K_{max} /variable R approach (Refs. 115, 133-137)

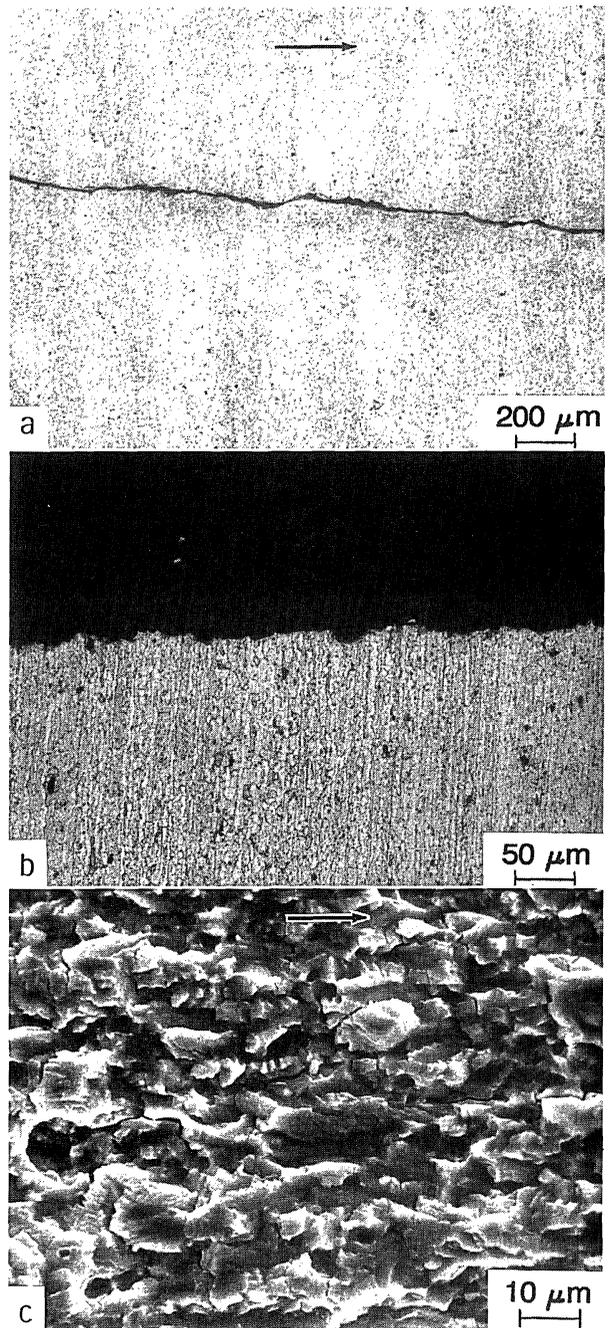


22 Variation in *a* fatigue crack extension rates and *b* crack closure levels for long cracks in 2090-T81 alloy at $R=0.1$, as function of specimen orientation. Note how growth rates parallel to rolling plane (S-L, S-T) are fastest, whereas growth rates normal to rolling plane (L-T, T-L, T-S) are slowest, consistent with non-linearity of crack path morphologies and resulting levels of crack closure (Ref. 115)

advance is marked by frequent deflection of the crack path at intersecting slip bands and high angle grain boundaries (Fig.23).^{144,145,151} At higher cyclic stress levels, several cracks may nucleate along coarsened slip bands, which tend to link up rapidly to form a network.¹⁴⁵

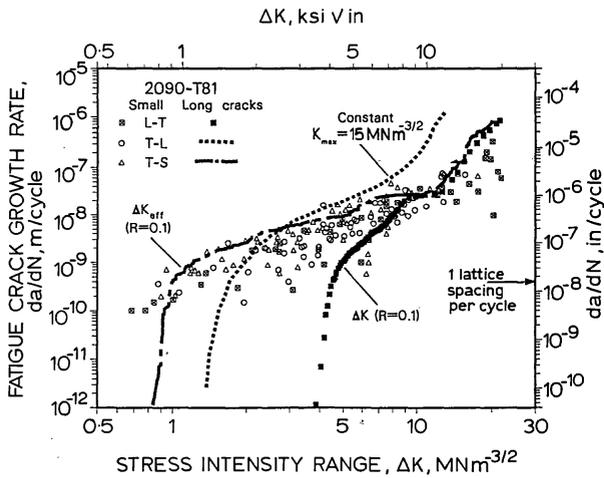
The growth rate response of such microcracks is compared with conventional long crack data in Fig. 24, as a function of ΔK (the use of K here is considered appropriate as computed maximum and cyclic plastic-zone sizes remain small (typically 4%) compared with crack size). Despite the large scatter in experimental results, it is clearly apparent that small cracks propagate at rates as much as 2-3 orders of magnitude faster than long cracks at equivalent applied stress-intensity levels;^{144,151} however, uncertainties do exist in the computation of ΔK for small cracks, owing to the stress concentrating effect of nucleating cracks near inclusions and to irregular crack geometries.¹⁴¹⁻¹⁴³ Nevertheless, small crack growth is observed at ΔK levels as low as $0.8 MN m^{-3/2}$, far below the threshold ΔK_{TH} for long cracks.¹⁴⁴⁻¹⁵¹

A major reason for the discrepancy between long and small crack behaviour is the restricted crack tip shielding associated with flaws of limited wake.¹⁴⁴⁻¹⁵²



23 SEM micrographs illustrating growth morphology of naturally occurring, microstructurally small (2-1000 μm) surface fatigue cracks in Al-Li alloys. As shown in *a* they initiate at local heterogeneities in microstructure, propagate by periodically deflecting *b* along intersecting slip bands, and *c* at grain boundaries. Micrographs were obtained from gold coated cellulose-acetate replicas of specimen surface for 2090-T81: horizontal arrow represents direction of loading (Ref. 123)

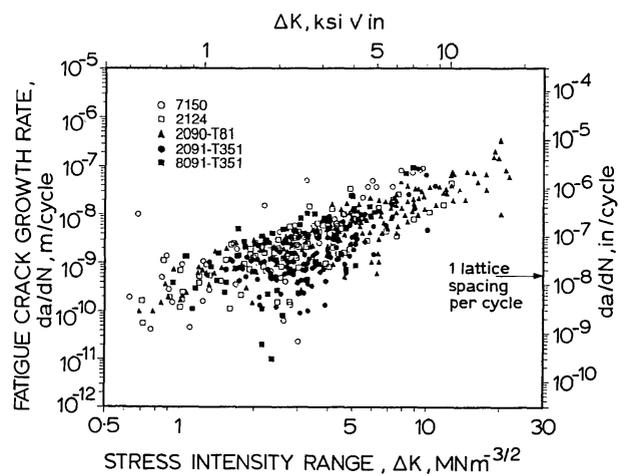
Since shielding mechanisms act primarily on the crack wake, small cracks generally are unable to develop the same degree of closure as a crack of a length larger than the equilibrium shielding zone (where, for example, fracture surface asperity contact is achieved); thus, at the same nominal ΔK , the small crack may



24 Growth rate behaviour of microstructurally small (2–1000 μm) surface, and long (>10 mm) fatigue cracks in peak aged 2090 Al-Li alloy. Note that at ΔK levels below 7–8 $\text{MN m}^{-3/2}$, small crack velocities are nearly 2–3 orders of magnitude faster than long crack da/dN at $R=0.1$. Closer correspondence between long and small crack data is seen when long crack results are characterised using ΔK_{eff} or measured using constant K_{max} /variable R schemes (Refs, 144, 151)

experience a larger local stress intensity and hence extend at a faster rate. In fact, *in situ* compliance measurements in 8090 and 2090 confirm that closure levels for small cracks are smaller than for corresponding long cracks.^{145,147} Accordingly, if long and small crack results are characterised in terms of ΔK_{eff} (after correcting for closure), or if long crack growth rates are evaluated using constant K_{max} /variable R procedures (which minimise the closure effect), the resulting data display a much closer correspondence, as shown for 2090 in Fig. 24. With increasing crack size, an equilibrium shielding zone is established in the crack wake, such that closure levels are similar, and small and long crack data merge (even when characterised in terms of ΔK); in commercial Al-Li alloys, this occurs typically at crack sizes above $\sim 700\text{--}1000\ \mu\text{m}$ and nominal ΔK levels above $\sim 7\ \text{MN m}^{-3/2}$.^{144,146} However, it should be noted that there is not a perfect match between long and small crack data at near-threshold levels. This may result from scatter, uncertainties in ΔK computation, and/or the fact that naturally occurring small cracks initiate at local heterogeneities in the microstructure which are not necessarily simulated in a long-crack test involving a pre-existing flaw and artificial load shedding schemes.

In contrast to long crack results where microstructural effects strongly influence crack growth behaviour (Figs. 12–14), a minimal influence of alloy chemistry and microstructure is apparent for small cracks in Al-Li alloys,^{123,135,137,151} in fact, despite extensive scatter, their small crack behaviour is comparable with that of traditional alloys^{123,150,151} (Fig. 25). This implies that the salient role of microstructure during fatigue crack propagation in these alloys occurs principally through the development of crack tip shielding



25 Growth rate response of naturally occurring, microstructurally small surface cracks in commercial Al-Li and traditional alloys. Note that results show no apparent influence of microstructure and composition; small crack resistance of Al-Li alloys remains comparable with other high strength Al alloys (Ref. 1513)

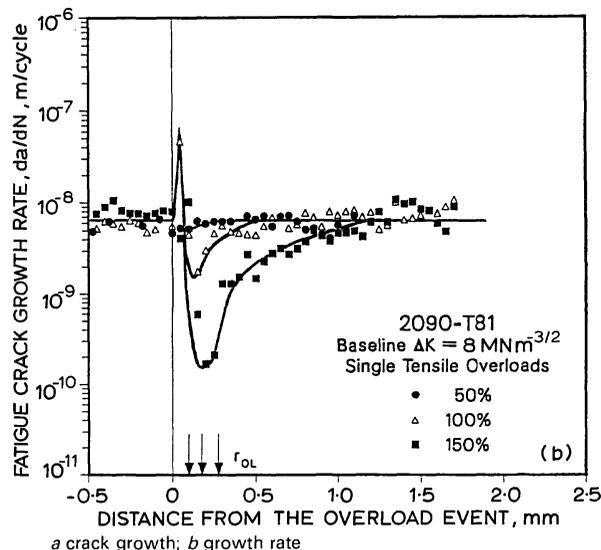
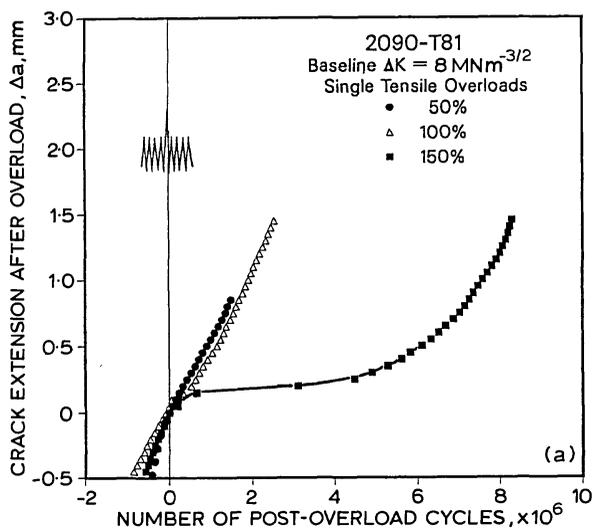
(crack deflection and closure). This is consistent with the fact that microstructural effects become less evident where the shielding effects are suppressed, i.e. at high load ratios, with small cracks, or where results are characterised in terms of ΔK_{eff} or using constant K_{max} /variable R procedures.¹⁵² Accordingly, in contrast to long crack behaviour, it is to be expected, and indeed is the case, that small crack growth rates are relatively insensitive to load ratio and specimen orientation.^{135,144,151} However, since the long crack behaviour of Al-Li alloys is, in general, superior to conventional aluminium alloys (because of extensive shielding), differences between long and small crack results are most significant in these materials. Thus, Al-Li alloys can be considered to display the best long crack properties of all high strength aluminium alloys yet, by comparison, the worst small crack properties, paradoxically for the same reason, i.e. crack tip shielding.¹⁵¹

Behaviour under variable amplitude loading

The design and selection of materials for aircraft structures is increasingly based on fatigue crack growth response under complex loading spectra which simulate in-service operating conditions. This is particularly important since the fatigue performance of a material can vary widely with the specific nature of these spectra. For example, tensile overload cycles of sufficient magnitude can act to retard crack growth and hence prolong fatigue life; compression overloads, on the other hand, tend to diminish this effect. The behaviour of Al-Li alloys in this respect is discussed below.

Influence of tensile overloads

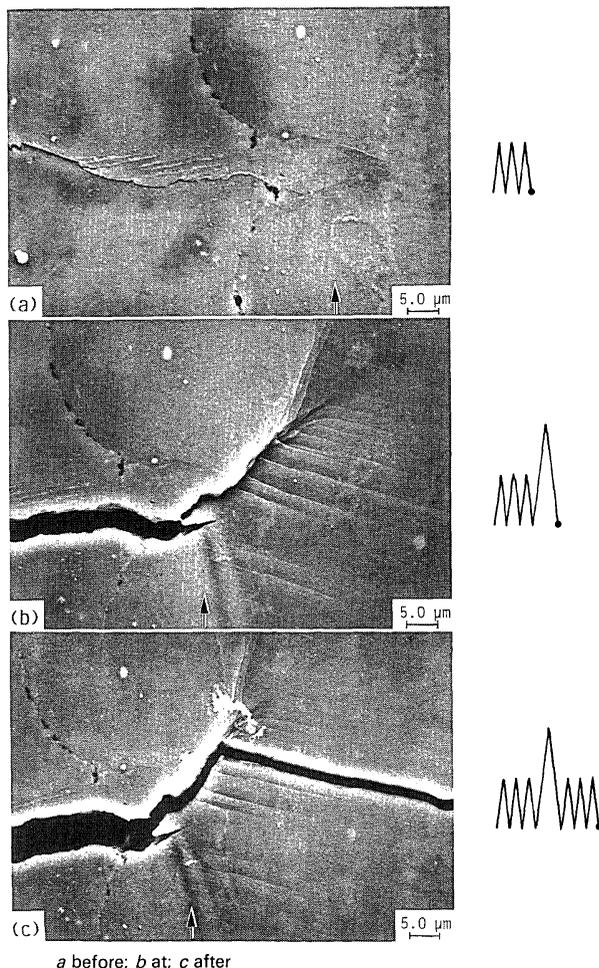
Experiments to examine the role of single tensile overloads are generally performed on fatigue cracks propagating under constant ΔK conditions by moni-



26 Transient fatigue response, as function of crack extension following tensile overloads in 2090-T81 (L-T orientation) at a constant baseline ΔK of $8 \text{ MN m}^{-3/2}$. Retardation distances following overload are typically 2-3 times computed maximum overload plastic-zone size, r_{OL} , as indicated by vertical arrows (Ref. 153)

toring the transient crack growth response, after resuming cycling at baseline stress intensity levels. Typical data for 2090-T81, showing the effect of 50, 100, and 150% single tensile overloads on crack propagation rates at a constant baseline ΔK of $8 \text{ MN m}^{-3/2}$, are plotted in Fig. 26 (Ref. 153). It is evident that 50% overloads have little effect; overloads of higher magnitude, however, result in an immediate acceleration followed by a period of prolonged retardation before growth rates return to baseline levels. The extent of the delay increases with the magnitude of the overload, and at near-threshold levels can completely arrest crack advance.¹⁵³

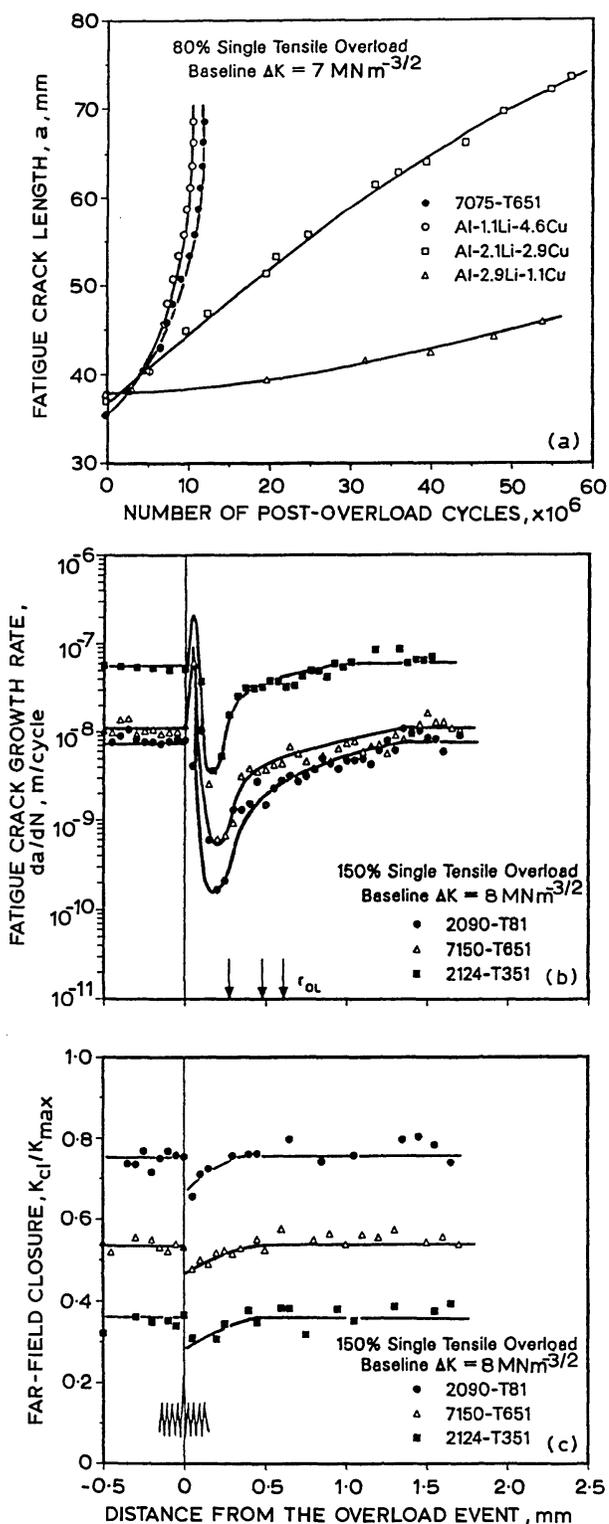
Mechanistically, studies on Al-Li and titanium alloys¹⁵³⁻¹⁵⁶ have indicated that the post-overload retardation is associated with residual compressive stresses in the overload plastic zone, and with crack closure which results in the immediate



27 SEM images of surface fatigue crack morphology, showing crack tip during application of a 100% single peak tensile overload in 2090-T81 at baseline ΔK level of $8 \text{ MN m}^{-3/2}$ ($R=0.1$). Imaging was performed after unloading; vertical arrows indicate position of overload event (Ref. 153)

vicinity of the crack tip as the crack grows into the zone; this effect is enhanced by lower CTODs in the overload zone which promote wedging of the mating fracture surfaces. Crack deflection and branching along flow bands (Fig. 27), though predominantly a surface effect, may also contribute to the delay.¹⁵³

Because Al-Li alloys develop such high shielding levels, they generally show a greater degree of post-overload retardation than traditional aluminium alloys.^{101,124,153-158} This behaviour is illustrated in Fig. 28a, where fatigue crack extension following an 80% tensile overload is plotted as a function of the number of cycles for three experimental Al-Li-Cu-Zr and 7075-T6 alloys.¹²⁴ Although the Al-Li alloys with low Li/Cu ratio show only marginally increased delays compared with 7075, alloys with higher Li/Cu ratio clearly exhibit far superior variable amplitude, fatigue crack propagation resistance. Mechanistically, the large improvement in variable amplitude fatigue properties with increasing lithium content is consistent with the larger volume fraction of δ' precipitates in these microstructures; this induces crystallographic crack growth which in turn promotes crack deflection



a crack extension; b crack growth rates; c far field, crack closure levels.

28 Post-overload, fatigue crack propagation behaviour following single tensile overloads in Al-Li alloys compared with other Al alloys. Note greater delay periods and retardations in higher Li-containing alloys, and specifically 2090, consistent with higher closure levels (Refs. 124, 153)

and roughness induced closure. Similar crack growth and crack closure behavioural trends have been reported¹⁵³ for commercial 2090-T81 Al-Li alloy

when compared with traditional 2124-T351 and 7150-T651 alloys (Fig. 28b and c).

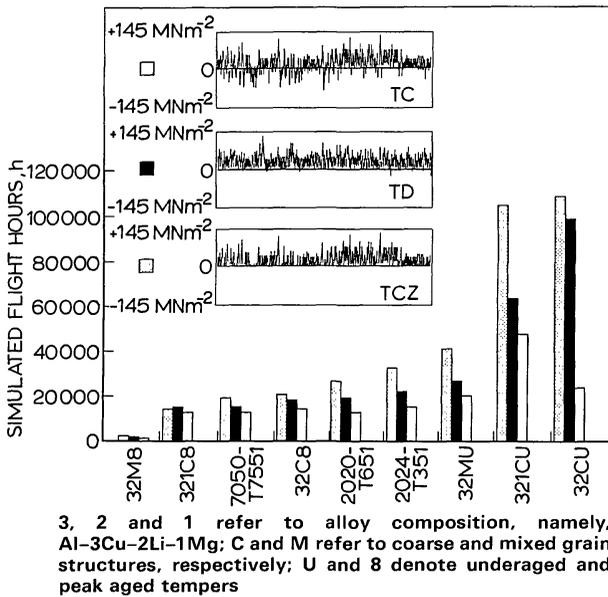
Compression overloads

Although compressive overloads are commonly experienced by numerous aircraft components in service, their influence on fatigue crack propagation has received surprisingly little attention. As the crack is presumed to remain closed during the negative portion of the loading cycle, the effect of compression cycles has long been considered minimal. Recent results on high strength steels, Al alloys, and Al-Li alloys, however, have indicated that periodic compression overloads can significantly accelerate near threshold crack growth rates, specifically by reinitiating the growth of long fatigue cracks previously arrested at ΔK_{TH} , or by reducing the fatigue threshold.^{138,159,160} This behaviour has been associated with diminished levels of crack tip shielding, resulting from a reduction in residual compressive stresses in the cyclic plastic zone and, more importantly, from the compaction of the fracture-surface asperities during the overload cycle. In view of their increased dependence on crack deflection and asperity wedging closure mechanisms, Al-Li alloys are far more sensitive to compressive overloads than traditional aluminium alloys; for example, it takes 400–500% compressive overloads to reinitiate the growth of arrested fatigue cracks in traditional 2124 and 7150 alloys, whereas a mere 200% overload is sufficient to cause the same effect in Al-Li alloy 2090-T81 (Ref. 138).

Spectrum loading

Comprehensive studies at Northrop^{161,162} on the spectrum fatigue resistance of several 2000 and 7000 series aluminium alloys and experimental Al-Li alloys provide further corroboration of the differing response of Al-Li alloys to tension and compression overloads. Specifically, where behaviour was compared for two load spectra, dominated either by tension overloads (TD)* or by tension and compression overloads of approximately equal magnitude (TC)*, Al-Li alloys can show a nearly fivefold improvement in spectrum fatigue lives compared with 2024 and 7050 under TD loading, consistent with the large crack growth retardations observed following the tensile overloads (Fig. 29). However, alloys with the highest fatigue lives under such TD spectra also show the largest differences between TD and TC spectra, consistent with the diminished role of shielding following compression cycles. In fact, using a modified TC spectrum which eliminates all compressive loads (TCZ), the spectrum fatigue life of all alloys was increased relative to either TD or TC spectra. For example, an under-aged Al-3Cu-2Li-0.12Zr alloy, which exhibited lifetimes greater than 100 000 flight hours under the TD spectrum, showed lifetimes less than 25 000 flight hours under the TC spectrum, a mere 25% of the TD life; under the modified TCZ spectrum, life was increased to nearly 110 000 flight hours.¹⁶²

* These spectra represent, respectively, the lower root wing and horizontal tail hinge moment load histories of the F-18 fighter.

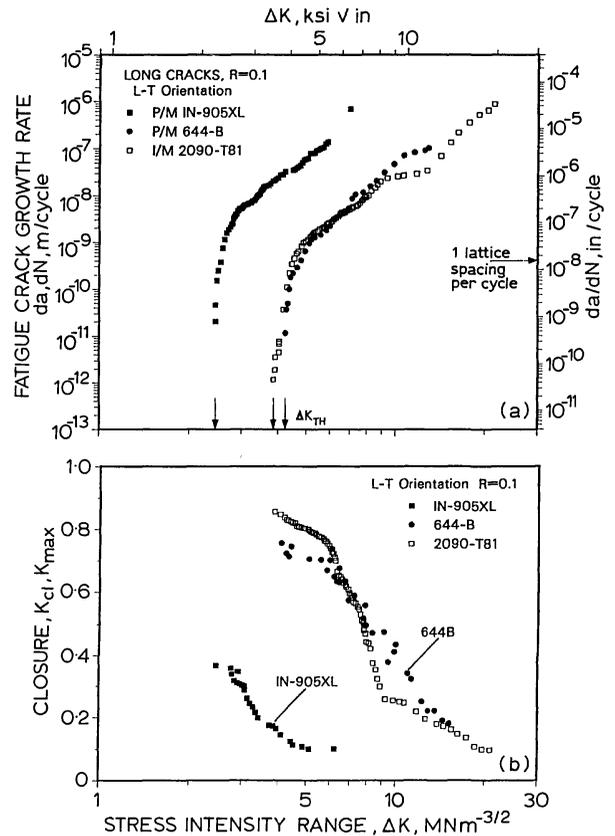


29 Comparison of spectrum fatigue lifetimes (in simulated flight hours) for experimental Al–Li–Cu, 2020–T651, 2024–T351, and 7050–T7551 Al alloys, under tension dominated (TD), tension–compression dominated (TC), and modified tension–compression dominated (TCZ) loading. Estimated lifetimes are based on growth of 6 mm flaw to final failure, for maximum peak stresses below 145 MN m^{-2} . Note that Al–Li alloys show remarkable improvements in lifetimes under TD spectrum loading over 2024 and 7050; behaviour under TC spectrum loading remains comparable (Refs. 161,162)

It is clear that as with behaviour under constant amplitude loading, microstructures that promote shielding, i.e. coarse grained unrecrystallised and underaged coherent particle hardened structures that deform by planar slip and induce deflected crack paths, are preferred for optimal performance under variable amplitude loading conditions.

Powder metallurgy alloys

In addition to the IM alloys discussed so far, several promising Al–Li alloys have been processed by powder metallurgy. The PM route offers flexibility in developing alloys with unique compositions and homogeneous microstructures, via such techniques as rapid solidification and mechanical alloying,^{163–168} as problems of segregation encountered during casting largely confine the chemical compositions of IM alloys to the maximum solid solubility of the alloying elements in aluminium. With PM techniques, Al–Li alloys can potentially be developed with further reduced density and improved strength and stiffness by incorporating higher Li contents exceeding 3%. In addition, high temperature characteristics of Al–Li alloys can be improved by alloying with large concentrations of transition or rare earth elements, such as Zr, Hf, Ti, Ce, or Co. The latter approach, which utilises dispersion hardening for resisting dislocation motion, is also the most effective means of reducing planarity of slip, and hence in improving

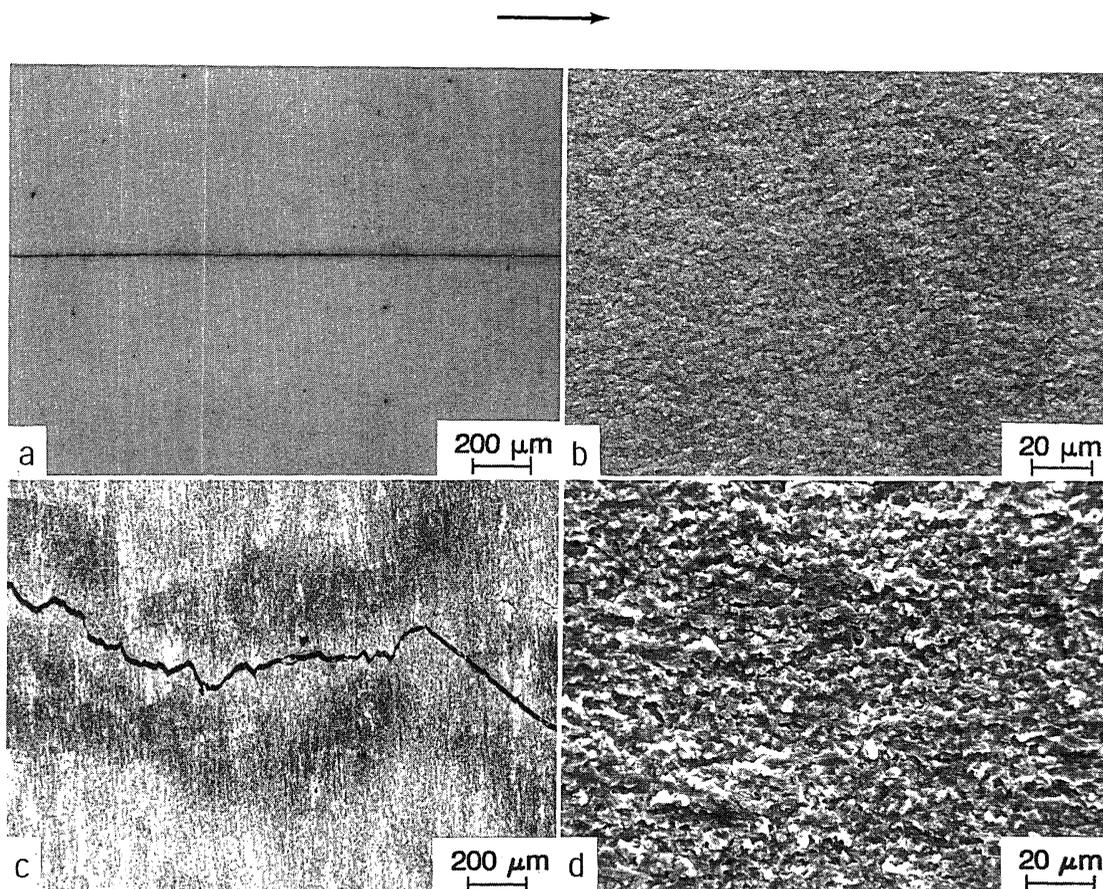


30 Constant amplitude *a* fatigue crack propagation and *b* crack-closure behaviour in PM Al–Li extrusions processed via rapid solidification (Allied Signal 644–B alloy) and mechanical alloying (IN–905XL). Growth rates in IM 2090–T81 plate are also plotted for comparison (Ref. 58)

the ductility and toughness of Al–Li forgings and extrusions, since pre-aging stretching procedures are not feasible for these product forms.

Prominent among the experimental PM alloys are rapid solidification processed (RSP) Al–3.0Li–0.9Cu–0.4Mg–0.4Zr alloy 644–B, developed by Allied Signal, and the mechanically alloyed (MA) Al–1.5Li–4Mg alloy IN–905XL, developed by Inco Alloys Int. The latter alloy also contains substantial amounts of carbon and oxygen in the form of oxide (Al_2O_3) and carbide (Al_4C_3) dispersions, which are formed during ball milling of elemental powders; carbon is added as an organic agent to prevent agglomeration. Microstructural details and mechanical properties are summarised in Tables 2–5.

Such near commercial PM alloys have roughly 10% lower density than traditional 7075 type alloys, and often exhibit far higher strength with moderate ductility and fracture toughness. The HCF and LCF resistance of mechanically alloyed PM extrusions is generally superior to corresponding IM alloys^{90,169} because of the more homogeneous mode of deformation in PM alloys. Moreover, no change in deformation mode is observed for PM Al–Li alloys between the high and low strain amplitudes consistent with the observation of a single slope on the Coffin–Manson plot.⁹⁰ However, crack propagation properties are invariably inferior to IM alloys (Fig. 30*a*);



31 Comparison of fatigue crack path and fracture surface morphologies in *a, b* MA IN-905XL, *c, d* RSP 644-B extrusions at ΔK levels between 6 and 8 $\text{MN m}^{-3/2}$. Crack paths in MA 905-XL are unusually linear, without features of crystallographic crack growth: arrow indicates direction of crack advance (Ref. 58)

fatigue crack growth thresholds are generally lower and crack propagation rates up to one or two orders of magnitude faster.^{58,170–172} The prime reason for this can be traced to a limited role of crack tip shielding. Essentially, the fine microstructures (grain size typically below 5 μm) and more homogeneous deformation of MA Al–Li alloys, which results from dispersion hardening by oxides and carbides rather than by ordered δ' precipitates (Li remains in solid solution), yields unusually linear crack profiles and exceptionally smooth fracture surfaces (Fig. 31*a, b*), with little sign of slip band cracking.^{58,172} As a result, MA microstructures, unlike coarser grained IM alloys, do not develop high shielding levels from deflection and related crack wedging closure mechanisms; in fact, measured K_{c1} values at near-threshold levels are a mere 35% of K_{max} in the Al–Li–Mg 905-XL alloy, compared with a corresponding value of $\sim 90\%$ of K_{max} in IM alloys (Fig. 30*b*). The beneficial effects of lithium on fatigue crack growth are thus not so apparent in MA Al–Li systems.⁵⁸

Rapid solidification processing, conversely, can lead to better fatigue resistance for PM alloys compared with MA microstructures.⁵⁸ For example, the crack growth and closure properties of RSP 644-B extrusions are comparable with IM 2090-T81 plate, particularly at lower ΔK levels ($\leq 10 \text{ MN m}^{-3/2}$). As with IM Al–Li alloys, the superior behaviour of RSP alloys can be attributed primarily to factors that

favour planar slip deformation, namely, precipitation hardening by shearable δ' particles and the presence of β' dispersoids that inhibit recrystallisation and impart preferred grain orientations.⁵⁸ Accordingly, fractographic features in the RSP 644-B alloy resemble those seen in IM Al–Li alloys; crack paths are highly deflected with rough transgranular fracture surfaces 3–4 μm sized facets and evidence of local slip band cracking (Fig. 31*c, d*). In fact, the RSP Al–Li alloy displays superior crack growth properties to most traditional RSP PM aluminium alloys.^{58,164}

In accordance with these arguments, crack growth rates and fatigue thresholds in MA alloys are often found to be less sensitive to load ratio,^{58,171,173} loading sequence,⁵⁸ specimen orientation,⁵⁸ specimen geometry,¹⁷¹ and crack size¹⁷⁴ because of the limited extent of shielding. For example, nominal ΔK_{TH} values in IN-905XL are equivalent to $\Delta K_{\text{eff,TH}}$ and are independent of load ratio for either tension-tension or tension-compression cycling. Similarly, growth rates are similar for small surface cracks ($< 1 \text{ mm}$) and long through thickness cracks in these alloys; in fact, small cracks have been reported to propagate slower.¹⁷⁴ Conversely, faceted crack growth, tortuous crack paths, and high closure levels are all prominent in the RSP 644-B alloy, such that crack growth behaviour in this material becomes sensitive to load ratio and specimen orientation.⁵⁸

High temperature properties

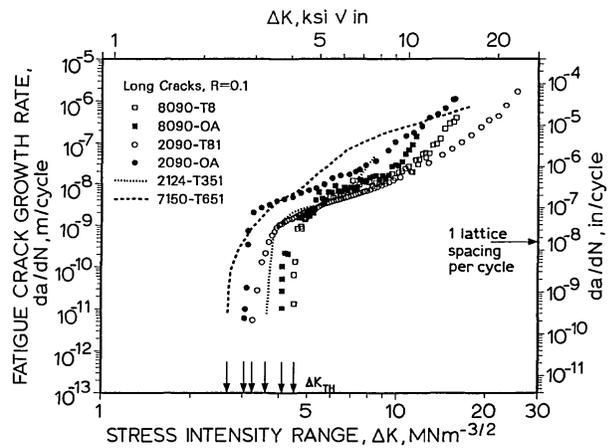
The use of aluminium alloys at moderately elevated temperatures (350–450 K) in advanced aerospace applications can be limited by severe overaging which degrades strength, toughness, and fatigue resistance. Accordingly, there has been a considerable motivation to develop higher temperature alloys, such as the Al-Fe-V-Si, Al-Fe-Ce, and Al-Fe-Co systems, to replace the more expensive titanium alloys for aerospace applications above 400 K. Naturally, there has been much interest in using ultra-low density Al-Li alloys for such applications as well.

Current results on the high temperature performance of Al-Li alloys suggest that despite a loss in properties from overaging, behaviour is generally superior to traditional aluminium alloys.^{98,101,169,175} Specifically, the crack initiation resistance of 2020-T651, 8090, and IN-905XL alloys has been found to be relatively stable up to temperatures of ~450 K; mean values of the HCF strength for 2020 and 905-XL are above the scatter band for 7075-T6.^{101,169} This improved thermal stability is presumed to result from slower coarsening kinetics of strengthening precipitates and dispersoids in these alloys.¹⁰¹ However, stretching before aging can diminish the high temperature fatigue resistance; for example, the endurance strength of 8090 has been found to decrease at higher temperatures because of dynamic recovery in the cold worked (stretched) material.⁹⁸ In addition, localised oxidation and grain boundary embrittlement at elevated temperatures can contribute to reductions in LCF resistance, particularly at low strain amplitudes.⁸⁶

Following prolonged exposures to high temperatures of up to 1000 h at 436 K, peak aged 2090 and 8090 Al-Li alloys have been shown to suffer only marginal reductions (~15%) in strength and toughness,^{175–177} even though such temperature excursions lead to grain boundary precipitation, an increase in the width of δ' -PFZs, and coarsening of matrix precipitates. However, 1000 h exposures at 533 K can lead to drastic reductions in strength, similar to behaviour in 2124-T3, 7075-T6, and 2219-T8 aluminium alloys.^{176,177} Fatigue crack growth resistance in Al-Li alloys can also be degraded by high temperature exposure, as crack paths become more linear leading to diminished crack tip shielding and enhanced crack growth rates.^{175,176} The increase in growth rates, however, is confined to high ΔK levels (where crack velocities exceed $\sim 10^{-9}$ m/cycle); near-threshold behaviour is generally unchanged. However, the fatigue properties of thermally exposed Al-Li alloys are still comparable with traditional 2124-T351, 7150-T6 and 7150-T7 alloys (Fig. 32).

Environmental effects

Environments such as moist air, water vapour, and aqueous halide solutions are known to accelerate crack propagation rates in most 2000 and 7000 series aluminium alloys under both monotonic and cyclic loading.^{178–200} For example, high strength aluminium alloys, including Al-Li alloys, are susceptible to intergranular stress corrosion cracking (SCC) under mono-



32 Influence of high temperature exposure of 1000 h at 436 K (overaging) on fatigue crack propagation rates in 2090-T81, compared with traditional alloys. Although overaging degrades crack growth resistance of Al-Li alloy, it still remains comparable, if not superior, with traditional alloys (Refs. 175, 176)

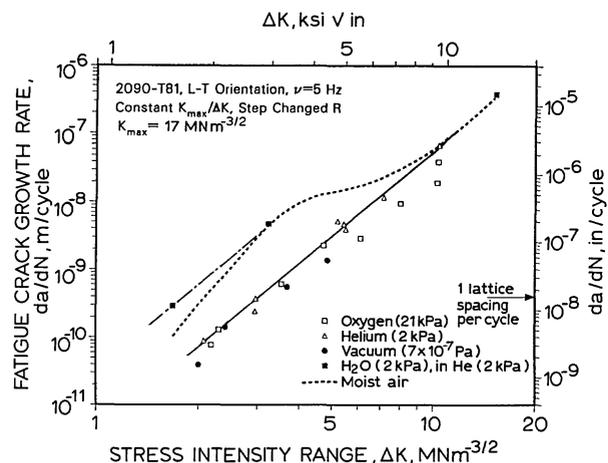
tonic loading, particularly in the short-transverse (S-T, S-L) orientations.^{188–192} Under cyclic loads, environmentally assisted crack growth may be observed at stress intensities either above or below the sustained-load threshold for environmentally assisted cracking (K_{Isc}). In the latter case, environmental damage mechanisms are associated with the synergistic effects of plastic deformation and chemical/electrochemical reactions.^{133–135, 178–187}

Because of the highly reactive nature of lithium, Al-Li alloys are often incorrectly perceived to be more sensitive to corrosive environments than traditional aluminium alloys. However, in many respects Al-Li-Cu alloys behave like 2000 series alloys, and are superior to 7000 series alloys, in moist air, water vapour, and aqueous chloride environments.^{133–137, 187}

Behaviour in gaseous environments

As with 2000 and 7000 series alloys, water vapour accelerates fatigue crack propagation rates in Al-Li alloys,^{133–135,179,180} as shown for 2090-T81 in Fig. 33 (crack closure effects were minimised in these data by measuring growth rates using constant K_{max} /variable R techniques or with physically short (<1 mm) cracks.¹³⁵ At equivalent ΔK levels, crack growth rates in pure water vapour are comparable with those in moist air, but are accelerated relative to behaviour in pure helium, oxygen, and vacuum. Crack velocities are independent of water vapour pressure for p_{H_2O} above ~ 0.2 Pa (at a frequency of 5 Hz); below this 'saturation pressure', growth rates decrease with decreasing partial pressure to approach inert environment crack velocities. Environmental effects are most pronounced at near-threshold levels and diminish with increasing ΔK , showing a plateau behaviour coincident with the transition; in both regimes, growth rates display a ΔK^4 power law relationship.¹³³

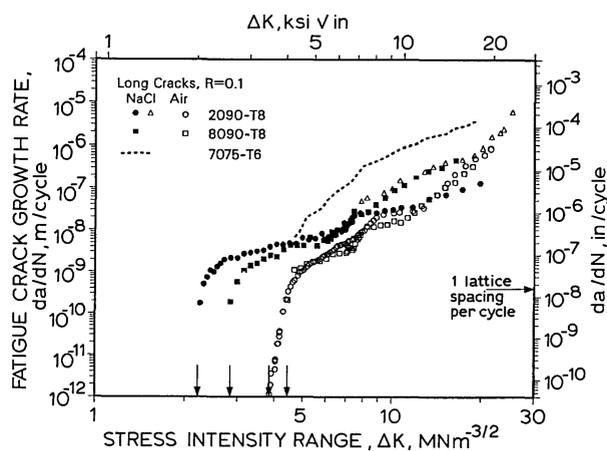
In common with other high strength aluminium alloys, hydrogen embrittlement has been suggested as



33 Influence of environment on intrinsic fatigue crack growth rates in 2090–T81 alloy (L–T orientation, 5 Hz), obtained from short through-thickness cracks or long cracks under constant K_{max} /variable R conditions. Note that crack velocities are significantly enhanced only by water vapour, particularly close to ΔK_{TH} ; behaviour in O_2 is comparable with that in inert environments (Ref. 133)

the prime mechanism of crack tip damage involved in the accelerated cracking of Al–Li alloys in moist air relative to vacuum.^{133,179–181,186} Mechanistically, this occurs in gaseous and aqueous environments which produce atomic hydrogen via reaction with fresh metal surface exposed at the crack tip by the cyclic deformation. Hydrogen is then adsorbed and transported into the plastic zone ahead of the crack tip by diffusion or moving dislocations,¹⁷⁹ leading to faster crack growth rates; cracking may be aided further by reduced slip reversibility in the presence of an aggressive environment.³⁵ Because nearly identical crack growth rates are observed in oxygen and inert environments (Fig. 33), neither oxygen nor oxide film formation on crack surfaces is considered to influence the morphology and reversibility of plastic deformation.¹³³ Moreover, the measured dependence of growth rates on H_2O partial pressure and dependence of saturation pressure on load ratio and frequency are consistent with an impeded molecular transport model of hydrogen embrittlement.^{133,180}

Fractographically, the embrittlement is reflected in differences in the crack path morphology. In inert environments, fatigue cracking in Al–Li alloys proceeds along intersecting $\{111\}$ slip planes, particularly at moderate to high ΔK levels exceeding $\sim 6 \text{ MN m}^{-3/2}$ (Refs. 135, 186), resulting in large crystallographic facets and unusually rough fracture surfaces (Fig. 18). Conversely, a flat transgranular cracking morphology (resembling cleavage) along $\{100\}$ or $\{110\}$ planes, combined with regions of faceted slip band cracking within subgrains, is seen at near-threshold levels. Hydrogen environments, on the other hand, lead to increased intersubboundary cracking at high ΔK s,^{133–135} and promote the cleavage like cracking at progressively higher ΔK levels,^{132–134} both factors that enhance crack growth rates; no evidence for ductile or brittle straitions has been reported in Al–Li alloys.

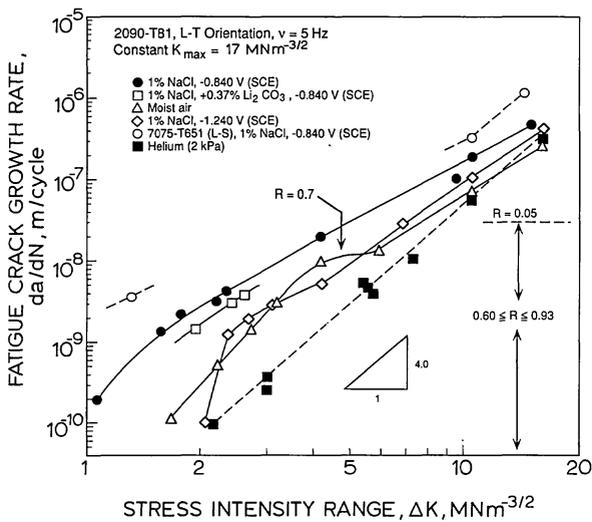


34 Comparison of fatigue crack propagation rates in 2090–T81 and 8090–T8 Al–Li alloys (L–T orientation, $R=0.1$, 50 Hz) compared with traditional 7075–T6 Al alloy in aqueous 3.5% NaCl environments (Refs. 186,193)

Apart from the intrinsic damage mechanisms described above, many extrinsic factors can affect environmentally assisted cracking in Al–Li alloys. For example, *in vacuo* testing can result in more tortuous crack paths with extensive faceting compared with laboratory air, possibly because of the greater ease of reversible slip in inert environments.³⁵ Additionally, the presence of corrosion products on crack surfaces can promote (oxide induced) crack closure and retard crack advance, particularly at near-threshold stress intensity levels.^{111,119} However, limited studies¹¹⁵ show that the role of such oxidation debris in commercial Al–Li alloys is generally minimal (at least in moist air environments), as experimentally measured oxide thicknesses are considerably smaller than the relevant crack tip opening displacements.¹¹⁵ Results of Al–Li–Cu alloys in gaseous oxygen and saline environments, on the other hand, suggest oxide induced closure effects may be important.¹³⁵

Behaviour in aqueous chloride environments

Aluminium–lithium alloys are particularly susceptible to accelerated fatigue crack propagation in aqueous chloride solutions.^{133,178–187} Typical growth rate data for commercial 2090 and 8090 Al–Li alloys in 3.5% aqueous NaCl (measured using conventional load shedding schemes at $R=0.1$) are illustrated in Fig. 34. Crack velocities in 2090–T81 are typically one to three orders of magnitude faster in salt water compared with moist air at low stress intensities, with a $\sim 50\%$ reduction in fatigue threshold. The acceleration in growth rates has been proposed to result from a diminished contribution from roughness induced closure, arising from the combined action of corrosive attack and mechanical fretting in flattening fracture surface asperities.¹⁸⁶ Alternatively, the reduced faceting and closure may result from an intrinsic change in cracking mechanism in chloride environments.¹³⁵ However, at equivalent ΔK levels, growth rates in Al–Li alloys in sea water environments are still an order of magnitude slower than in 7075–T6 (Fig. 34). Studies on the role of microstructure in influencing corrosion fatigue of Al–Li



35 Intrinsic corrosion fatigue crack growth rates for 2090-T81 and 7075-T6 alloys (L-T orientation, 5 Hz) in aqueous 1% NaCl under impressed anodic (-0.84 V) and cathodic (-1.24 V) potentials, compared with helium and moist air. Anodic polarisation produces the largest corrosion fatigue effect, which is reduced by Li_2CO_3 additions (Refs. 133-135)

alloys in chloride solutions are scarce; limited results suggest that overaging does not significantly lower fatigue resistance as crack paths remain predominantly transgranular.¹⁹³

More detailed studies¹³³⁻¹³⁵ on the intrinsic fatigue crack growth kinetics show that aqueous NaCl environments accelerate crack propagation in Al-Li alloys under both anodic and cathodic polarisation (Fig. 35). The magnitude of the effect is a function of frequency and Li_2CO_3 additions, and is most pronounced at near-threshold levels under anodic polarisation; in 2090-T81 for example, growth rates are almost two orders of magnitude faster than in inert or water vapour containing environments. In contrast, cathodic potentials reduce fatigue crack growth rates in 2090 to levels approaching those observed in moist air, as with other aluminium alloys.^{178,182} In addition, passivating ions (0.4% Li_2CO_3) can also retard crack growth rates in NaCl (Fig. 35) by forming a protective film on the crack surface.^{133,187}

Mechanistically, the intrinsic environmental damage of Al-Li alloys in aqueous NaCl environments is associated with combined effects of hydrogen and surface films.^{133,187} Cathodic polarisation studies are not conclusive in ruling out hydrogen embrittlement mechanisms because interpretations are complicated by lack of knowledge regarding the crack tip solution chemistry.^{133, 179} The observed effect of frequency on crack growth rates, however, does strongly suggest an active role of surface films. In most 7000 series aluminium alloys (at ΔK levels of $\sim 5-8 \text{ MN m}^{-3/2}$), decreasing frequency results in increasing growth rates;¹⁹⁴⁻¹⁹⁹ hydrogen embrittlement models generally relate this to enhanced surface reactions or mass transport during the longer time period of the cycle. Fatigue crack propagation rates in 2090-T81, conversely, decrease by a factor of two as the frequency is reduced from 5 to 0.1 Hz at ΔK levels $\sim 10 \text{ MN}$

$\text{m}^{-3/2}$ or above, though little effect of frequency is apparent close to ΔK_{TH} .^{133,193} Based on these observations, it has been suggested that mass transport is not the rate limiting step in Al-Li alloys; rather, increased crack tip strain rates at higher frequencies are reasoned to destabilise (protective) surface films, resulting in enhanced anodic dissolution or crack advance by film rupture. This effect may be coupled with time based hydrogen production and entry into the metal.^{133,179}

Fracture surfaces in NaCl and moist air environments are nominally similar though, because of the more aggressive nature of salt water, a greater degree of intersubgrain boundary fracture and reduced slip band cracking is seen at high ΔK levels.¹³⁴⁻¹³⁶ At low ΔK levels, on the other hand, the crystallographic fracture morphology resembling transgranular cleavage is more prevalent in NaCl than in moist air environments. The associated reduction in crack path tortuosity caused by an environmentally induced change from faceted slip band cracking in benign environments to relatively linear intersubgranular and $\{100\}$ or $\{110\}$ type cracking may be the prime cause of the lower closure levels measured in conventional corrosion fatigue tests.¹⁸⁶

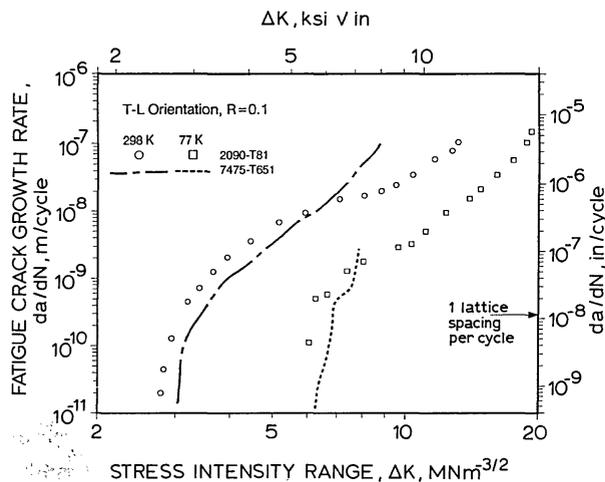
Unlike behaviour reported for steels^{179,198,199} where local crack tip electrochemistry effects greatly enhance environmental effects when crack sizes are small ($\sim 150 \mu\text{m}$ to 5 mm), the corrosion fatigue behaviour of short cracks ($\sim 300 \mu\text{m}$ to 4 mm) in Al-Li alloys is essentially identical to that of long ($> 5 \text{ mm}$) cracks.^{133,135} In fact in 2090-T81 alloy, K -based similitude adequately describes the growth kinetics of fatigue cracks for all crack sizes above $300 \mu\text{m}$ in length.

In summary, it may be concluded that the intrinsic fatigue crack growth resistance of Al-Li alloys in moist air and sea water environments is comparable with 2000 series alloys and in fact superior to 7000 series alloys^{133-137,179} (Figs. 20, 35). In addition, limited studies indicate that Al-Li alloys show superior crack initiation and $S-N$ fatigue resistance in these environments, at least compared with 7075 type alloys, a fact attributed to their reduced susceptibility to hydrogen absorption and hydrogen assisted intergranular fracture.¹⁸³⁻¹⁸⁵

Cryogenic properties

The cryogenic properties of Al-Li alloys have recently been the focus of much attention following reports that indicated remarkable improvements in their strength, ductility, and fracture toughness at liquid nitrogen (77 K) and liquid helium (4 K) temperatures compared with room temperature.³⁸⁻⁴² Specifically, studies for the L and L-T orientations on 2090-T81 alloy plate showed a 45% increase in strength, 67% increase in tensile elongation, 12% increase in modulus, and most importantly 90% increase in fracture toughness between 298 K and 4 K.* These properties

* The dramatic improvements in toughness at cryogenic temperatures are not, however, typical of all Al-Li microstructures or product forms; thin sheet Al-Li alloys, such as 2090-T83 for example, may actually exhibit lower toughness with decrease in temperature.^{69,200}

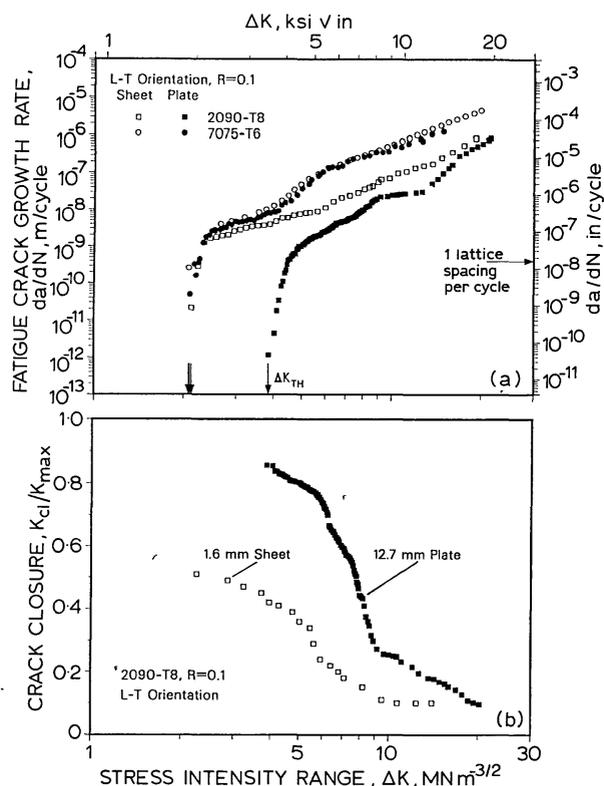


36 Fatigue crack propagation behaviour in 2090-T81 and 7475-T6 alloys at ambient and liquid nitrogen temperatures (T-L orientation, $R=0.1$, 50 Hz) (Ref. 38)

combined with lower specific modulus and superior fatigue resistance have made Al-Li alloys very attractive for many cryogenic structural applications; these include liquid hydrogen and liquid oxygen fuel tanks on future hypersonic and transatmospheric aerospace vehicles, supporting structures for magnets in particle accelerators, and high energy fusion devices.

As with many high strength aluminium alloys, the HCF resistance of Al-Li alloys is substantially improved at cryogenic temperatures.²⁰¹⁻²⁰³ For example, the smooth bar axial fatigue strength of 2090-T81 (at a life of 10^5 cycles, $R=0.1$) increases from $\sim 290 \text{ MN m}^{-2}$ at 298 K to $\sim 700 \text{ MN m}^{-2}$ at 4 K ($\sim 90\%$ of the tensile strength), behaviour which is second only to 7075-T6; in addition, the Al-Li alloy is less notch sensitive.²⁰¹ Studies on the Al-Li-Cu-Mg-Zr system²⁰³ indicate similar trends, and further show that though the fatigue strength at 298 K can be improved by increasing the Zr content, the temperature dependence of fatigue strength is compromised.

Although data are limited, fatigue crack growth rates in Al-Li alloys such as 2090 show a decrease at lower temperatures, typical of pure aluminium and many aluminium alloys.^{38,204-206} Moreover, the properties are comparable with traditional aluminium alloys at cryogenic temperatures (Fig. 36). The generally improved fatigue resistance of aluminium alloys at cryogenic temperatures has been attributed to an increase in the monotonic or cyclic yield strength, which in turn increases the plastic work required for fracture; this has been associated with more homogeneous plastic deformation,^{204,205} or a reduction in rates of thermally activated processes at low temperatures.²⁰³ Alternatively, the reduced low temperature crack velocities in 2090-T8 Al-Li alloy may be attributed to the smaller CTODs per cycle, resulting from an increase in strength, elastic modulus, and strain hardening rate, and to limited environmental effects at cryogenic temperatures. However, the explanations are not definitive because of the lack of intrinsic low temperature, crack propagation data in inert environments.



a fatigue crack propagation; *b* crack closure behaviour

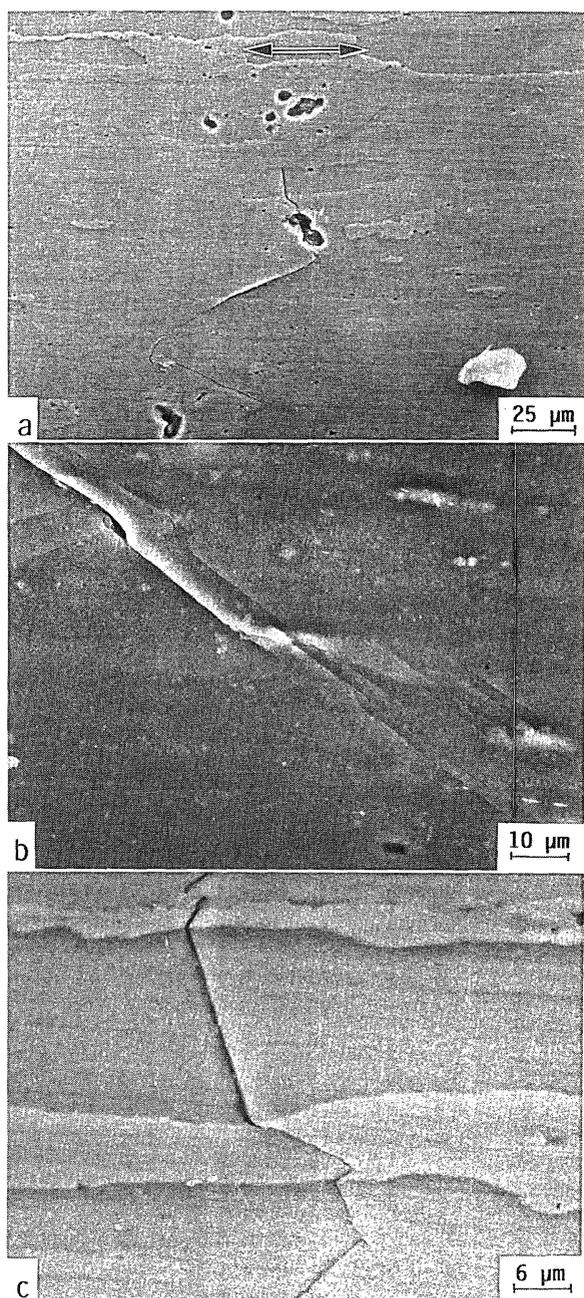
37 Fatigue behaviour in 1.6 mm thin 2090-T83 sheet and 12.7 mm thick T81 plate, as a function of ΔK (L-T orientation, $R=0.1$, 50 Hz, moist air), compared with 7075-T6. Properties in plate were measured on 6.4 mm thick samples ($t/2$ location); sheet behaviour was examined in as received thickness (Refs. 70, 207)

Influence of wrought product form

The predominant role of crack tip shielding in promoting superior fatigue crack growth resistance in Al-Li alloys via crack deflection mechanisms implies that the effect may not necessarily be seen in all product forms.^{70,207} For example, at the same ΔK , fatigue cracks in 1.6 mm thick 2090-T83 sheet propagate significantly faster than in 12.7 mm thick 2090-T81 plate, particularly as ΔK levels (at $R=0.1$) approach the fatigue threshold (Fig. 37a), though both rolled sheet and plate have nominally similar composition and hardening precipitates.²⁰⁷ Such faster crack growth rates in the sheet are concurrent with markedly lower crack closure levels compared to plate (Fig. 37b). This is in contrast to traditional aluminium alloys, such as 7075-T6, which rarely show such differences in behaviour between sheet and plate, at least for ΔK levels below $\sim 20 \text{ MN m}^{-3/2}$.*

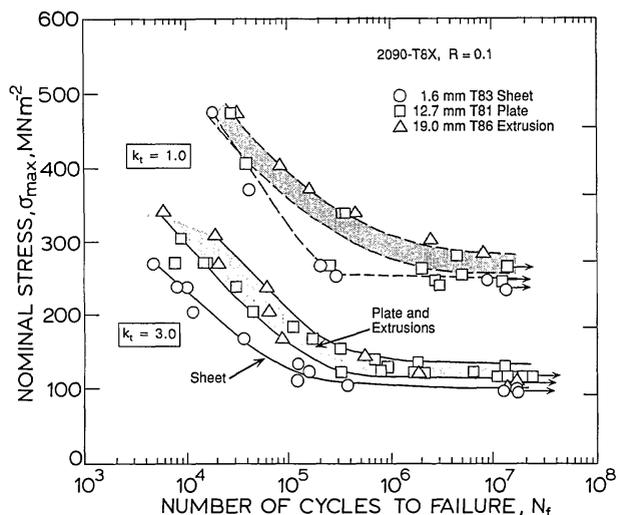
Such differences in fatigue behaviour in Al-Li alloy sheet and plate are principally associated with microstructurally induced variations in crack path deflection and resulting crack closure effects (Figs. 16-18, 37, 38). In the T81 plate material, marked slip planarity induced by coherent δ' and deformation texture

* Stress state effects may be expected to affect crack growth rates at higher ΔK^3 levels as deformation conditions in the sheet alloys approach that of plane stress.²⁰⁷



38 Comparison of crack path and fracture surface morphologies during fatigue crack growth in 2090-T83 sheet, showing crack paths: *a* along crack growth direction, *b* specimen thickness, and *c* on fracture surface; arrow indicates general direction of crack advance (Ref. 207)

promotes crack deflection through crystallographically faceted slip band cracking (Figs. 16–18). Such effects are not as prominent in sheet²⁰⁷ because recrystallisation occurs readily in thin section products, below ~4 mm in thickness, despite the presence of Zr. The finer, partially recrystallised grain structure and, reduced texture induce essentially linear crack paths and hence far smoother fracture surfaces (Fig. 38). It should be noted, however, that even without significant crack closure effects, Al–Li sheet alloys still display superior fatigue crack growth resistance than traditional alloys, particularly at



39 Influence of wrought product form on HCF resistance of 2090-T8 ($k_t=1$, or 3, $R=0.1$, longitudinal orientation) in ambient air. Note that plates and extrusions show better fatigue resistance compared with sheet (Ref. 70)

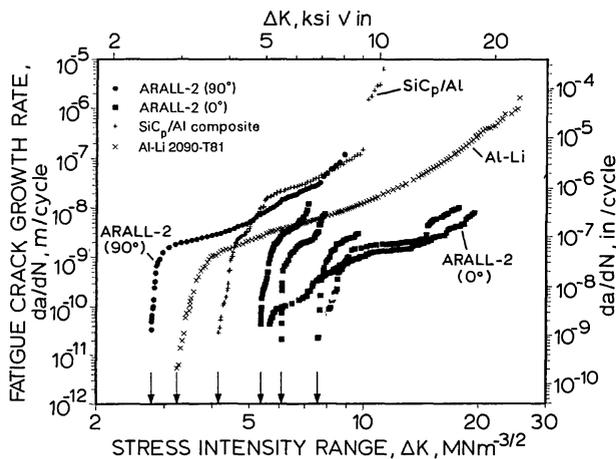
higher growth rates (Fig. 37*a*), presumably because of an increased modulus and the beneficial role of δ' in promoting planar slip deformation and crack deflection.

The influence of load ratio and specimen orientation on the crack growth behaviour of Al–Li sheet is similar to that described above for plate alloys. Higher load ratios induce faster growth rates, particularly at near-threshold levels, and the lowest crack growth resistance is found for the $L+45^\circ$ orientation.^{70,207} However, owing to the limited role of crack closure in sheet, the magnitude of the load ratio and orientation effects is less pronounced than in plate alloys.²⁰⁷ In fact, the larger CTOD levels associated with higher load ratios minimise the closure effect to such an extent that, at $R>0.75$, sheet and plate material show essentially similar behaviour.

Differences in the $S-N$ fatigue behaviour of commercial Al–Li alloys have also been noted with differing wrought product form (Fig. 39). Specifically, for the three principal product variants of 2090-T8X alloy, T86 extrusions exhibit the highest smooth specimen fatigue strength, followed by T81 plate and T83 sheet; behaviour, as in plate alloys, is anisotropic.⁷⁰ For notched tests (with a stress concentration factor k_t of 3), extrusions and plate show equivalent behaviour, with the fatigue strength of the sheet material being slightly inferior. Such results have been attributed to delamination induced reduction in lateral constraint in the thicker product forms,⁷⁰ though they are more likely to be associated with the lower strength of the 2090 sheet.

Concluding remarks

The resurgence of interest in Al–Li alloys has resulted in the development of a series of low density, monolithic aerospace materials with strength–toughness properties comparable with traditional aluminium alloys and a considerable price advantage over com-



40 Comparison of fatigue crack growth resistance of advanced structural aerospace materials, namely, Al-Li alloys (2090-T81), metal matrix composites (SiC-particulate reinforced 7091 PM aluminium alloy) and ARALL-2 laminates (From Refs. 11, 14, 115)

peting composites. Moreover, they offer the designer a material with improved fatigue crack growth properties over many composites (Fig. 40) and traditional Al alloys, provided in-plane (e.g. L-T, T-L) orientations are used. For other orientations such as S-T and S-L, delamination cracking can lead to poor toughness and crack growth properties in thick plate sections. Furthermore, short-transverse delamination failures are responsible for the low fatigue lives measured under anticlastic bending fatigue.²⁰⁸ Under these circumstances, the performance of Al-Li alloys is inferior to traditional alloys.

Since the beneficial fatigue properties can be traced primarily to extrinsic crack tip shielding mechanisms, designers must appreciate the implications of this phenomenon on life prediction. For example, many of the contrasting observations on the fatigue crack growth behaviour of Al-Li alloys can be explained as consequences of crack tip shielding, resulting from highly tortuous crack paths induced by planar slip. This leads to superior long crack properties, under constant amplitude loading at positive, low load ratios, and under tension dominated spectrum loading, all instances where crack closure due to wedging phenomena is not restricted. However, factors which limit crack path tortuosity and asperity wedging can compromise the superiority of these alloys. This can occur at high load ratios, under compression dominated spectrum loading, with microstructurally small cracks, and with refined and overaged microstructures; in these instances, the fatigue behaviour of Al-Li alloys is essentially similar to 2000 and 7000 series alloys.

A further consequence of the role of crack tip shielding in Al-Li alloys is complications with the assessment of fatigue crack growth resistance. In addition to loading spectra, crack size, and specimen orientation, crack growth behaviour tends to be strongly dependent on wrought product form and specimen geometry. For example, thin specimens generally will exhibit significantly faster crack growth rates than thick specimens of nominally identical

composition and microstructure, principally by reducing through-thickness crack path meandering. Similarly, centre cracked, rather than compact tension, bend, or double cantilever beam, specimen geometries may also limit the fatigue crack growth property advantage of Al-Li alloys by reducing off-angle cracking and inhibiting crack deflection. A similar suppression of shielding in Al-Li alloys has been reported by testing single edged notched, tension specimens with fixed grips compared with freely rotating grips.¹³³

Finally, it should be noted that primary sources for crack deflection and crack path tortuosity, which confer the excellent fatigue crack growth properties in commercial Al-Li plate alloys, are marked slip planarity, strong deformation texture, and weak short-transverse properties, all microstructural features which limit toughness and which may be considered undesirable for many service applications. Although new alloys can, and have been, developed with less anisotropy to solve this problem, their fatigue crack growth properties are likely to be inferior to those of the early commercial alloys.

These factors suggest that the use of Al-Li alloys, or the direct substitution for traditional alloys, in fatigue critical designs requires a greater mechanistic understanding and more detailed analysis of the correlation of laboratory scale coupon tests to the performance of structures in service; with Al-Li alloys, it is by no means certain that the superior fatigue properties measured in laboratory tests will always be realised in actual structures.

Acknowledgments

This work was supported by the Director, Office of Energy Research, Office of Basic Energy Sciences, Materials Sciences Division of the US Department of Energy under Contract No. DE-AC03-76SF00098. The authors would particularly like to thank Dr R. J. Bucci of Alcoa for his help and encouragement throughout this work. In addition, they thank Drs P. E. Bretz and R. R. Sawtell, also from Alcoa, P. O. Wakeling, formerly of Alcan; Drs G. R. Chani, G. V. Scarich, and K. M. Bresnahan of Northrop; Dr W. E. Quist of Boeing; G. J. Petrak of WRDC; Dr S. Das of Allied Signal Inc.; Prof. R. P. Gangloff of the University of Virginia; Prof. N. J. Kim of Pohang Institute of Technology (Korea); Dr K. V. Jata of the University of Dayton Research Institute; Dr R. S. Piascik of NASA Langley; and Prof. P. P. Pizzo of San José State University, for their collaboration. The assistance of the late Dr W. Yu, H. F. Hayashigatani, J. E. Miles, D. Kovar, and J. C. McNulty with the experiments, Ms Madeleine Penton with preparation of the manuscript, and the key reader and referees for their critical comments, is also gratefully acknowledged.

References

1. F. H. FROES: *Mater. Des.*, 1989, **10**, (3), 111.
2. M. A. STEINBURG: *Sci. Am.*, 1986, **255**, (4), 66.
3. D. R. TENNEY, B. LISAGOR, and S. DIXON: *J. Aircraft*, 1989, **26**, 953.
4. P. CANNON: *J. Met.*, 1989, **40**, (5), 10.

5. J. J. DE LUCCIA, R. E. TRABOCCO, J. WALDMAN, and J. F. COLLINS: *Adv. Mater. Proc.*, 1989, **137**, (5), 39.
6. T. M. F. RONALD: *Adv. Mater. Proc.*, 1989, **137**, (5), 29.
7. W. E. QUIST, A. L. WINGERT, and G. H. NARAYANAN: in 'Aluminum–lithium: Development, application and super-plastic forming', (ed. S. P. Agrawal and R. J. Kar), 227–279; 1986, Metals Park, OH, ASM International.
8. T. W. CHOU, R. L. McCULLOUGH, and R. B. PIPES: *Sci. Am.*, 1986, **225**, (4), 192.
9. S. V. NAIR, J. K. TIEN, and R. C. BATES: *Int. Met. Rev.*, 1985, **30**, 275.
10. A. L. GEIGER and M. JACKSON: *Adv. Mater. Proc.*, 1989, **137**, (7), 23.
11. J.-K. SHANG, W. YU, and R. O. RITCHIE: *Mater. Sci. Eng.*, 1988, **102A**, 181.
12. S. V. KAMAT, J. P. HIRTH, and R. MEHRABIAN: *Acta Metall.*, 1989, **37**, 2395.
13. R. J. BUCCI, L. N. MUELLER, L. B. VOGELSANG and J. W. GUNNICK: 'Aluminum alloys: Contemporary research and applications', (ed. A. K. Vasudévan and R. D. Doherty), Treatise on Materials Science and Technology, Vol. 31, 295–322; 1990, New York, Academic Press.
14. R. O. RITCHIE, W. YU, and R. J. BUCCI: *Eng. Fract. Mech.*, 1989, **32**, 361.
15. R. MARISSSEN: PhD thesis, Delft University of Technology, 1988.
- *16. T. H. SANDERS, Jr and E. A. STARKE, Jr (eds): 'Aluminum–lithium alloys'; 1981, Warrendale, PA, The Metallurgical Society of AIME.
- *17. T. H. SANDERS, Jr and E. A. STARKE, Jr (eds.): 'Aluminum–lithium alloys II'; 1984, Warrendale, PA, The Metallurgical Society of AIME.
- *18. C. BAKER, P. J. GREGSON, S. J. HARRIS, and C. J. PEEL (eds.): 'Aluminium–lithium alloys III'; 1986, London, the Institute of Metals.
- *19. C. CHAMPIER, B. DUBOST, D. MIANNAY, and L. SABETY (eds.): 'Proc. 4th int. conf. on aluminium–lithium alloys', *J. Phys.*, **C3**, (9), 1987, Paris, Les Editions de Physique.
- *20. T. H. SANDERS, Jr and E. A. STARKE, Jr (eds.): in 'Proc. 5th int. conf. on aluminium–lithium alloys'; 1989, Edgbaston, UK, Materials and Component Engineering Publications.
21. E. A. STARKE, Jr, T. H. SANDERS, Jr, and I. G. PALMER: *J. Met.*, 1981, **33**, (8), 24.
22. R. GRIMES, A. J. CORNISH, W. S. MILLER, and M. A. REYNOLDS: *Met. Mater.*, 1988, **4**, (7), 357.
23. C. A. STUBBINGTON: *Met. Mater.*, 1988, **4**, (7), 424.
24. T. H. SANDERS, Jr, and E. S. BALMUTH: *Met. Prog.*, 1978, **121**, (3), 32.
25. R. C. DORWARD and T. R. PRITCHETT: *Mater. Des.*, 1988, **9**, (2), 63.
26. R. S. JAMES: in 'Metals handbook', 10 edn, Vol. 2, 178–199; 1990, Materials Park, OH, ASM International.
27. W. E. QUIST and G. H. NARAYANAN: in 'Aluminum alloys: Contemporary research and applications', (ed. A. K. Vasudévan and R. D. Doherty), Treatise on Materials Science and Technology, Vol. 31, 219–254; 1990, New York, Academic Press.
28. C. J. PEEL: *Mater. Sci. Technol.*, 1986, **2**, 1169.
29. E. S. BALMUTH and R. SCHMIDT: in Ref. 16, p. 69.
30. W. E. QUIST, G. H. NARAYANAN, and A. L. WINGERT: in Ref. 16, p. 313.
31. J. C. EKVAL, J. E. RHODES, and G. G. WALD: In 'Design of fatigue and fracture resistant structures', STP 761, 328–341; 1982, Philadelphia, PA, American Society for Testing and Materials.
32. R. E. LEWIS, D. WEBSTER, and I. G. PALMER: 'A feasibility study for development of structural aluminum alloys from rapidly solidified powders for aerospace structural applications', Final Report No. AFML–TR–78–102, Lockheed Palo Alto Research Laboratory, Palo Alto, CA, 1978.
33. B. NOBLE, S. J. HARRIS, and K. DINSDALE: *J. Mater. Sci.*, 1982, **17**, 461.
- *34. T. H. SANDERS, Jr and E. A. STARKE, Jr: *Acta Metall.* 1982, **30**, 927.
- *35. K. V. JATA and E. A. STARKE, Jr: *Metall. Trans.*, 1986, **17A**, 1011.
36. W. S. MILLER, M. P. THOMAS, D. J. LLOYD, and D. CREBER: *Mater. Sci. Technol.*, 1986, **2**, 1210.
37. D. DEW-HUGHES, E. CREED, and W. S. MILLER: *Mater. Sci. Technol.*, 1988, **4**, 106.
38. J. GLAZER, S. L. VERZASCONI, E. N. C. DALDER, W. YU, R. A. EMIGH, R. O. RITCHIE, and J. W. MORRIS: *Adv. Cryogen. Eng.*, 1986, **32**, 397.
39. D. WEBSTER: *Met. Prog.*, 1984, **125**, (4), 33.
40. R. C. DORWARD: *Scr. Metall.*, 1986, **20**, 1379.
41. K. T. VENKATESWARA RAO, W. YU, and R. O. RITCHIE: *Metall. Trans.*, 1989, **20A**, 485.
42. K. V. JATA and E. A. STARKE, Jr: *Scr. Metall.*, 1988, **22**, 1553.
43. S. SURESH, A. K. VASUDÉVAN, M. TOSTEN, and P. R. HOWELL: *Acta Metall.*, 1987, **35**, 25.
- *44. A. K. VASUDÉVAN, R. D. DOHERTY, and S. SURESH: in 'Aluminum alloys: Contemporary research and applications', (ed. A. K. Vasudévan and R. D. Doherty), Treatise on Materials Science and Technology, Vol. 31, 445–462; 1990, New York, Academic Press.
45. R. O. RITCHIE: *Mater. Sci. Eng.*, 1988, **A103**, 15.
46. A. G. EVANS: in 'Fracture mechanics: Perspectives and directions (20th symp.)', (ed. R. P. Wei and R. P. Gangloff), STP 1020, 267–291; 1989, Philadelphia, PA, American Society for Testing and Materials.
47. H. M. FLOWER and P. J. GREGSON: *Mater. Sci. Technol.*, 1987, **3**, 81.
48. S. F. BAUMANN and D. B. WILLIAMS: *Metall. Trans.*, 1985, **16A**, 1203.
49. T. H. SANDERS, Jr, and E. A. STARKE, Jr: in Ref. 20, p. 1.
50. D. B. WILLIAMS: in Ref. 20, p. 551.
51. M. H. TOSTEN, A. K. VASUDÉVAN, and P. R. HOWELL: *Metall. Trans.*, 1988, **19A**, 51.
52. R. J. RIOJA, P. E. BRETZ, R. R. SAWTELL, W. H. HUNT, and E. A. LUDWICZAK: in 'Aluminium alloys: Their physical and mechanical properties', (ed. T. H. Sanders, Jr and E. A. Starke, Jr), 1781–1797; 1986, Warley, UK, Engineering Materials Advisory Services.
53. J. C. HUANG and A. J. ARDELL: in Ref. 18, p. 455.
54. D. B. WILLIAMS and P. R. HOWELL: in 'Aluminum alloys: Contemporary research and applications', (ed. A. K. Vasudévan and R. D. Doherty), Treatise on Materials Science and Technology, Vol. 31, 365–388; 1990, New York, Academic Press.
55. J. M. PAPA ZIAN and P. S. GILMAN: *Mater. Sci. Eng.*, 1990, **A125**, 121.
56. F. W. GAYLE and B. VANDERSANDE: *Acta Metall.*, 1989, **37**, 1033.
- *57. J. DHERS, J. DRIVER, and A. FOURDEUX: in Ref. 18, p. 233.
- *58. K. T. VENKATESWARA RAO and R. O. RITCHIE: *Metall. Trans.*, 1991, **22A**, 191.
59. K. T. VENKATESWARA RAO and R. O. RITCHIE: *Mater. Sci. Technol.*, 1989, **5**, 882.
60. A. K. VASUDÉVAN, W. G. FRICKE, R. C. MALCOLM, R. J. BUCCI, M. A. PRZYSTUPA, and F. BARLAT: *Metall. Trans.*, 1988, **19A**, 731.
61. J. HIRSCH, O. ENGLER, K. LUCKE, M. PETERS, and K. WELPMANN: in Ref. 19, p. 605.
62. G. R. YODER, P. S. PAO, M. A. IMAM, and L. A. COOLEY: *Scr. Metall.*, 1988, **22**, 1241.
63. G. R. YODER, P. S. PAO, M. A. IMAM, and L. A. COOLEY: *Scr. Metall.*, 1989, **23**, 1455.
64. C. J. WARREN and R. J. RIOJA: in Ref. 20, p. 417.
65. M. PETERS, J. ESCHWEILER, and K. WELPMANN: *Scr. Metall.*, 1986, **20**, 259.
66. A. K. VASUDÉVAN, M. A. PRZYSTUPA, and W. G. FRICKE: *Scr. Metall.*, 1990, **24**, 1429.
67. S. J. HARRIS, B. NOBLE, and A. DODD: in Ref. 20, p. 1061.
68. C. P. BLANKENSHIP and E. A. STARKE, Jr: *Fat. Fract. Eng. Mater. Struct.*, 1991, **14**, 103.
69. K. T. VENKATESWARA RAO and R. O. RITCHIE: *Acta Metall. Mater.*, 1990, **38**, 2309.
- *70. R. J. BUCCI, R. C. MALCOLM, E. L. COLVIN, S. MURTHA, and R. S. JAMES: 'Cooperative test program for the evaluation of engineering properties of Al–Li alloy 2090–T8X sheet, plate and extrusion products', Report NSWC TR 89–106, Alcoa Technical Center, Pittsburgh, PA, 1989.
71. P. MEYER, Y. CANS, D. FERTON and M. REBOUL: in Ref. 19, p. 131.
72. M. A. REYNOLDS and E. CREED: in Ref. 19, p. 195.

* Important references on fatigue of Al–Li alloys.

73. W. S. MILLER, J. WHITE, M. A. REYNOLDS, D. S. McDARMAID, and G. M. STARR: in Ref. 17, p. 151.
74. A. F. SMITH: in Ref. 20, p. 1587.
75. N. J. KIM, D. RAYBOULD, R. L. BYE, and S. K. DAS: in Ref. 20, p. 123.
76. M. R. MITCHELL: in 'Fatigue and microstructure', 385-437; 1979, Metals Park, OH, American Society for Metals.
77. E. J. COYNE, T. H. SANDERS, Jr, and E. A. STARKE, Jr: in Ref. 16, p. 293.
78. Y. XIAO and P. BOMPARD: in Ref. 17, p. 737.
- *79. N. ESWARA PRASAD, G. MALAKONDAIAH, K. N. RAJU and P. RAMA RAO: in 'Advances in fracture research', Proc. 7th Int. Conf. of Fracture, (ed. K. Salama et al.), 1103-1112; 1989, London, Pergamon Press.
80. T. S. SRIVATSAN, K. YAMAGUCHI, and E.A. STARKE, Jr: *Mater. Sci. Eng.*, 1986, **83**, 87.
81. T. S. SRIVATSAN and E. J. COYNE: *Mater. Sci. Technol.*, 1989, **5**, 548.
82. Y. BRECHET, F. LOUCHET, and J. L. VERGER-GAUGRY: in Ref. 19, p. 709.
- *83. Y. BRECHET, F. LOUCHET, and T. MAGNIN: *J. Mater. Sci.*, 1990, **25**, 3053.
84. S. HAN, A. TOURABI, and B. WACK: in Ref. 19, p. 701.
85. J. SCHNEIDER, H.-J. GUDLADT, and V. GEROLD: in Ref. 19, p. 745.
86. T. S. SRIVATSAN and E. J. COYNE, Jr: *Int. J. Fatigue*, 1988, **10**, 91.
87. D. KHIREDDINE, R. RAHOUADI, and M. CLAVEL: *Scr. Metall.*, 1988, **22**, 167.
- *88. D. KHIREDDINE, R. RAHOUADI, and M. CLAVEL: *Acta Metall.*, 1989, **37**, 191.
- *89. W. X. FENG, F. S. LIN, and E. A. STARKE, Jr: *Metall Trans.*, 1984, **15A** 1209.
- *90. R. T. CHEN and E. A. STARKE, Jr: *Mater. Sci. Eng.*, 1984, **67**, 229.
91. B. VELTEN, A. K. VASUDÉVAN, and E. HORNOGEN: *Z. Metallkd.*, 1989, **80**, 21.
- *92. J. M. GENTZBITTEL and F. FOUGERES: *Scr. Metall.*, 1987, **21**, 1411.
- *93. Y. BRECHET, F. LOUCHET, C. MARCHIONNI, and J. L. VERGER-GAUGRY: *Philos. Mag.*, 1987, **56**, 353.
94. J. LENĐVAI, H.-J. GUDLADT, and V. GEROLD: *Scr. Metall.*, 1988, **22**, 1755.
- *95. Z. DI, S. SAJI, and S. HORI: in Ref. 19, p. 753.
96. R. E. LEE and W. J. D. JONES: *J. Mater. Sci.*, 1974, **9**, 476.
97. S. H. CHEN: MS thesis, China, National Tsing Hua University, 1985.
- *98. P. J. E. BISCHLER and J. W. MARTIN: in Ref. 19, p. 761.
99. L. FARCY, C. CARRE, M. CLAVEL, Y. BARBAUX, and D. ALIAGA: in Ref. 19, p. 769.
- *100. M. PETERS, K. WELPMANN, W. ZINK, and T. H. SANDERS: in Ref. 18, p. 239.
- *101. P. E. BRETZ, L. N. MUELLER, and A. K. VASUDÉVAN: in Ref. 17, p. 543.
102. P. J. GREGSON, J. NEWMAN, and A. GRAY: *Mater. Sci. Technol.*, 1989, **5**, 65.
103. Y. MUTOH, G. H. FAIR, B. NOBLE, and R. B. WATERHOUSE: *Fat. Fract. Eng. Mater. Struct.*, 1987, **10**, 261.
104. V. J. BOLAM, P. J. GREGSON, and A. GRAY: in Ref. 20, p. 1097.
105. R. W. HERTZBERG: 'Deformation and fracture mechanics of engineering materials'. 3rd edn; 1989, New York, Wiley.
106. R. O. RITCHIE: *Int. Met. Rev.*, 1979, **24**, 205.
107. S. SURESH: 'Fatigue of materials'; 1991, New York, Cambridge University Press.
108. W. ELBER: in 'Damage tolerance in aircraft structures', STP 486, 230-242; 1971, Philadelphia, PA, American Society for Testing and Materials.
109. N. WALKER and C. J. BEEVERS: *Fat. Eng. Mater. Struct.*, 1979, **1**, 135.
110. K. MINAKAWA and A. J. McEVILY: *Scr. Metall.*, 1981, **15**, 633.
111. S. SURESH, G. F. ZAMISKI, and R. O. RITCHIE: *Metall. Trans.*, 1981, **12A**, 1435.
112. R. O. RITCHIE and S. SURESH: *Metall. Trans.*, 1982, **13A**, 937.
113. J. C. NEWMAN and W. ELBER (eds.): 'Mechanics of fatigue crack closure', STP 982; 1988, Philadelphia, PA, American Society for Testing and Materials.
114. J. E. ALLISON: 'Fracture mechanics: Eighteenth symp.', STP 945, 913-933; 1988, Philadelphia, PA, American Society for Testing and Materials.
- *115. K. T. VENKATESWARA RAO, W. YU, and R. O. RITCHIE: *Metall. Trans.*, 1988, **19A**, 549.
116. W. A. HERMAN, R. W. HERTZBERG, and R. JACCARD: *Fat. Eng. Mater. Struct.*, 1988, **10**, 303.
- *117. S. J. HARRIS, B. NOBLE, and K. DINSDALE: in Ref. 17, p. 219.
118. P. S. PAO, K. K. SANKARAN, and J. E. NEAL: in Ref. 16, p. 307.
- *119. A. K. VASUDÉVAN, P. E. BRETZ, A. C. MILLER, and S. SURESH: *Mater. Sci. Eng.*, 1984, **64**, 113.
120. S. SURESH, A. K. VASUDÉVAN, and P. E. BRETZ: *Metall. Trans.*, 1984, **15A**, 369.
121. E. ZAIKEN and R. O. RITCHIE: *Mater. Sci. Eng.*, 1985, **70**, 151.
122. R. D. CARTER, E. W. LEE, E. A. STARKE, Jr, and C. J. BEEVERS: *Metall. Trans.*, 1984, **15A**, 555.
- *123. K. T. VENKATESWARA RAO and R. O. RITCHIE: *Mater. Sci. Technol.*, 1989, **5**, 896.
- *124. J. PETIT, S. SURESH, A. K. VASUDÉVAN, and R. C. MALCOLM: in Ref. 18, p. 257.
- *125. A. K. VASUDÉVAN and S. SURESH: *Metall. Trans.*, 1985, **16A**, 475.
126. C. Q. CHEN and H. X. LI: in Ref. 20, p. 973.
127. R. TINTILLIER, H. J. GUDLADT, V. GEROLD, and J. PETIT: in Ref. 20, p. 1135.
128. R. TINTILLIER, H. S. YANG, N. RANGANATHAN, and J. PETIT: in Ref. 19, P. 777.
129. M. PETERS, V. BACHMANN, and K. WELPMANN: in Ref. 19, p. 785.
130. D. S. McDARMAID and C. J. PEEL: in Ref. 20, p. 993.
131. H. D. PEACOCK and J. W. MARTIN: in Ref. 20, p. 1013.
132. J. M. NEWMAN, P. J. GREGSON, P. D. PITCHER, and P. J. E. FORSYTH: in Ref. 20, p. 1043.
- *133. R. S. PIASCIK and R. P. GANGLOFF: *Metall. Trans.*, 1991, **22A**, 2415.
- *134. R. S. PIASCIK and R. P. GANGLOFF: *Metall. Trans.* 1992, **23A**, in press.
135. R. S. PIASCIK: PhD thesis, University of Virginia, 1989.
136. R. P. GANGLOFF, R. S. PIASCIK, D. L. DICUS, and J. C. NEWMAN: in 'Proc. 17th ICAS Cong.', London, 1990, Royal Aeronautical Society.
- *137. K. T. VENKATESWARA RAO, R. S. PIASCIK, R. P. GANGLOFF, and R. O. RITCHIE: in Ref. 20, p. 955.
- *138. W. YU, and R. O. RITCHIE: *J. Eng. Mater. Technol.*, 1987, **109**, 81.
139. S. SURESH: *Metall. Trans.*, 1983, **14A**, 2375.
140. F. L. HADDLETON, S. MURPHY, and T. J. GRIFFIN: in Ref. 19, p. 809.
141. R. O. RITCHIE and J. LANKFORD (eds): 'Small fatigue cracks'; 1986, Warrendale, PA, The Metallurgical Society of AIME.
142. K. J. MILLER and E. R. DE LOS RIOS (eds): 'The behaviour of short fatigue cracks'; 1986, London Institution of Mechanical Engineers.
143. S. SURESH and R. O. RITCHIE: *Int. Met. Rev.*, 1984, **29**, 445.
- *144. K. T. VENKATESWARA RAO, W. YU, and R. O. RITCHIE: *Metall. Trans.*, 1988, **19A**, 563.
- *145. M. R. JAMES: *Scr. Metall.*, 1987, **21**, 783.
146. Z. CHEN, A. ZEGHLOUL, and J. PETIT: *Scr. Metall.*, 1989, **23**, 1005.
- *147. X. SU, J. M. LARSEN, and J. R. JIRA: in 'Fatigue '90', Proc. of 4th Int. Conf on Fatigue and Fatigue Thresholds, (ed. H. Kitagawa and T. Tanaka), 1049-1054; 1990, Edgbaston, UK, Materials and Component Engineering Publications.
- *148. J. M. NEWMAN, P. J. GREGSON, P. J. E. FORSYTH, and P. D. PITCHER: *Scr. Metall.*, 1989, **23**, 1157.
149. D. J. NICHOLLS and J. W. MARTIN: *Fat. Fract. Eng. Mater. Struct.*, 1990, **13**, 83.
150. D. J. NICHOLLS and J. W. MARTIN: *Mater. Sci. Eng.*, 1990, **A128**, 141.
- *151. K. T. VENKATESWARA RAO, W. YU, and R. O. RITCHIE: *Eng. Fract. Mech.*, 1988, **31**, 623.
- *152. R. W. HERTZBERG, W. A. HERMAN, and R. O. RITCHIE: *Scr. Metall.*, 1987, **21**, 1541.
- *153. K. T. VENKATESWARA RAO and R. O. RITCHIE: *Acta Metall.*, 1988, **26**, 2949.
154. R. P. WEI and R. I. STEPHENS (eds): 'Fatigue crack growth under spectrum loads', STP 595; 1976, Philadelphia, PA, American Society for Testing and Materials.
155. C. M. WARD-CLOSE, A. F. BLOM, and R. O. RITCHIE: *Eng. Fract. Mech.*, 1989, **32**, 613.
156. D. J. ALEXANDER and J. F. KNOTT: in 'Fatigue '87', Proc. of 3rd Int. Conf on Fatigue and Fatigue Thresholds, (ed.

- R. O. Ritchie and E. A. Starke, Jr), 395–406; 1987, Warley, UK, Engineering Materials Advisory Services.
157. C. ESPINASSE and C. BATHIAS: in Ref. 19, p. 793.
- *158. N. OHRLOFF, A. GYSLER, and G. LÜTJERING: in Ref. 19, p. 801.
159. M. T. YU and T. H. TOPPER: *J. Eng. Mater. Technol.*, 1985, **107**, 19.
160. E. ZAIKEN and R. O. RITCHIE: *Eng. Fract. Mech.*, 1985, **22**, 35.
161. G. V. SCARICH and P. E. BRETZ: 'Fatigue crack growth resistance of aluminum alloys under spectrum loading: Vol. I – Commercial 2XXX and 7XXX alloys.' Technical Report No. NOR 85–141, Northrop Corporation, Hawthorne, CA, 1985.
- *162. G. V. SCARICH, K. M. BRESNAHAN, and P. E. BRETZ: 'Fatigue crack growth resistance of aluminum alloys under spectrum loading: Vol. II – Aluminum–lithium alloys', Technical report No. NOR 85–141, Northrop Corporation, Hawthorne, CA 1985.
163. W. E. QUIST, G. H. NARAYANAN, A. L. WINGERT, and T. M. F. RONALD: in Ref. 18, p. 625.
164. N. J. KIM, R. L. BYE, and S. K. DAS: in Ref. 19, p. 309.
165. W. WANG and N. J. GRANT: in Ref. 17, p. 447.
166. R. D. SCHELLENG: *JOM*, 1989, **41**, (1), 32.
167. P. S. GILMAN: in Ref. 17, p. 485.
168. F. W. GAYLE, N. F. LEVOY, and J. B. VANDERSANDE: *J. Met.*, 1987, **39**, (5), 33.
169. H. KEMPER, B. WEISS, and R. STICKLER: in Proc. 8th Light Metals Cong., Leoben, June 1987, Institut für Metallkunde und Werkstoffprüfung, Montanuniversität.
170. W. RUCH and E. A. STARKE, JR: in Ref. 18, p. 121.
- *171. H. KEMPER, B. WEISS, and R. STICKLER: in 'Fatigue 87', Proc. of 3rd Int. Conf. on Fatigue and Fatigue Thresholds, (ed. R. O. Ritchie and E. A. Starke, Jr), 789–800; 1987, Warley, UK, Engineering Materials Advisory Services.
172. K. V. JATA, W. RUCH, and E. A. STARKE, JR: in 'Light metals', 55–62; 1985, Europe, Materials Research Society.
173. H. KEMPER, B. WEISS, and R. STICKLER: *Eng. Fract. Mech.*, 1989, **32**, 591.
174. M. D. WRIGHT and C. J. BEEVERS: in Ref. 20, p. 1087.
- *175. K. T. VENKATESWARA RAO and R. O. RITCHIE: *Mater. Sci. Eng.*, 1988, **100**, 23.
176. K. VENKATESWARA RAO, J. C. McNULTY, and R. O. RITCHIE: unpublished work.
177. R. C. MALCOLM, F. J. CORDIER, and R. J. BUCCI: 'The effects of various thermal conditions on the tensile properties of aluminum alloy 2090–T8E41 plate', Report No. 56–86–AH429, Alcoa Technical Center, Pittsburgh, PA 1986.
178. R. M. PELLOUX, R. E. STOLTZ, and J. A. MOSKOWITZ: *Mater. Sci. Eng.*, 1976, **25**, 193.
179. R. P. GANGLOFF: in 'Environment induced cracking of metals', (ed. R. P. Gangloff and M. B. Ives), 55–109; 1990, Houston, TX, National Association of Corrosion Engineers.
180. M. GAO, P. S. PAO, and R. P. WEI: *Metall. Trans.*, 1988, **19A**, 1739.
181. R. P. GANGLOFF and D. J. DUQUETTE: in 'Chemistry and physics of fracture', (ed. R. M. Latanision and R. H. Jones), 612–645; 1987, The Netherlands, Martinus Nijhoff.
182. R. E. STOLTZ and R. M. PELLOUX: *Metall. Trans.*, 1972, **3**, 2433.
183. C. P. DERVENIS, E. I. MELETIS, and R. F. HOCHMAN: *Mater. Sci. Eng.*, 1988, **A102**, 151.
184. M. REBIERE and T. MAGNIN: *Mater. Sci. Eng.*, 1990, **A128**, 99.
185. R. E. RICKER: PhD thesis, Troy, NY, Rensselaer Polytechnic Institute, 1982.
- *186. P. S. PAO, M. A. IMAM, L. A. COOLEY, and G. R. YODER: *Corrosion*, 1989, **45**, 530.
- *187. R. S. PIASCIAK and R. P. GANGLOFF: in 'Environment induced cracking of metals', (ed. R. P. Gangloff and M. B. Ives), 233–240; 1990, Houston, TX, National Association of Corrosion Engineers.
188. M. O. SPIEDEL: *Metall. Trans.*, 1975, **6A**, 631.
189. A. GRAY, N. J. H. HOLROYD, and J. WHITE: in Ref. 20, p. 1175.
190. J. G. RINKER, M. MAREK, and T. H. SANDERS, JR: *Mater. Sci. Eng.*, 1984, **64**, 203.
191. R. C. DORWARD and K. R. HASSE: *Corrosion*, 1988, **44**, 932.
192. N. J. H. HOLROYD, A. GRAY, G. M. SCAMANS, and R. HERMANN: in Ref. 16, p. 310.
193. K. T. VENKATESWARA RAO, D. KOVAR, and R. O. RITCHIE: unpublished work, 1990.
194. N. J. H. HOLROYD and D. HARDIE: *Corros. Sci.*, 1983, **23**, 527.
195. A. M. GREEN and J. F. KNOTT: in 'Advances in fracture research', Proc. of 7th Int. Conf. of Fracture, (ed. K. Salama *et al.*), 1747–1756; 1989, London, Pergamon Press.
196. F. J. BRADSHAW and C. WHEELER: *Int. J. Fract. Mech.*, 1969, **5**, 255.
197. R. J. SELINES and R. M. PELLOUX: *Metall. Trans.*, 1972, **3**, 2525.
198. R. P. WEI and R. P. GANGLOFF: in 'Fracture mechanics: Perspectives and directions', STP 1020, (ed. R. P. Wei and R. P. Gangloff), 233–264; 1989, Philadelphia, PA, American Society for Testing and Materials.
199. R. P. GANGLOFF and R. O. RITCHIE: in 'Fundamentals of deformation and fracture', (ed. B. A. Bilby *et al.*), 529–558; 1985, Cambridge, UK, Cambridge University Press.
200. K. T. VENKATESWARA RAO and R. O. RITCHIE: *Scr. Metall.*, 1989, **23**, 1129.
201. L. M. MA, J. K. HAN, R. L. TOBLER, R. P. WALSH, and R. P. REED: *Adv. Cryogen. Eng.*, 1990, **34**, 1143.
202. R. L. TOBLER, J. K. HAN, L. MA, R. P. WALSH, and R. P. REED: in Ref. 20, p. 1115.
203. Y. B. XU, Z. G. WANG, Y. ZHANG, H. H. ZHAO, and Z. Q. HU: in Ref. 20, p. 1147.
204. P. K. LIAW and M. E. FINE: *Metall. Trans.*, 1981, **12A**, 1927.
205. J. MCKITTRICK, P. K. LIAW, S. I. KWUN, and M. E. FINE: *Metall. Trans.*, 1981, **12A**, 1535.
206. P. K. LIAW and W. A. LOGDSON: *Acta Metall.*, 1988, **36**, 1731.
- *207. K. T. VENKATESWARA RAO, R. J. BUCCI, K. V. JATA, and R. O. RITCHIE: *Mater. Sci. Eng.*, 1991, **A141**, 39.
208. R. CROOKS and M. R. MITCHELL: in Ref. 20, p. 1023.

THE PREHISTORY OF METALLURGY IN THE BRITISH ISLES

R. F. Tylecote

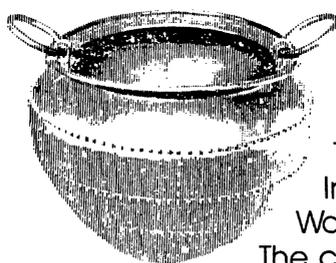
Based upon the author's 'Metallurgy in Archaeology' (published in 1962), which was one of the few books on this subject and became a minor classic in its time, this revised text incorporates the results of work done in the scientific investigations of archaeology between 1960 and 1982.

Since this is a subject of interest to metallurgists as well as archaeologists, the work has not been treated as a textbook on metallurgy for the archaeologist, but is addressed to both archaeologists and metallurgists.

CONTENTS

The native metals

Copper and copper alloys



The production and properties of tin and its alloys

Lead, silver and antimony

Fabrication

The coming of iron

The Roman Iron Age

Iron in the early medieval period

Water-powered bloomeries: the end of an era

The charcoal blast furnace and the finery

Fuels and their ashes

Appendixes

Index

Book 319 (1986) 297 x 210 mm 256pp ISBN 0 904357 72 4 Casebound

UK £25.00 (Institute of Materials and Historical Metallurgy Society members discount price £20.00)

Overseas US\$50.00 (Institute of Materials and Historical Metallurgy Society members discount price US\$40.00)

Please send orders with remittance, quoting membership number (where applicable) to:

The Institute of Materials

Sales & Marketing Department

1 Carlton House Terrace

London

SW1Y 5DB

Tel: 071-839 4071; Fax: 071-839 2078; Tlx: 8814813