Indigenization of high-strength, light-weight Al–Li alloys for aerospace

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Abstract. A research programme was initiated at the Defence Metallurgical Research Laboratory, Hyderabad, a decade ago for the indigenous development of Al–Li alloys in order to finally meet the requirements of the space and aircraft industries in the country. This paper describes the systematic studies carried out in the laboratory to overcome the initial difficulties in producing sound ingots, optimize the subsequent heat treatments and processing schedules, and to finally obtain reproducible microstructure and mechanical properties in the semi products (i.e. sheets and extrusions) developed. Laboratory-scale sheet and extrusion products meet tensile property specification of 8090 alloy. One of these semiproducts, i.e. round bar extrusion, is currently being supplied for the stallite programmes. Commercial-scale sheet products made in Russia under an Indo-Russian joint programme have been made available for the aircraft programme. Recent alloy development studies in the laboratory are discussed within the context of the present paper.

Keywords. Al–Li alloy; ingot casting; heat treatment; mechanical processing; sheet; extrusion; mechanical properties; corrosion.

1. Introduction

Since the recognition of the age-hardening capability of an aluminium alloy containing Cu and Mg in 1906 (Wilm 1911), many complex Al alloys have been developed. Amongst them, Al–Cu–Mg based 2000, Al–Zn–Mg based 7000 and Al–Mg–Si based 6000 series alloys have become very popular. Both 2000 and 7000 series alloys are well known for their use as structural materials for aircraft. In aircraft engineering, there are continuing pressures to develop materials with high specific properties. In the mid 1970s, the changing economics of civilian and military aircraft fleet operation costs, the advent of high-strength, low-density fibre-reinforced composite materials, and in particular of organic matrices with carbon fibre reinforcement accentuated the search efforts for new metallic materials. In metallic systems, one method of achieving superior specific properties is to reduce the density of an alloy while maintaining or increasing the property levels. Reductions in alloy density are more effective than increases in strength or stiffness at reducing aircraft mass (see figure 1). A 10% increase in tensile strength gives approximately 3% mass saving, whereas a 10% decrease in density achieves a 10% mass saving, provided the mechanical properties of the low-density alloy match those of the alloy it replaces (Ekneil *et al* 1984). A direct consequence of these practical considerations is that a family of Li-containing Al-based alloys, which were in fact known since the 1950s, reemerged as potential candidates for reexamination and large-scale development.

1.1 Why Al–Li Alloy

Li (having atomic number 3) is the lightest metallic element in the periodic table. Compared to the density of 2.7 g/cm³ of Al, the density of Li is 0.53 g/cm³. As a

Figure 1. Effect of individual property changes on structural weight savings.

result, Li addition to Al reduces the density by about 3% for each wt% Li added. This is most impressive as weight saving in structure is dominated by a linear relationship with density. Li is an “open” alkali metal (Hume Rothery *et al* 1969). The atomic diameter of Li refers to a relatively compressible structure with a very dilute gas of valency electrons. When Li is dissolved in Al, which is much less compressible, the much stronger bonding forces of Al control the structure to such an extent that the apparent atomic diameter (AAD) value of Li is much smaller than would be expected from Vegard’s law (Hume Rothery *et al* 1969). This is the reason why Li addition reduces the lattice parameter of Al, and simultaneously increases the elastic modulus (Mukhopadhyay and Chaturvedi 1989). For a binary alloy containing 2.5 wt% Li, the elastic modulus increases nonlinearly from 66 to 80/82 GPa, the maximum effect on modulus increase being obtained due to the presence of Li in the solid solution in Al (Noble *et al* 1982).

Li is one of the four elements (Zn, Ag, Mg and Li) that have more than 10 at% solid solubility in Al. The maximum solid solubility of Li in Al is about 4 wt% (13.9 at%) at the eutectic temperature of 601°C. Below this, the solid solubility falls rapidly to a low value at room temperature (McAlister 1982). As with the other age-hardenable aluminium alloys, this decrease in solid solubility with decreasing aging temperature provides the basis for strengthening via precipitation hardening by solution treatment and subsequent aging operations. In the case of Al–Li alloys, the major strengthening upon aging heat treatments stems from the widespread uniform precipitation of a metastable coherent phase, δ' , based on Al_3Li (LiI_2 structure) (Silcock 1959–1960). Li is therefore unique among the possible alloying elements for Al in that it provides the basis for major precipitation hardening while simultaneously reducing density and increasing modulus.

1.2 Historical development of Al–Li alloy

The attempts to realize the potential benefits of increased modulus and reduced density of Al–Li alloys date back 60 years, although it was not until the 1950s when

the Al–Cu–Li AA 2020 was developed in the USA by the Aluminum Company of America (ALCOA). Compared to 7075, which is widely used in the manufacture of aircraft components, AA 2020 has approximately 3% lower density and 7% higher elastic modulus; commercial application of this alloy was made by the US Navy. However, further research revealed that the alloy was notch-sensitive and had a lower fracture toughness than 7075. New US airplane damage tolerance requirements could not be met and the result was the withdrawal of AA 2020 from commercial production, although it continued to be used successfully in existing aircraft. In the USSR, at about the same time, VAD 23, an alloy similar in composition to 2020, was developed. Later, the Soviets investigated and developed a series of Al–Mg–Li alloys referred to as 01420. However, these alloys too exhibited poor ductility and fracture toughness properties. Attempts to improve the mechanical properties were made, but without any success. It is now realized that development work was carried out in those days without a proper understanding of the strengthening and deformation mechanisms in Al–Li alloys. In the mid and late 70s, the deformation mechanisms and the principal factors responsible for the brittle behaviour of Al–Li alloys were identified, and the difficulties associated with the casting of high-quality Al–Li ingot were largely overcome. This very substantial progress together with continuing large-scale development programmes inspired Al producers to reexamine and produce more complex Al–Li alloys via both ingot and powder metallurgy routes. In 1983, the British ALCAN International in association with RAE developed, via ingot metallurgy route, the first of the new generation of Al–Li alloys referred to as DTD XXXA (the Aluminum Association code being AA 8090). The DTD XXXA alloy was designed to replace medium-strength 2014-T6 and 2024-T3 plate and sheet in aircraft. In the following years, a higher-strength alloy DTD XXXB was also developed in the UK to substitute for 7000 series Al alloys. Since the introduction of these two alloys, several other Al–Li alloys of varying compositions have been developed by other leading Al producers in the US, France and USSR. The specific weight of these alloys ranges from 2.5 to 2.6 g/cm³ and the elastic modulus from 78 to 81 GPa. This compares with the average values of 2.79 g/cm³ and 71 GPa respectively for the existing aircraft alloys based on Al. It should be pointed out at this stage that during the initial period of Al–Li alloy development, several investigators examined the Al–Li alloy products manufactured by both powder and ingot metallurgy routes. It was realized that the strength–ductility combinations are either equal to or even better in the ingot metallurgy route products than in those obtained via powder metallurgy route. The cheaper ingot metallurgy route has therefore been obviously chosen for commercial production of today's Al–Li-based alloys.

1.3 Development efforts at DMRL

In the mid 80s, a research programme was initiated at the Defence Metallurgical Research Laboratory (DMRL), Hyderabad, for the indigenous development of Al–Li alloys (via ingot metallurgical route) in order to finally meet the requirements of the space and aircraft industries in the country. It is the purpose of the present paper to describe the systematic studies carried out at various stages of the continuing Al–Li alloy development programme in the laboratory. Current applications of these alloys, together with recent developments on this alloy type, are also discussed within the

Table 1. Compositions limit for various 8090-type Al–Li alloys.

	Li	Cu	Mg	Zr	Fe	Si
8090	2.2–2.7	1.0–1.6	0.6–1.3	0.04–0.16	< 0.3	< 0.2
DMRL	2.0–2.5	1.0–1.6	0.6–1.3	0.06–0.12	< 0.15	< 0.1
1441	1.6–2.1	1.6–2.1	0.6–1.3	0.04–0.16	< 0.15	< 0.1

context of the present paper. At DMRL, the Al–Li development programme has so far been concerned with mainly the Al–Li–Cu–Mg–Zr-based 8090 alloy (see table 1).

2. Melting, casting and homogenization

2.1 Melting and casting

During the initial stages of the Al–Li alloy development programme, melting and casting of these alloys posed great challenges because the related technologies were not available in the open literature and are mostly proprietary in nature. Melting of Li poses a number of problems due to its high reactivity (Averil *et al* 1981; Divecha and Karmarkar 1981; Starke *et al* 1981). Molten Li reacts with water vapour, oxygen and nitrogen. As a result of its reaction with water vapour, Li oxide forms and hydrogen gets dissolved in the liquid. Dissolved hydrogen can cause microporosity if not removed prior to pouring. Even exposure of molten Li to air during pouring is sufficient to cause oxidation and harm ingot quality. Gas pick-up is accordingly found to be more serious in Al–Li alloys than in conventional Al alloys (Starke *et al* 1981). In order to minimize reaction of Li with the atmosphere, two approaches have been adopted. The first approach consists of covering molten Li by halide fluxes. LiCl and LiF are highly stable compounds and are likely to protect the molten metal (Kamaludeen *et al* 1987). The second approach to minimize Li reaction with the atmosphere is to melt it in an inert atmosphere. Another major problem arising due to Li reactivity is its attack on crucible materials. The attack on crucible materials increases with temperature and concentration of dissolved gases. A low-carbon variety of 304 stainless steel has been reported to give excellent service for Li melting. However, this crucible could not be used for Al melting due to the continual increase in Fe pick-up with time (Chakravorty and Chakravorty 1990). Accordingly, Al and Li have to be melted separately. Two approaches have been adopted in this regard. In one investigation, Li was encapsulated in super-purity Al and then plunged into molten Al with a graphite rod (Jones and Das 1959–60). Another method is to transfer Al melt to Li melt (Chakravorty and Chakravorty 1990).

In DMRL, a twin-crucible melting technique with a combination of flux and argon protection has been developed for the production of up to 50 kg cast ingot (Chakravorty *et al* 1987; Chakravorty 1988; Gokhale *et al* 1988, 1990; Singh *et al* 1989). Special designs were incorporated in the metal mould to minimize melt turbulence and to promote feeding of solidification shrinkage. The argon flushing in the metal mould together with suitable trace-alloy additions are found to improve casting quality. Further studies involving evaluation of the casting characteristics of Li-containing Al alloys showed that these alloys undergo severe metal mould reactions

with sand moulds prepared with sodium silicate binders (Chakravorty and Chakravorty 1991). Alternative mould materials for successful castings have been identified. A high-quality ingot is characterized by its low alkali impurity content, low gas content, low inclusion content, consistent composition, high surface quality, and should necessarily be crack-free. The tendency for hot tearing and porosity formation decreases with reduction in the as-cast grain size. Also, a fine as-cast grain size results in wider distribution of the nonequilibrium solidification products, thus leading to faster dissolution during subsequent processings. Furthermore, in the case of the Al–Li alloys, the final grain structure depends to a great extent on the initial cast structure, since recrystallization in the wrought structure is very difficult to achieve. The latter is partly due to the presence of Zr which is added to control the grain structure during hot working and solution treatment (Ryum 1969; Nes 1972). The grain (albeit the cast grain structure) refining additions of TIBAl has been optimized in this regard. Our studies have shown that 0.4% of TIBAl is required for optimum grain (in the cast structure) refinement (Chakravorty 1988).

2.2 As-cast microstructure and homogenization

Figure 2a shows a light micrograph of a typical as-cast microstructure of an Al–Li–Cu–Mg–Zr-based 8090 alloy produced at DMRL. The micrograph shows that elemental partitioning has occurred during solidification resulting in second-phase precipitation at the boundaries between the grains. Small precipitates observed within the grains formed during cooling of the ingot. SEM studies together with the X-ray mapping of the grain-boundary phases within the as-cast 8090 alloy revealed that the major precipitate at the grain boundary contains Cu and Mg. While Li could

Figure 2. Light micrographs showing (a) interdendritic copper-rich eutectic phases in the as-cast material and (b) homogenized material produced due to annealing at 535°C for 24 h.

not be analysed using X-ray analysis, SIMS analysis of the as-cast 8090 alloys revealed that the Cu, Mg-bearing major grain-boundary phases also contain Li (Mukhopadhyay *et al* 1992). TEM studies conclusively showed that the major coarse g.b. phases are either $T_2(Al_6CuLi_3)$ and/or their approximant variants (Satyaprasad 1990).

With due considerations of the presence of a low-temperature eutectic in the as-cast alloy at about 547°C, the alloy composition, the size of the excess phase particles and that of the dendritic cells in the as-cast structure, an optimum homogenization schedule was determined. Figure 2b shows a light micrograph of the homogenized alloy. It was found that the Fe- and the Si-bearing particles (which separated from the melt early during the solidification) survived the optimum homogenization treatment.

3. Raw materials selection

Deterioration of the strength–ductility combinations as functions of Fe content in Al alloys is a well-known phenomena (Speidel 1975). Presence of a higher amount of Si, i.e. > 0.08 wt%, in Al–Li alloys poses an additional problem by promoting formation of an AlLiSi phase at grain boundaries. The AlLiSi phase survives the optimum homogenization treatment and serves as an active nucleation site for surface pitting upon exposure to salt water and greatly diminishes the resistance of the materials to stress corrosion cracking (Lewis *et al* 1987). The above considerations necessitated using high-purity Al containing 0.08% Fe and 0.07% Si for Al–Li production. Also, alkali impurities such as Na and Ca are also found to result in grain-boundary weakening (Webster 1981). This required further control on raw material impurity contents and processing. Based on the purity requirements projected by DMRL, the National Aluminum Company (NALCO) undertook the development of 99.85% purity Al and has since been able to produce Al to the required purity on a tonnage scale. 99.8% minimum purity Li metal was imported from LITHCO and Cyprus Foote Mineral Company. Other raw materials were available indigenously.

4. Basis for commercial Al–Li alloy composition, and optimum precipitate microstructure

In commercial Al–Li alloys, the major strengthening is due to the precipitation of ordered ($L1_2$ type) and coherent δ' (Al_3Li) phase which forms during or immediately after quenching. The microdeformation characteristics in these alloys are dominated by the interaction of mobile dislocations and the coherent and ordered δ' precipitates. Up to the peakaged condition, such coherent δ' precipitates are sheared by the moving dislocation and planar slip is the result. It has been shown by numerous investigators that such slip coplanarity is the major cause of poor ductility and fracture toughness in these alloys (Sanders and Starke 1982; Noble *et al* 1982; Gregson and Flower 1985). Recent commercial alloys are therefore suitably designed to form semicoherent phases such as $S(Al_2CuMg)$ and/or $T_1(Al_2CuLi)$ in order to disperse coplanar slip leading to slip homogenization in the material. This is the basis for the development of Al–Li–Cu- and Al–Li–Cu–Mg-based alloys. These alloys normally contain trace amounts of Zr to refine grain and to inhibit recrystallization. Zr rather than Cr or

Mn has been chosen additionally because of reduced quench sensitivity considerations (Holl 1969).

Strain localization can occur in regions denuded of strengthening precipitates called precipitate-free zones (PFZs). These PFZs form near grain boundaries and some equilibrium phases. Studies have shown that the presence of coarse grain-boundary precipitates such as T_2 and δ constitute a major basis for premature grain-boundary failure and low ductility and fracture toughness of these alloys (Vasudevan and Doherty 1987).

Although, as mentioned earlier, planar slip results in low toughness, it can lead to improved fatigue crack growth resistance in three ways. Shearable precipitates promote slip reversibility, cause crack deflections along slip planes, and cause crack branching (Vasudevan *et al* 1984). Accordingly, "Our current understanding of an ideal microstructure for this alloy is the one characterized by the presence of a uniform and fine distribution of both shearable (δ') and non-shearable (such as S) precipitates in the matrix, a reduced width of the PFZs around the grain boundaries, and by the minimal presence of coarse grain boundary precipitates" (Mukhopadhyay and Chakravorty 1993). At DMRL, the 8090 composition was selected as the baseline composition for initial development. The casting quality as well as ductility of the semiproducts were found to decrease with Li content exceeding 2.5 wt%. Therefore, a lower Li content of < 2.5 wt% is used in the alloy production. Similarly, a lower Cu content as well as a higher Mg:Cu ratio (within the 8090 alloy composition) reduce quench sensitivity and coarse g.b. precipitation, and maximize uniform distribution of desirable Al_2CuMg -based S precipitates in the heat-treated microstructure (Mukhopadhyay *et al* 1990b). Details of the alloy design aspects of these alloys have been examined by several workers (Flower and Gregson 1987; Harris *et al* 1987; Mukhopadhyay *et al* 1987b; Sainfort and Dubost 1987; Sanders and Starke 1989).

5. Heat treatment

High-temperature annealing treatments of Al–Li-based alloys, such as homogenization and solution treatments, involve loss of lithium from the surfaces due to oxidation and hence care should be taken to optimize such treatments. For commercial Al–Li-based alloys, the recommended solution treatment temperatures lie in a narrow range, viz. $535 \pm 5^\circ\text{C}$. Solution treatment time, on the other hand, varies from 30 to 90 min depending on the microstructure of the alloy produced due to prior deformation; materials processed at lower temperatures require extended solution treatment in order to solutionize T_2 -type coarse particles which form due to thermal exposure (Mukhopadhyay *et al* 1987a).

Aging heat treatments of alloys of commercial interest involve temperatures lying within the metastable $\alpha + \delta'$ solvus. The maximum aging temperature is always chosen below 200°C . This is because the strengthening falls off sharply when aged above 200°C ; this is attributed in turn to a decrease in δ' (Al_3Li) volume fraction and rapid coarsening of δ' precipitates. Keeping in mind the sluggish precipitation kinetics in Li-containing commercial Al alloys (Flower *et al* 1986, Gregson *et al* 1986), the lower limit of the aging temperature is specified at or above 150°C . The commonly used aging temperature ranges from 150 to 190°C . With increasing aging temperature, the diffusion in the grain-boundary regions is enhanced, leading to rapid grain-boundary

Figure 3. Age hardening curves developed at (a) 170°C and (b) 190°C.

Table 2. Properties of DMRL 8090 sheet products.

Product	Direction	0.2% PS (MPa)	UTS (MPa)	Elong. (%)	Fracture toughness (MPa√m)
Hot rolled sheet	L	404	475	6.0	(L-T):66 (T-L):46
	LT	416	487	7.0	
	45°	377	460	9.1	
Cold rolled sheet	L	370	455	8.5	—
	LT	375	447	8.5	
	45°	354	430	9.5	
Specified	—	360	420	6.0	75

precipitation and formation of the associated PFZs. The present trend is therefore to use a lower aging temperature, combined with prior deformation by stretching (to increase the density of the heterogeneous nucleation sites for precipitation in terms of dislocations), thus attaining the peakaged strength within a shorter time as well as minimizing the extent of grain-boundary precipitation and the associated PFZs (Ashton *et al* 1986).

Figure 4. Development of microstructure in the alloy when solution-treated quenched, 2% stretched and aged at 170°C for 24 h: (a) δ' (Al_3Li) phase, (b) S (Al_2CuMg) phase and (c) δ' PFZ adjacent to the grain boundary (arrowed).

Figure 3 shows the age-hardening curves determined for the DMRL alloy at temperatures of 170 and 190°C. It was found that preaging treatments such as 2% stretch or natural aging for 14 days did not change the hardness of the materials appreciably, except during the initial stages of the artificial aging. The 170°C aging produced a maximum hardness. Microstructural studies revealed that the treatment involving a 2% stretch followed by aging at 170°C produced the optimum heat-treated

microstructure (see figure 4). The longitudinal tensile properties of the hot-rolled sheets subjected to different heat treatments (see table 2) support the microstructural analyses.

6. Mechanical processing

Homogenized ingots were scalped to remove the oxidized layers formed on the surfaces and were subsequently processed by rolling or extrusion. Hot-working parameters were first optimized by evaluating the intrinsic workability of the alloys over a wide

Figure 5. (a) Process stability parameter contour map for DMRL Al-Li cast and homogenized alloy and (b) process efficiency parameter contour map for DMRL Al-Li cast and homogenized alloy.

Figure 6. Light micrograph showing development of grain structure at the mid thickness region of a 4 mm hot-rolled sheet.

Figure 7. Longitudinal tensile properties of hot-rolled sheets as functions of varying aging treatments.

temperature-strain domain, and then implementing suitable combinations of parameters during extrusion and rolling trials. The intrinsic workability of the alloy was determined from the flow curves from hot-compression tests, and then evaluating the stability and efficiency parameters (Prasad 1990). Figure 5, a and b, shows the stability and efficiency maps for the DMRL as-cast and homogenized alloys respectively (Prasad 1993). Positive stability parameters signify stable and uniform deformation, while higher efficiencies are generally associated with greater degree of recrystallization. Higher temperatures and lower strain rates are found to give uniform deformation and greater degree of recrystallization. Compared to many of the other Al alloys, the rather strong influence of the extrusion temperature on the degree of recrystallization observed in the present 8090 alloys during extrusion is closely related to the dissolution of the coarse T_2 -type precipitates in the microstructure; recrystallization is maximum at temperatures at which these coarse precipitates dissolve (Mukhopadhyay *et al* 1990b).

6.1 *Sheets*

Sheets of thicknesses 1 to 4 mm were successfully produced. A typical light microstructure of 4 mm (hot-rolled) sheet product is shown in figure 6. The tensile properties of the sheet products evaluated at different orientations with respect to the rolling direction are also presented in figure 7. It is noteworthy that the differences in the property levels between any two of the three directions are small, and that these properties are comparable to the minimum properties specified for aircraft applications.

Fatigue crack growth rates were evaluated in L-T and T-L orientations under constant-amplitude loading as well as LCA spectrum loading conditions. The alloys produced at DMRL were found to have equivalent or lower FCG rates compared to those of 2014 alloy (see figure 8, a and b).

6.2 *Extruded products*

Rods of 15 to 40 mm and plates of 100×25 and $110 \times 12 \text{ mm}^2$ have been extruded. The as-extruded microstructure revealed that 60% of the peripheral cross-sectional area shows recrystallized grain structure, while the core portion (40%) shows partially recrystallized grain structure (figure 9). The variation in the grain structures observed across the round bar extrudes is also found in the case of many other similarly treated Al alloys. These changes are consistent with the previously documented results on Al-Li alloys suggesting the presence of a strain gradient across the extrude section, where the strain is maximum at the surface and minimum at the core (Mukhopadhyay *et al* 1987, 1990a). Table 3 presents the average tensile properties of the extrudes subjected to the T8 temper. The properties favourably compare with the 8090 minimum property requirements. This product is currently being supplied by DMRL/NFTDC to ISRO Satellite Centre for satellite applications. Table 3 shows that the notch tensile strength values of the DMRL extrusions are comparable to those evaluated from thick plates of commercial Al-Li alloy. Fracture toughness values in C-R orientation of the DMRL extrusions are comparable to those of thick plates of 8090 alloy in L-ST orientation, but inferior to that of 2014 alloy.

Figure 8. Fatigue crack propagation of DMRL hot-rolled alloy, 3 mm thick, $R = -0.3$, in comparison with 1440 (3-2), L73 (2014) and 8090 C alloys: (a) LT and (b) TL orientation.

Figure 9. Light micrographs showing variation in grain structure across the extrude section: (a) periphery is fully recrystallized, while (b) the core is unrecrystallized.

7. Commercial-scale production trials

Alloy 1441 was produced in collaboration with NFTDC, Hyderabad, and Russian industries. About 2 tonnes of sheet materials were produced in sizes $1200 \times 3000 \times 1.2 \text{ mm}^3$ (bare and clad), $1200 \times 4000 \times 2 \text{ mm}^3$ (bare and clad) and $1200 \times 4000 \times 8 \text{ mm}^3$ (bare). The tensile properties of the sheet products presented in table 4 are found to be well balanced in different directions with respect to the rolling plane, and also

Table 3. Properties of DMRL 8090 extruded rods.

Alloy	0.2% PS (MPa)	UTS (MPa)	Elong. (%)	NTS PS	KIC (MPa $\sqrt{\text{m}}$)
DMRL	395	465	8.1	1.44	(C–R) 18.9
8090A	380	460	7.0	1.40	(C–R) 10.0
Spec	360	430	4.0	—	—

(Tensile and notch tensile properties in L direction)

Table 4. Tensile properties of 1441 alloys.

Product type	Direction	0.2% PS (MPa)	UTS (MPa)	Elong. (%)
8 mm	L	379	450	9.6
Unclad	LT	375	472	8.9
	450	360	460	6.9
2 mm	L	412	465	8.7
Unclad	LT	389	469	9.7
	450	396	473	9.2
2 mm	L	399	456	9.3
Clad	LT	425	482	8.8
	450	392	461	10.3
1.2 mm	L	401	455	9.7
Unclad	LT	365	445	12.0
	450	375	445	12.3
1.2 mm	L	394	441	9.0
Clad	LT	404	452	8.3
	450	378	430	10.4
Specs		360	420	6.0

compare favourably with the specified property requirements. Fatigue properties have also been evaluated extensively and are found to be largely independent of test orientation and sheet thickness. Cladding is found to increase the fatigue crack growth rate by a small extent (Saxena 1993). This result is consistent with the previously documented results on smooth specimen fatigue behaviour (S–N curve) (Polmear 1989).

8. Work in progress and future plans

The alloys are being evaluated for corrosion properties at DMRL as well as in collaboration with BE college, Shibpur; IIT, Kanpur; and DMSRDE, Kanpur. Initial work indicates that the exfoliation corrosion resistance of DMRL Al–Li alloys is superior to that of 2014 alloy (Banerjee 1993), and that it is dependent on the degree

of aging with the underaged alloys being immune to exfoliation corrosion (Tripathi 1993). Recent studies have shown that the rate of corrosion of DMRL alloys in corrosive media (such as 3.5% NaCl solution) is markedly reduced in the presence of trace additions of Zn (Singh *et al* 1994). Work involving development of 8090 alloys having combination of higher strength, toughness and stress corrosion cracking (SCC) resistance are in progress (Singh *et al* 1994).

Plans to scale up the melting and casting process are in progress.

9. Summary

Laboratory-scale Al–Li–Cu–Mg-based 8090 alloy semiproducts have been developed with microstructure and tensile properties comparable to those of commercial Al–Li alloys. One of the semiproducts, i.e. extruded rod products, are being supplied to ISRO for the space programmes. Alloys having improved combinations of strength, toughness, and corrosion and stress corrosion cracking resistance properties are being developed. Scale-up of the existing production technology is in progress.

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