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Progress in joining of advanced materials

Part 2: Joining of metal matrix composites and joining of other advanced materials

G. Çam and M. Koçak

Advanced materials generally require novel joining techniques. Developments in new materials research should be conducted hand in hand with work on weldability and joining capacity aspects. Sound joint quality for any new material has always been considered a milestone in a research and development scheme for a new material, particularly in terms of widespread applications. Better understanding of the microstructuremechanical properties relationships of the bonded or welded joints will feed back to the materials development activities both in conventional and new materials areas. The two joining processes diffusion bonding and laser welding are considered in this literature review, since these processes are capable of joining a wide range of materials of interest in the aerospace industry, as well as in many other industrial applications, and offer remarkable advantages over conventional fusion welding processes. Of particular interest is the ability to join the more difficult aerospace alloys with minimal component distortion and high reproducibility of joint quality. The purpose of the review is to outline progress made in this area and to make suggestions for future work. Part 1 (STWJ, 1998, 3, (3), 105-126) dealt with solid state joining, fusion joining, and joining of intermetallics, while Part 2 covers joining of metal matrix composites and joining of other advanced materials.

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JOINING OF METAL MATRIX COMPOSITES

Metal matrix composite (MMC) materials have attracted great interest in recent years owing to their superior strength/weight ratios. These materials have wide fields of potential application as a result of not only their very high specific strength and elastic modulus but also their high resistance to elevated temperature and better wear resistance than many materials. However, the poor formability of MMCs, arising from their anisotropic properties and stiff reinforcement, considerably hinder their practical application.

- A MMC consists of two distinctly different materials:
- (i) the matrix, typically an alloy of aluminium, titanium, magnesium, copper, iron, etc.
- (ii) the reinforcement, a fibre or an intermetallic compound such as an oxide, carbide, or nitride.

The term MMC is nowadays used to encompass metal matrix composites reinforced by particles, whiskers, or

either continuous or discontinuous fibres, as well as metal matrix oxide dispersion strengthened (ODS) materials. Since these groups of materials have different strengthening mechanisms, the joining phenomena and problems encountered in joining are also different. Of these groups of materials, fibre reinforced MMCs are the most difficult to join in general.

The purpose of this section of the review is to provide brief information on the weldability of ODS alloys and MMCs, particularly fibre reinforced MMCs, and the problems involved.

Joining of ODS alloys

Alloying can only improve high temperature properties in a certain usable range of temperatures. Simple alloys, even those solid solution strengthened and precipitation hardened, reach the limits of their use below the requirements of many engineering applications. Oxide dispersion strengthened nickel and iron based alloys are potential materials to be used in service at temperatures well above 1100°C. In these materials, the high temperature strength is increased by introducing a small, stable oxide dispersoid into the metal matrix (with elongated grain structure providing considerable strength). The fine oxide dispersoids act to pin the dislocation movement in a manner similar to that in the case of second phases in precipitation hardened alloys.

For example, mechanically alloyed Incoloy alloy MA 956, which is a ferritic ODS superalloy containing yttrium oxide (yttria) dispersoid, exhibits excellent strength and resistance to corrosion at temperatures up to 1260° C. Because of the complex nature of the alloy's structure, joining procedures must be carefully designed to produce joints with optimum properties and avoid degradation of the base metal properties. Excessive weld heating of the ODS alloy can cause oxide coalescence to the point that the size of the oxide dispersoid is no longer effective in pinning the dislocations. Melting, however, diminishes the elongated grain structure and leads to an equiaxed grain structure, which results in reduced strength.¹⁷⁴

Solid state joining processes therefore offer good potential for successive joining of ODS alloys since melting of the base metal is avoided. Work by Moore and Glasgow¹⁷⁵ has demonstrated that diffusion bonding of a nickel base ODS alloy, Inconel alloy MA 6000, is feasible. Hammelmann *et al.*¹⁷⁶ studied hipping–diffusion bonding of ODS materials, namely MA 754, MA 956, MA 6000, and Incoloy 800H, using plasma sprayed encapsulation and reported that the correct choice of hipping parameters was essential. They suggested that a maximum temperature of 1050°C must be used to avoid uncontrolled recrystallisation and recommended hipping pressure ranges of 150–200 MPa for MA 956 and Incoloy 800H, of 250–350 MPa for MA 754, and of 380–400 MPa for MA 6000. Use of transient liquid phase (TLP) diffusion bonding is also promising in maintaining the parent alloy microstructure in the joint region. Recent work has demonstrated that transient liquid phase diffusion bonding of ferritic ODS alloys, namely MA 957, MA 956, and PM 2000, using a novel amorphous foil based on an Fe–B–Si composition, can be successfully applied.¹⁷⁷ However, the results showed that success in achieving grain growth across TLP bonds is not only dependent on the composition but is also strongly dependent on the recrystallisation behaviour of individual alloy types, possibly owing to differences in prior thermomechanical processing.

Results thus far indicate that solid state welding processes are excellent candidates for joining ODS alloys. Since no melting of the base metal is involved, the problems of grain structure alteration and deslagging of the dispersoid can be avoided. Accordingly, solid state joints in ODS alloys exhibit properties superior to those of fusion weldments. Considerable research has been and continues to be conducted to define suitable solid state joining procedures for ODS alloys.

Fusion processes have also been explored for joining ODS alloys. Gas tungsten arc (GTA) welding is commonly used for joining of sheet metals and, thus, can be applied to joining of ODS alloys. Both autogenous and filler metal weldments have been made with alloy MA 956 sheet. However, it must be remembered that the alloy is ferritic and, therefore, exhibits a ductile-brittle transition temperature. Consequently, a preheat of about 200°C is recommended. Furthermore, slow cooling may be required for complex welded structures to avoid post-weld cracking. Even then, the grain structure of the weldment becomes equiaxed whereas that of the base metal is still elongated. In addition, the dispersoid floats out of the weld area while molten, e.g. when deslagging. As a result, GTA weldments exhibit strength well below that of the base metal. In many cases, design considerations can account for the lower strength by placing weldments in areas of lower stress and/or temperature. The GTA welding technique can also be used for applications where corrosion resistance at high temperatures is required rather than strength.

Flash butt welding of alloy MA 956 leads to a weldment structure which is to some extent comparable to that of GTA weldments. In resistance spot welding of alloy MA 956, grain growth in the weldment in the opposite direction to that of the grain structure of the base metal occurs, which results in inferior weld properties. Resistance seam welding results in agglomeration and rejection of the dispersoid in the weldment and, thus, also leads to inferior weld properties.

General comments on ODS alloys

Power beam welding processes, such as electron beam and laser welding, melt a minimum of base metal as a result of the concentrated heat application in these processes. Porosity and dispersoid agglomeration can be significantly reduced and, thus, weldments of better quality than those produced by other fusion processes can be realised. Joining of ODS alloys by power beam processes has not been widely applied so far; suitable procedures for joining of ODS alloys by power beam processes as well as diffusion bonding are yet to be fully established.

Joining of reinforced MMCs

Reinforced MMCs are of very great interest in meeting the demands of high performance structures owing to their high strength and low density. The lower manufacturing cost and ability to shape discontinuously reinforced composites using standard forming processes such as forging, rolling, and extrusion make this class of MMC especially attractive for aerospace, automotive, and electronic applications. Presently, suitable powder metallurgy and casting techniques are available for production of MMCs, however, the difficulty of joining these materials, particularly fibre





reinforced MMCs, has to be overcome. Fibre reinforced MMCs which are attractive high performance materials with potentially wide application suffer especially from the difficulty of joining fibres at the joint; joining of these popular structural materials is of paramount importance. To date, extensive work has been conducted on the joining of fibre reinforced MMCs, especially the aluminium matrix composites, whereas work on the joining of other MMCs has been limited. An excellent review on joining of MMCs was produced by Ellis.¹⁷⁸

An urgent need exists to establish new joining techniques to produce reliable joints to use the full potential of fibre reinforced MMCs.

As pointed out above, fibre reinforced MMCs are the most difficult to join in general. Taking into consideration the nature of anisotropic fibre reinforced MMCs, it is vitally important to prevent melting (e.g. using solid state diffusion welding) or at least to minimise the melted region (e.g. using power beam processes) in order to join these materials satisfactorily. However, it is now well established that power beam (i.e. electron beam (EB) and laser beam (LB)) weld processes run a high risk of causing Al₄C₃ formation when welding SiC reinforced aluminium matrix composites.¹⁷⁹

Both the elevated temperature and pressure used in joining might affect the weld properties. Typical degradations of MMCs that might take place during joining are:¹⁸⁰

- (i) deterioration of strength at the fibre/matrix interface owing to formation of brittle phases at the interface at elevated temperature
- (ii) degradation in strength of MMCs owing to destruction or change in configuration of the fibre caused by plastic deformation of the matrix during the applied force used in the joining process (Fig. 1)
- (iii) poor joint efficiency owing to discontinuity in the fibre and to the change in configuration of the fibre in the reaction layer at the bonding interface (Fig. 2).

Use of conventional fusion welding processes to join MMCs has a major difficulty caused by the problems related to porosity and clustering of reinforcing particles (e.g. SiC) in the weld region. Since solid state diffusion bonding and friction welding do not involve melting of the parent plate, they can be considered to be ideal joining processes for aluminium based MMCs.

Degradation of the strength of MMCs owing to formation of a brittle phase at the fibre/matrix interface can be typically experienced in aluminium matrix composites after heating to a sufficiently high temperature. For example, brittle AlB₂ forms at the fibre/matrix interface in Al–B composite when it is heated to a temperature above about 425° C, leading to the deterioration of strength.¹⁸³ In a



2 Schematic illustration of discontinuity of fibres at bonding interface of metal matrix composite reinforced with fibres¹⁸²

similar manner, carbon fibres react with the aluminium matrix in Al–C composite and form brittle Al_4C_3 at the fibre/matrix interface at temperatures above about 575°C.¹⁸⁴ In order to avoid the reaction between the fibre and the matrix, fibres are sometimes coated with carbides such as SiC, TiC, or ZrC. Although these coatings retard the reaction at the interface, brittle reaction products will still form after prolonged heating.

Since plastic deformation of the matrix causes destruction or change in configuration of the fibre leading to deterioration of joint strength, joining techniques requiring a considerable amount of deformation are not convenient for fibre reinforced MMCs. Consequently, possibilities for success are limited, probably to solid state pressure bonding and friction welding. However, friction welding might be successfully applied in the case of an MMC containing short fibres or oxide particles as the strengthening phase. Although diffusion bonding can minimise plastic deformation during bonding, significant plastic deformation at the faying interface is still required to achieve complete atomic joining. Even this deformation might cause damage and destruction of the fibre, leading to degradation of the original properties of the MMC at the joint.

The possible interfaces developing at the joining interface of the MMC are fibre/fibre, fibre/matrix, and matrix/matrix. It is rather easy to generate reactions at the matrix/matrix interface, while difficulties may sometimes be experienced at the fibre/matrix interface. However, it is quite difficult to obtain real bonding at the fibre/fibre interface. As a result of discontinuity, a decrease in joint efficiency will inevitably occur at the joint interface. The decrease in joint strength in the case of MMCs strengthened with long fibres is a considerable problem.

Although diffusion bonding is an essential technique for the primary forming of MMCs, attempts to use the method for secondary processing have not produced satisfactory results. All the attempts to join aluminium matrix MMCs particularly yielded poor joint efficiency. The large deformation and/or high temperature required to bond aluminium or aluminium alloy matrix MMCs, to eliminate the effect of refractory oxide film covering the surface (an effective diffusion barrier), is responsible for the poor joint properties. Nevertheless, diffusion bonding of aluminium based MMCs is still possible, in particular by using appropriate interlayers, which can result in TLP bonding.

Several attempts have been made to improve the joinability of aluminium matrix MMCs. For example, silver plating of the joint surface has been demonstrated as an effective way of improving joint strength. The diffusion bonding of a 20%B/7093 butt joint with a doubler of the same material failed by shear at a tensile stress level in the base material of 240 MPa.¹⁸⁵ The effect of surface finishing of the joint surface on the joint strength was also investigated and electrolytical etching of the joining surface was reported to yield an increase in the joint strength in $Al_2O_3/Al.^{186}$

Particular success has been reported with 8090/SiC MMC¹⁸⁷ and this may in part be caused by the effect of lithium in destabilising the alumina layer. Use of insert metals that form liquid during bonding by reacting with aluminium matrix MMCs has also been investigated.¹⁸⁶ Among the insert metals investigated for aluminium matrix MMCs, silver and copper, which form eutectic by reacting with aluminium at the bonding temperature, are effective. With a liquidus temperature lower than that of aluminium, Al-Si alloy can also be used as an insert metal, although the mechanical properties are generally poor. It should also be noted that diffusion bonding temperatures are usually at or above solution treatment temperatures, and a full post-heat treatment is required to generate optimum properties. Some aluminium MMCs have insufficient hot strength to be able to withstand diffusion bonding without excessive deformation, which is a further limitation on the flexibility of the process. Nevertheless, diffusion bonding of aluminium matrix MMCs is possible, but not with all alloys. There is, however, considerable scope for optimising interlayers to promote TLP bonding.

Resistance welding has been applied mainly to aluminium matrix MMCs. The heating time in resistance welding can be reduced and hence the reaction between the matrix and the fibre can be minimised. Moreover, void or crack formation can normally be prevented by the pressure during welding. Consequently, if the welding conditions are properly selected, sufficiently high joining strength can be obtained with resistance welding. For example, a resistance welded joint of B/6061 fractured at 90% strength of the base material in a shear test.¹⁸⁸

It has been reported that capacitor discharge welding can be used for joining of SiC/Al (40 vol.-%SiC particulate reinforced aluminium).¹⁸⁹ Since this welding technique involves an extremely short heating time of 0.1 μ s, the reaction between the matrix and the fibre can practically be avoided. Microstructural observations have revealed that the joints produced with this technique include neither defects, such as cracks and voids, nor a reaction layer between the matrix and the fibre. Although the properties of these joints have not been investigated, this welding technique seems to have great potential for joining MMCs.

In addition to the problems described above, fusion welding of MMCs involves:

- (i) decomposition or dissolution of fibres at high temperature during welding
- (ii) occurrence of weld cracks and voids owing to poor wettability between the fibre and the molten metal.

This does not mean that the appropriate selection of welding methods and parameters would not make feasible application of fusion welding to certain kinds of MMC. For instance, an aluminium based alumina particulate reinforced MMC, namely AA 6061 aluminium alloy matrix containing 20 vol.-%Al₂O₃ particles, was successfully welded using the gas metal arc (GMA) process with conventional 1.6 mm diameter ER 5356 filler alloy.¹⁹⁰ It was demonstrated that alumina does not react with molten aluminium during GMA welding. However, a post-weld heat treatment (PWHT) was found to be necessary to achieve maximum mechanical properties. It has also been reported that low heat input welding with a wide groove enables 1039/20%B to be joined successfully,¹⁸⁵ although a filler having good wettability with the fibres is necessary to ensure a sound joint free from weld defects such as cracks and voids.

Another experimental result has demonstrated that it is possible to successfully join SiC/6061 composite by GTA welding with the use of 4043 alloy filler.¹⁹¹ A C/Al composite can also be joined by GTA welding with the use of 4043 alloy filler.192 However, an SiC/Al joint made by GTA welding without filler exhibited lots of voids and cracks owing to gas absorbed in the base material and poor fluidity of the weld metal. In order to minimise voids and cracks, degassing of materials under vacuum before welding and use of filler metal are recommended.¹⁹¹ In other GTA and GMA weldability work on a 6061 aluminium matrix composite reinforced with 25 vol.-%B₄C particles $(6061/25 \text{ vol.-}\%B_4C_p)$, no decrease of fracture strength as a result of GTA welding was observed.¹⁹³ It was also reported that the GMA process produced better quality welds than the GTA process and that preheating at 93°C resulted in a slower cooling rate and, thus, less porosity in the GMA weld.

As pointed out above, friction welding might be successfully applied in the case of an MMC containing short fibres or oxide particles as the strengthening phase. Several variants of the technique have been used to join these materials, such as both inertia and continuous drive rotary friction welding and linear friction welding.¹⁹⁴ For instance, Bollat¹⁹⁵ demonstrated that SiC fibre reinforced aluminium could be successfully joined to an aluminium alloy by the inertia welding process. Inertia-friction welding was also conducted on a 6061 aluminium matrix MMC reinforced with 15 vol.-%Al₂O₃ particles in the T6 temper condition.¹⁹⁶ It was shown that the similar and dissimilar welds were sound and free from any discontinuities. However, the joint properties were evaluated using simple transverse tensile and bending tests and the results showed limited ductility of the weld zones and that failure locations were at or near the weld interface. A more recent investigation has also demonstrated that continuous drive friction welding of an SiC particulate reinforced aluminium matrix MMC is feasible.¹⁸¹ A decrease in the heat affected zone (HAZ) strength as a result of dissolution of hardening precipitates within the aluminium matrix was observed, although full HAZ strength recovery was achieved by use of an appropriate PWHT, which involved solution heat treatment at 535°C followed by artificially aging at 160°C for 10 h, as shown in Fig. 1.

Although a sound weld metal free from defects such as voids or cracks might be obtained by the above mentioned GMA and GTA welding procedures, the joint properties of MMCs made by fusion welding rarely match those of fibre reinforced MMC base mateial. Joint properties are governed by the weld metal not strengthened with fibres. Moreover, welding studies conducted on SiC/Al particulate reinforced composite indicate a problem of loss of SiC reinforcement in the fusion zone, with the consequent formation of Al_4C_3 rendering the fusion zone brittle.^{189,191,197} Figure 3 illustrates the sharp increase in the hardness owing to formation of brittle Al_4C_3 .¹⁹⁷ The welding of SiC reinforced aluminium matrix MMC must, therefore, be carried out under conditions of minimal superheating to prevent formation of brittle Al_4C_3 . Recent investigation has shown that EB welding is a possible process for improving joint properties of MMCs, because the fibre free area is rather narrow. However, the joint efficiency of a 6061/50 vol.-%B joint made by EB welding with Al-Si filler remained as low as 19% in the case of a squire groove butt joint made with one weld from each side.¹⁹⁸

A study by Katayama¹⁹⁹ showed that laser beam welding performed on 15 vol.-%SiC_p/2024 at a low heat input did not produce brittle Al₄C₃ and did not alter the distribution of SiC particulates in the fusion zone. Another investigation of laser welding used both a continuous wave (CW) CO₂ laser and a pulsed Nd-YAG laser at various travel speeds



a as welded, travel speed 2 m min⁻¹, Ar flowrate 10 L min⁻¹; *b* after low temperature aging, travel speed 1 m min⁻¹, aged for 24 h at 400°F (200°C)

3 Variation of hardness in LB welding SiC/2024 (Ref. 197)

and energy inputs to weld an SiC particulate reinforced aluminium matrix MMC (A356 aluminium alloy matrix reinforced with 15 vol.-%SiC particles).²⁰⁰ The results demonstrated that use of laser processing at low energy inputs and short thermal cycles precluded dissolution of the SiC reinforcement and subsequent formation of the deleterious Al_4C_3 phase. It was also observed that microstructural phases produced in pulsed Nd-YAG welds were similar to those produced in CW CO₂ welds except that they were generally finer.

In another study, the feasibility of laser welding of continuous fibre reinforced aluminium composites was investigated.²⁰¹ The results suggested that a critical output power existed which determined the transition between welding and cutting.

General comments on MMCs

While efficient production and shaping processes for MMCs have been developed, little fundamental research has been accomplished on effective joining methods of MMC materials. Although the joint efficiencies reported are improving, they still remain below the level to fully exlpoit the potential of these materials. It is now widely accepted that thermal joining processes are more promising for achieving better joint strength levels, particularly by shortening the thermal cycle and, hence, the time available for reactions responsible for degradation of the joint strength. Moreover, further work is needed as there is inadequate knowledge of the metallurgical phenomena occurring during joining of these materials and also very few data are available on the mechanical properties of the joints or their performance in high temperature, high wear, or environmentally aggressive conditions.

JOINING OF OTHER ADVANCED MATERIALS Nickel base superalloys

Superalloys, which are also referred to as high temperature alloys, are used at temperatures of 540°C and above as a result of their excellent high temperature properties. Outstanding properties of these alloys include:

- (i) high strength at elevated temperatures
- (ii) resistance to environmental attack, including nitridation, carbonisation, oxidation, and sulphidation
- (iii) excellent resistance to creep, creep rupture strength, toughness, and metallurgical stability
- (iv) useful thermal expansion characteristics
- (v) resistance to thermal fatigue and corrosion.

Although there is intensive research and development activity to replace nickel base superalloys with lighter materials, they are still widely used for many applications in aerospace industry. Nickel base superalloys possess superior high temperature properties, but they have the disadvantage of relatively high density. Superalloys comprise solid solution strengthened and precipitation hardened nickel, iron, and cobalt base alloys; the present review only addresses nickel base superalloys.

- Superalloys can be divided into two main groups:
- (i) solid solution strengthened alloys
- (ii) precipitation hardened alloys.

Both the solid solution and the precipitation hardening groups of alloys are typically austenitic (fcc phase, often referred to as γ). Cobalt, iron, chromium, molybdenum, tungsten, and tantalum are all solid solution hardeners in nickel base alloys, whereas chromium and molybdenum are the main solid solution hardeners in iron base superalloys. The atomic diameters of these elements differ from that of nickel by 1–13%. Lattice expansion owing to atomic diameter oversize (size factor) can be related to the hardening observed. Above $0.6T_m$, which is the range of high temperature creep where T_m is the melting temperature, diffusional processes determine the deformation and slow diffusion elements such as molybdenum and tungsten are the most effective solid solution strengtheners in nickel base superalloys.

The precipitation hardening alloys contain a second phase, γ' -Ni₃(Al,Ti), which is produced by means of controlled heat treatment. Dispersion of this second phase is achieved by additions of aluminium, titanium, and niobium. Precipitation of γ' -Ni₃(Al,Ti), which is the principal strengthening phase in many nickel and Ni-Fe base superalloys, in a nickel matrix provides significant strengthening to the material. This unique intermetallic phase has an fcc structure similar to that of the matrix and a lattice constant having 1% or less mismatch with the lattice constant of the γ matrix. Such close matching results in low surface energy and, therefore, provides long term stability. The γ' phase imparts great stiffness to these alloys at room temperature and up to the upper limit of their aging temperature ranges. Strengthening of alloys by γ' precipitation is a function of γ' particle size. The hardness of the alloy increases with particle size growth, which is a function of temperature and time. The volume percentage of γ' precipitated is also important because high temperature strength increases with the amount of the phase present. The amount of γ' formed is a function of the hardener content of the alloy; aluminium, titanium, niobium, and tantalum are strong γ' formers. However, it should be remembered that these materials lose their strength rapidly when held at service temperatures above their aging ranges as a result of the agglomeration of their second phases. This condition, commonly referred to as overaging, results in relatively large γ' particles. In order to retard coarsening, alloying elements (which increase the volume percentage of γ') or high partitioning, slow diffusing elements such as niobium or tantalum (which provide for the formation of the desired precipitate) can be added.

Moreover, the γ' phase can transform to other Ni₃X precipitates if the alloy is supersaturated in titanium, niobium, or tantalum. Titanium rich metastable γ' can transform to Ni₃Ti or an hcp η phase. Formation of η phase, which is found in iron, cobalt, and nickel base superalloys with high titanium/aluminium ratios after extended exposure, can alter mechanical properties. Excessive niobium content results in metastable η transforming to γ'' (a body centred tetragonal phase which is the principal strengthening phase in Inconel 718) and ultimately to the equilibrium Ni₃Nb phase (an orthorhombic phase observed in overaged Inconel 718). Both γ' and γ'' can be present at peak hardness, whereas transformation to the coarse, elongated orthorhombic Ni₃Nb results in a decrease in hardness. The types of phases precipitated are functions of alloy chemistry and heat treatments before service or the temperature-time exposure in application.

Carbide strengthening is another strengthening mechanism in superalloys. Nickel is not a carbide former itself. Carbides are formed by reaction of carbon with alloying elements in nickel; the carbides most frequently found in nickel base alloys are MC, M_6C , M_7C_3 , and $M_{23}C_6$ (where M is the metallic carbide forming element or elements).

Normally, MC is a large blocky carbide, random in distribution, and generally not desired. The M_6C carbides are also blocky; formed in grain boundaries they can be used to control grain size, but precipitated in a Widmanstätten pattern throughout the grain they can reduce ductility and rupture life. Predominantly Cr_7C_3 , M_7C_3 carbides form intergranularly and are beneficial if precipitated as discrete particles. They can cause embrittlement if they agglomerate, forming continuous grain boundary films. This condition will occur over an extended period of time at high temperatures. The $M_{23}C_6$ carbides usually tend to precipitate as discrete particles at grain boundaries and are influential in determining the mechanical properties of nickel base alloys, e.g. enhancing rupture properties.

Desired carbide structures and morphologies can be attained by heat treatment. The alloy chemistry, its prior processing history, and the heat treatment given to the material all influence carbide precipitation and ultimately performance of the alloy.

Physical metallurgy of nickel base superalloys

Most solid solution (non-precipitation hardening) nonferrous alloys exhibit sufficient ductility in the as welded condition. These materials, e.g. Inconel 625, can readily be joined using conventional welding processes. The GTA welding procedure for nickel based superalloys is already established and commonly used in industry; there is more extensive experience of GTA welding Inconel 625 than titanium and such alloys usually do not require pre-weld or post-weld heat treatments. Resistance heated diffusion bonding is, however, unsuitable for joining Inconel 625 plates owing to the rapid formation of an oxide film at the bonding interface.

The potential problem in welding these alloys is grain size. As grain size of material increases, weldability decreases, although larger grain sizes are generally required for better creep properties. Therefore, a minimum grain size is specified for maximum creep resistance in some applications. The compromise normally implemented is to limit welding processes to those methods that use lower heat input such as LB and EB welding.

For precipitation hardenable alloys, the thermal cycle(s) introduced by the welding process can significantly alter the microstructure of the base material in the weld region, thus markedly influencing their properties. The heating cycle involved in welding can result in the solutionising of age hardening constituents (base metal degradation) which will reduce the strength of the material. Temperatures in regions immediately next to the fusion zone will reach and

exceed the recrystallisation temperature, resulting in grain growth and elimination of the retained strain energy from thermomechanical processing. Such effects can significantly alter the properties of the weldment. These alloys are usually welded in the annealed (or solution treated) condition and are subsequently aged to precipitate the second phase.

The physical metallurgy of precipitation hardenable nickel base superalloys that affects weldability is, however, centred on the precipitates used for age hardening (γ' and γ'') and those associated with solidification and solidification segregation (primarily carbides and Laves phases). There are two types of cracking encountered in welding these alloys, namely strain age cracking and liquation cracking. Age hardening causes problems associated with strain age (i.e. reheat) cracking, whereas carbides and Laves phases can initiate melting and lead to cracking of grain boundaries in the HAZ (liquation cracking).

Nickel base superalloys that are strengthened by γ' -Ni₃(Al,Ti) precipitates are more prone to strain age cracking than are those strengthened by γ'' -Ni₃(Al,Ti,Nb,Mo) precipitates because of the more sluggish reaction of the γ'' phase. Inconel 718 is strengthened by γ'' phase and may, therefore, not be extremely susceptible to strain age cracking. For example, alloy 718 which is hardened by the more sluggish reaction of the γ'' phase is not prone to strain age cracking. However, alloy 718 does have problems when welded using conventional GTA welding, including microcracking in the HAZ, very poor impact strength in the fusion zone, poor fusion in the root pass, and heat to heat variation in weld penetration.

Initially, most of the welding research associated with nickel base superalloys was carried out on wrought forms of alloys such as René 41, Inconel 718, Waspaloy, and Inconel X. Several of these findings for the wrought nickel base superalloys will also apply for the cast nickel base superalloys. Boucher *et al.*²⁰² first correlated the grain size of the wrought superalloys to HAZ cracking. They showed that grain sizes smaller than ASTM 6 would provide resistance against HAZ microcracks in alloys such as Waspaloy and 718, but the grain size of investment cast structures can not generally be controlled to such fine grain sizes. However, aspects such as tramp element contamination, constitutional liquation of carbides, and strain age cracking can be applied in some form to the cast welding metallurgy of the wrought compositions. The original work done by Pease²⁰³ regarding tramp elements such as sulphur, phosphorus, lead, etc. was excellent and still is a good guide to follow with respect to cast alloy tramp element limits. The maximum specification levels for the elements sulphur, phosphorus, oxygen, and nitrogen in cast alloy 718 are 0.005, 0.005, 0.002, and 0.0035 respectively. Kelly^{204,205} has evaluated the effect of the tramp elements on the susceptibility of cast alloy 718 and René 220C to HAZ microcracking. He reported that the maximum specification levels are adequate to prevent HAZ microcracking.

Strain age cracking was first investigated and understood for the wrought alloy René 41 by Berry and Hughes,^{206,207} giving results that can be applied to cast compositions. They demonstrated that strain age cracking was an aspect of the γ' aging reaction in the HAZ of weldments. This reaction might occur during welding or in subsequent heat treatment, but regardless of the timing. The γ' precipitation in the matrix results in strengthening of the matrix so that all the solidification strains of the weldment are transferred to the grain boundaries in the HAZ causing them to fail and form microcracks. Elements which adversely affect strain age cracking are carbon, sulphur, and boron. Carbides tend to be initiation sites, while both sulphur and boron were correlated with cracking by Berry and Hughes, but no mechanism was determined.

Prager and Shira²⁰⁸ proposed the plot in Fig. 4, which was modified by Kelly²⁰⁹ to include alloys such as René



4 Plot of weldability v. Al and Ti content based on strain age cracking susceptibility: after Prager and Shira;²⁰⁸ includes several Ni base superalloys added by Kelly²⁰⁹

220C, Inconel 939 and 909, and René 108, which demonstrates that the phenomenon of strain age cracking is caused by the total Ti + Al content. As illustrated in Fig. 4, both Waspaloy and René 41 fall just below the safe limit and, therefore, require controlled grain size and special heat treatment for welding. Several cast nickel base superalloys such as Inconel 100 and 713C are also included in Fig. 4. Both titanium and aluminium form γ' , the nickel base superalloy strengthening phase that forms rapidly at certain temperatures. If the total Ti + Al content is kept below 4 wt-%, strain age cracking problems will be avoided. The most effective means of preventing strain age cracking is to overage the material before welding or to use controlled heating and cooling cycles for welding and subsequent heat treatment.

Constitutional liquation of phases found in the HAZ of weldments is another phenomenon that was first proposed by Pepe and Savage²¹⁰ and later discovered to be active in the liquation reaction of carbides in nickel base superalloys by Duvall and Owczarski.²¹¹ Enrichment of niobium, titanium, and molybdenum have been reported by several investigators to cause liquation cracking in nickel base superalloys.^{211–216} This process occurs when phases that form during solidification, such as MC carbides and Laves phases, initiate melting in the HAZ during welding and spread along the grain boundaries. The melting, often referred to as liquation, occurs as a result of a reaction between the dissolving precipitate and the matrix, resulting in liquid grain boundary films in the HAZ, well away from the fusion zone. Such a location prevents backfilling and promotes liquation cracking. The fact that liquation from precipitates can extend well away from the fusion zone not only increases the chance of liquation cracking but also increases the potential length of the crack, accentuating the stress intensity of the crack. The liquid film is incapable of absorbing the strain produced as a result of the contraction of the liquid metal in the fusion zone during solidification and therefore causes the grain boundaries it occupies to separate and form HAZ microcracks. Such cracking may be termed liquation cracking, hot cracking, or microfissuring.

Cieslak *et al.*^{217,218} investigated the solidification of alloy 625, a nickel base alloy containing relatively little niobium and practically no titanium or aluminium, and reported

that this alloy, even with reduced silicon content but with just 3.5 wt-%Nb, continued to have Ni₂Nb Laves phase as its terminal solidification phase. This is an indication that the as cast structures of any nickel base superalloy containing niobium would be expected to have the Laves phase present in the HAZ of any weldment, which leads to a eutectic reaction with the nickel matrix that can cause HAZ microcracks. Furthermore, before formation of the Laves phase, alloy 625 formed NbC, which would provide a means of generating a carbide constitutional liquation reaction in the absence of the Laves phase. In fact, MC carbide (NbC) and Laves phase (Ni2Nb) were reported to form in the HAZ of Inconel 625 weldments.²¹⁸ Wlodek²¹⁹ showed that silicon stabilises this niobium rich Laves phase. Cieslak²²⁰ also reported that for Inconel 625, low niobium content heats will be less susceptible to hot cracking than high niobium content heats. It was also reported that carbon alloying additions to the niobium bearing alloys promoted the formation of the γ -MC (NbC) constituent at the expense of the γ Laves constituent. Silicon, however, promoted formation of the γ Laves constituent with or without the presence of carbon and promoted formation of a γ -M₆C constituent at low carbon levels.

Additionally, Knorovsky *et al.*²²¹ have reported that both the Laves phase and NbC formed in alloy 718 on solidification, while Baeslack *et al.*²²² demonstrated that a Ni₂Ta Laves phase can form in tantalum containing alloys. Duvall *et al.*^{211,212} identified carbides in several alloys and Pease²⁰³ proposed that zirconium will form low melting phases in nickel base alloys. In the alloys containing a total of greater than 4 wt-%Ti + Al, the normal strain age cracking problem reported for wrought alloys would also be a concern even if liquation cracking is avoided.

Kelly^{204,209} has reported that boron is the most influential element in the formation of HAZ microcracks. However, it can not be dispensed with since it is required in nickel base superalloys for intermediate temperature strength and ductility. It appears that boron by itself does not rupture the grain boundaries of the HAZ; however, it is directly involved in an as yet undefined reaction that prepares the HAZ grain boundaries to be wet by any low melting constituent that is present. Until the role of boron in the grain boundary reactions is well understood, it can only be cautiously suggested that the lower the boron content the better the weldability.

Baeslack *et al.*²²² demonstrated that by increasing the eutectic reaction temperature of the alloy system the tendency for HAZ microcracking will be reduced. For instance, by replacing niobium with tantalum in alloy 718 the eutectic reaction temperature increases from 1185 to 1225° C.

Other factors, such as large grain size, ^{202,215,223-225} heat treatment, and high impurity levels^{203–205,223,226–232} may also promote HAZ cracking in superalloys. It is also worth noting that HAZ microcracking is apparently a function of both the boron at the grain boundaries and the lowest liquating temperature phase reaction present in the system. In the absence of a Laves phase, carbide constitutional liquation would determine HAZ microcracking. Kelly²⁰⁴ has reported that liquation of carbides in the HAZ of nickel base superalloy weldments occurs at higher temperatures than the Laves phase eutectic reaction in alloy 718 and it is likely that this would be valid for other Laves containing alloys.

The welding metallurgy of cast nickel base superalloys is further complicated by the fact that most castings are heat treated in some manner before welding to relieve casting stresses. This might lead to the formation of low melting temperature grain boundary films. For this reason, it is a good rule to avoid thermal cycles that might lead to liquation, i.e. application of a lower thermal cycle just sufficient to relieve casting stresses, or to leave strain age prone alloys in the solution condition for welding.

Joining of nickel base superalloys

The GTA welding procedure for nickel base alloys is already established and commonly used in industry. There is also more extensive experience of GTA welding Inconel 625 than titanium. Maak and Tinkler²³³ reported that EB welding is a feasible process for joining thick plates (>12.7 mm) of Inconel 625, which is a solid solution alloy. It should be noted that there is no published LB weldability study of Inconel 625. Although it has not been well demonstrated, there is no indication that solid solution alloys may present any major difficulty in welding by the LB process. This therefore needs to be investigated.

Welding of precipitation hardenable superalloys is, however, much more complicated, particularly that of crack sensitive alloys. In a study of the weldability of Haynes alloy 242 (a Ni–Mo–Cr superalloy), it was reported²³⁴ that the hot cracking resistance of alloy 242 based on varestraint testing is better than that of Inconel 718 and slightly less than that of Carpenter Custom Age 625 Plus. However, when compared with other alloys derived from the Ni–Mo–Cr system, alloy 242 appears to have one of the lowest hot cracking resistances.

Mills^{235,236} has investigated the effect of different PWHT procedures on the mechanical properties of Inconel 718 weldments, including tensile strength, ductility, and fracture toughness. Two different heat treatment schedules, conventional heat treatment (consisting of annealing for 1 h at 954°C, air cooling to room temperature, followed by aging for 8 h at 718°C, furnace cooling to 621°C, holding for 18 h, and air cooling to room temperature) and modified heat treatment (comprising solution treating for 1 h at 1093°C, cooling to 718°C at 55 K h⁻¹, holding for 4 h, cooling to 621° C at 55 K h⁻¹, holding for 16 h, and air cooling to room temperature) have been used as PWHT. Both heat treatment schedules resulted in similar tensile properties comparable with base metal strength and ductility values, whereas the as welded material displayed lower strength levels and higher ductilities relative to those observed in the heat treated welds. However, both heat treatment schedules decreased the fracture toughness $K_{\rm lc}$ of the as welded material ($K_{\rm lc} = 160$, 183, and 228 kJ m⁻² at room temperature, 427°C, and 538°C respectively). The modified heat treatment produced higher fracture toughness ($K_{Ic} =$ 141, 117, and 100 kJ m⁻² at room temperature, 427°C, and 538°C respectively) than the conventional heat treatment ($K_{\rm Ic} = 21$, 32, and 32 kJ m⁻² at room temperature, 427°C, and 538°C respectively).

David *et al.*^{237,238} have recently studied the weldability of single crystal nickel base superalloys (PWA 1480). They reported that these alloys exhibit a propensity for cracking and that crack free welds could be made over a very narrow range of welding conditions with a 500°C preheat. Furthermore, the loss of single crystal nature in the weld zone was encountered, which may degrade the weld properties.

Laser welding (using a pulsed Nd-YAG laser) of some nickel base superalloys, namely Inconel 600, 718, and 82, has also been investigated.²³⁹ No cracking was observed in Inconel 82, whereas Inconel 600 and 718 did crack. Busch and Kunze²⁴⁰ have, however, reported crack free weldments of Inconel 718 by both LB and EB welding processes. The type of laser beam was not specified.

The γ carbide and γ Laves eutectics can be suppressed by LB and EB welding processes, owing to rapid solidification. This was recently demonstrated for Inconel 718 by Gobbi *et al.*^{241,242}

Heat affected zone cracking can also be minimised by optimising welding parameters to reduce the dimension of the bead nailhead. The use of Nd-YAG pulsed lasers could prevent microfissuring easily by producing better bead profiles.²⁴¹ Power beam (LB and EB) welding processes

can be successfully applied in conjunction with appropriate metallurgical measures to avoid liquation cracking.

General comments on nickel base superalloys

Some nickel base alloys, particularly solid solution alloys, can be successfully welded without difficulty and precipitation hardenable alloys are prone to strain age cracking or HAZ liquation. Many factors, such as composition, grain size, prior heat treatments, welding parameters, and welding processes, can play an important role in determining the weldability of these alloys. For instance, applying a conventional arc welding process (high heat input), particularly to a coarse grain material, can frequently result in HAZ microand macrofissures caused by eutectic formation. The weldability of precipitation hardenable alloys is limited by either strain age cracking (not a problem when the strengthening phase is γ'') or HAZ liquation owing to the presence of carbide and/or Laves phases.

Conventional welding processes have been shown to produce crack free sound joints in several superalloys. Some successes in power beam welding of superalloys have been reported. However, there is still a need for further research to obtain a full understanding of the relationship between the welding parameters and the performance of the power beam joints of these alloys. Evaluation of the mechanical properties of the power beam welds has also not yet been fully completed for these alloys. Limited data on the performance of power beam superalloy joints have been reported.

Al-Li alloys

Aluminium alloys can be divided into two main groups, namely non-heat treatable and heat treatable. Non-heat treatable wrought aluminium alloys include 1xxx alloys, 3xxx alloys, 4xxx alloys containing only silicon, and 5xxx alloys. These alloys can only be strengthened by solid solution hardening, cold working, and grain refinement. From the standpoint of engineering applications, the most important non-heat treatable aluminium alloys are the 5xxx series alloys. They are used in applications where moderate strength levels are required, such as in the automotive industry, for pressure vessels, and in components for marine and cryogenic service. The relatively high strength of these alloys is provided by solid solution hardening from magnesium. More importantly, magnesium promotes work hardening by lowering the stacking fault energy, thus reducing dynamic recovery.

Heat treatable aluminium alloys exhibit low density and corrosion resistance and good strength and toughness levels. Consequently, they are used in a wide range of engineering applications, which include aerospace, transport, shipbuilding, tankage, and piping industries. The commercial wrought heat treatable aluminium alloys include the 2xxx (Al-Cu or Al-Cu-Mg), 6xxx (Al-Mg-Si), and 7xxx (Al-Zn-Mg) series alloys.

The high strength of heat treatable aluminium alloys is provided by the finely dispersed precipitates that form during aging heat treatments. For example, the precipitation sequence in the Al-Cu system starts with the saturated solid solution developing solute clusters referred to as GP (copper rich) zones. This is followed by the formation of transitional (non-equilibrium or metastable) precipitates of approximate composition Al₂Cu (θ' and θ''). The final structure consists of equilibrium precipitates, e.g. θ phase Al₂Cu, which do not contribute to precipitation strengthening. The GP zones are more resistant to movement of dislocations through the lattice and hence are the most effective strengtheners. In practice, hardening in high strength aluminium alloys is achieved by a combination of GP zones and relatively widely spaced, semicoherent or incoherent precipitates. Some alloys in which diffusion is faster age harden at room temperature (natural aging), whereas heating to a temperature below the GP zone solvus line is required for slow diffusing alloys (artificial aging).

Most of the heat treatable alloys, however, contain combinations of magnesium with one or more elements, such as copper, silicon, and zinc. Characteristically, even small additions of magnesium in conjunction with these elements promote precipitation hardening, e.g. in the Al-Cu-Mg system (2xxx series) where strengthening is provided by the Al₂Cu phase and the addition of magnesium intensifies precipitation. Alloys in the 6xxx (Al-Mg-Si) series contain silicon and magnesium approximately in the proportions required for formulation of Mg₂Si. The dispersion of particles of the equilibrium η phase (MgZn₂) and η' precipitates are the main strengtheners in the 7xxx series (Al-Zn-Mg) alloys.

Minor additions of silver (~0.1 at.-%) could enhance the response to age hardening of all aluminium alloys containing magnesium, namely the commercial Al-Mg, Al-Cu-Mg, and Al-Zn-Mg-Cu systems. In all these alloys, silver promotes nucleation of a finely dispersed intermediate precipitate. For instance, ~0.1 at.-%Ag addition to Al-Cu-Mg alloys can stimulate an improved response to age hardening over a wide range of compositions. Recent attention has been directed towards alloys with high copper/magnesium ratios in which silver promotes nucleation of a new precipitate, designated Ω , that forms as finely dispersed plates in the {111} matrix planes. The Ω precipitate is generally believed to be an orthorhombic form of the equilibrium phase θ phase (Al₂Cu) that precipitates in the {100} planes in the binary Al-Cu system.²⁴³

The Al-Li alloys are also heat treatable and capable of providing high strength with heat treatment through the precipitation of T_1 -Al₂CuLi, δ' -Al₃Li, or S'-Al₂CuMg depending on the alloy composition. The most potent strengthening phase in Al-Li alloys is T_1 -Al₂CuLi, which is favoured at higher copper/magnesium ratios, i.e. in alloy 2090. Furthermore, zirconium forms fine Al₃Zr dispersoids which have the primary role of assisting control of recrystallisation and grain size during casting and hot working.

A new Al–Li based alloy, Weldalite 049, developed in the 1980s, was designed to replace alloys 2219 and 2014.^{244–246} Weldalite 049 is an Al–(4·0–6·3)Cu–1·3Li–0·4Ag–0·14Zr alloy that has extremely high strength in several tempers because, in part, of the enhanced precipitation behaviour caused by the Ag + Mg additions. It is significantly stronger than other aluminium alloys and has a typical longitudinal T8 tensile yield strength (mean value of more than 22 heats) of over 690 MPa from $10\cdot24 \times 2\cdot3$ cm extruded 'plate'. This exceptionally high strength in the T8 temper is a result of an ultrafine distribution of T₁-Al₂CuLi platelets, with no δ' -Al₃Li present.

Physical metallurgy of aluminium alloys

The present review discusses only physical metallurgy of heat treatable aluminium alloys and welding of Al-Li alloys. A majority of heat treatable aluminium alloys can readily be welded. However, these alloys possess certain characteristics inherent to all aluminium alloys, such as a tenacious oxide layer, high thermal conductivity, a high coefficient of thermal expansion, high reflectivity, solidification shrinkage almost twice that of ferrous alloys, relatively wide solidification temperature ranges, a tendency to form low melting constituents, and high solubility of hydrogen when in the molten state. Therefore, some difficulties, namely crack sensitivity, propensity for porosity, and strength loss in the weld metal owing to dissolution of precipitates, can be encountered in welding of these alloys. Furthermore, the high reflectivity of aluminium alloys leads to difficulties in laser beam welding. These aspects must be considered before successful welding can be employed.

Crack sensitivity. Aluminium alloys are sensitive to weld metal cracking as a result of their large solidification temperature range, high coefficient of thermal expansion, and large solidification shrinkage. The weld crack sensitivity of heat treatable aluminium alloys is particularly of prime concern because of the greater amounts of alloying additions present in these alloys, which also exhibit a tendency to form low melting constituents.

Weld cracking in aluminium alloys may be classified into two main groups based on the mechanism responsible for cracking and the crack location being either solidification cracking or liquation cracking. Solidification cracking occurs within the weld fusion zone and is caused by solidification shrinkage. Liquation cracks, however, occur next to the fusion zone and are caused by the formation of low melting constituents. Liquation cracking takes place in precipitation hardenable alloys as a result of the relatively large amount of alloying additions available to form eutectic phases. These constituents have low melting points and so liquate (melt) during welding, accompanied by tears pro-vided that sufficient stress is present.²⁴⁷⁻²⁵⁰ Higher heat input widens the partially melted region and makes it more prone to tearing. Thus, liquation cracking is not expected to be of prime concern in power beam welding of aluminium alloys, which is a low heat input process. Consequently, only solidification cracking is discussed below.

Solidification cracking (hot tearing). Solidification cracking or hot tearing is cracking in the fusion zone caused by the inability of the liquid to support the strain imposed by solidification shrinkage and thermal stresses.^{247,251–253} The degree of restraint of welded assemblies plays an important role in crack sensitivity by increasing the external stress on the solidifying weld. Solidification cracking takes place within the weld fusion zone and is influenced by weld metal composition, welding parameters, and the aspect ratio of the weld bead.

The crack susceptibility (castability) index for Al-Cu-Mg ternary alloys generated by Pumphrey and Moore²⁵⁴ in 1948 is normally used to predict the weldability of aluminium alloys. Crack sensitivity curves of various binary aluminium systems have also been experimentally determined $^{255-260}$ and can be used to predict the relative crack sensitivity behaviour of complex alloy systems. The hot cracking susceptibility of any aluminium alloy in general appears to be related to its copper and magnesium contents and may involve any Al-Cu-Mg eutectic.²⁵³ Alloys with high copper and low magnesium contents, e.g. alloy 2219, are readily weldable, as are alloys with low copper and high magnesium contents, e.g. alloy 5083. However, several high strength, precipitation hardenable aluminium alloys rely on complex alloying additions (large amounts of both copper and magnesium) to develop the required properties. These additions dramatically increase weld crack sensitivity, the most notable examples being alloys 2024 (Al-4·4Cu-1·5Mg) and 7075 (Al-5·6Zn-2·5Mg-1·6Cu). Moreover, in complex alloy systems, minor alloying additions of copper or magnesium may dramatically increase crack sensitivity by widening the temperature range.^{254,261,262}

High heat inputs, such as high currents and slow welding speeds, contribute to weld solidification cracking.²⁶³ Another factor contributing to hot tearing is the weld metal grain size; weldability of aluminium alloys can be improved by refinement of the weld metal grain size.²⁶⁴ Weld metal grain sizes produced by power beam welding processes are usually fine as a result of the rapid cooling involved in these processes. Low heat input power welding processes may thus reduce weld crack sensitivity.

Čieslak and Fuerschbach²⁶⁵ studied the hot cracking susceptibility of alloy 6061 when welded with pulsed and

continuous wave Nd-YAG lasers. They reported that pulsed Nd-YAG laser welds are extremely susceptible to weld metal hot cracking while continuous Nd-YAG laser welds produce crack free joints. The susceptibility to hot cracking for pulsed lasers is attributed to thermally induced stress gradients from the pulsed energy input. Wittig *et al.*^{266,267} also observed that alloy 2024–T3 Nd-YAG pulsed laser welds contained severe centreline hot cracking while a continuous wave Nd-YAG laser produced sound welds. Kutsuna *et al.*²⁶⁸ have recently reported that crack free joints in alloy 2219 can be readily produced with crossflow type CO₂ lasers with a power of 4 kW at a welding speed of 2 mm min⁻¹.

Crack sensitivity of Al-Li alloys. Most Al-Li alloys are strengthened by precipitates such as T₁-Al₂CuLi, which does not require large quantities of both copper and magnesium. It is therefore predicted that Al-Li alloys with high and low magnesium contents, e.g. alloys 2094 and 2090, should not be prone to cracking. Conversely, alloy 2091, and Al-Li alloy with intermediate amounts of both copper and magnesium, may exhibit crack sensitivity. Alloy 2319 can be used as a filler to join both 2090 and Weldalite 049 and alloy 5356 can be used to join alloy 8090.

Only a few studies aimed at understanding comprehensively the technique and use of welding to join lightweight Al-Li alloys have been conducted. Cross et al.251 carried out studies to evaluate the weldability of high purity Al-Li binary alloys using the varestraint test to assess their susceptibility to hot tearing. They observed that the susceptibility of an alloy to hot tearing varies with lithium content and that the depth of penetration increases with increasing heat input, as is observed in other aluminium alloys.²⁵² They also found the depth of penetration to increase with lithium concentration. The increase in penetration was rationalised as being caused by the decrease in thermal conductivity that accompanies increasing lithium content. It was observed that the high purity binary Al-Li alloys are readily weldable and have good resistance to hot tearing. Namba and Sano²⁶⁹ also reported that the susceptibility of Al-4.7Mg-(0.3-1.3)Li alloy to hot cracking during GTA welding is similar to that of alloy 5083, which is quite low. Hot tearing is cracking in the weld zone caused by the inability of the liquid to support the strain imposed by solidification shrinkage.^{251,252}

Although Al-2.5Li-1.2Cu-0.7Mg-0.12Zr alloy 8090 was designed for mechanically fastened applications, several investigations have been performed to assess its fusion weldability. For example, Gittos²⁷⁰ performed GTA and GMA studies on alloy 8090 using alloy 5556A (a slight modification of 5556), 4043, 2319, and parent alloy fillers. To reduce weldment porosity, the plate surfaces were mechanically milled before welding and inert gas backing was used. No hot cracking was observed in the weldments made with 5556A filler, while hot cracking took place in the weldments made with the other three fillers. It was also reported that the inherent hot cracking susceptibility of alloy 8090 is similar to that of 6082, which is a hot cracking sensitive alloy. Gittos²⁷⁰ concluded that lithium is not a neutral addition to Al-Cu-Mg alloys in terms of influencing hot cracking because its presence changes the normal ranking of filler alloys compared with their performance in conventional Al-Cu-Mg alloys.

Edwards and Stoneham²⁷¹ also performed a weldability study on alloy 8090–T6 using the bead on plate (i.e. autogenous) Houldcroft test to assess hot cracking. It was reported that the susceptibility of alloy 8090 to hot cracking is similar to that of alloy 2014, a hot cracking sensitive alloy. Lippold²⁷² investigated the weldability of alloys 2090 and 8090 by varestraint and Houldcroft tests. He reported that varestraint test results indicated that alloys 2090 and

8090 are more susceptible to weld solidification cracking than other commercial aluminium alloys, such as 2219, 5083, and 6061. Furthermore, Houldcroft tests showed that alloy 2090 is more resistant to cracking than alloys 8090 and 2091. In a recent weldability study, a trans-varestraint test was used to compare Al-Li alloys with conventional aluminium alloys.²⁷³ Results indicated alloy 2219 to be the most weldable and alloys 2090 and 2094 to be the least weldable of those investigated. Skillingberg²⁷⁴ performed GTA and GMA weldability studies on alloy 8090 and reported that alloy 8090 was weldable using alloy 1100, 4043, 5356, and parent fillers, although significant hot cracking was observed using alloy 2319. Hot cracking was also observed in the weldments made with alloy 5356, but no cracking was observed in 2090 weldments produced using parent filler. However, extensive weld development programs^{246,275-277} have demonstrated that these Al-Li alloys can readily be welded in practice with little concern for cracking. For example, Gnanamuthu and Moores²⁷⁷ successfully demonstrated use of CW CO₂ laser welding to join 2.7 mm thick plates of alloy 8090 without using filler material. No hot tearing was encountered at speeds up to 95.2 mm s⁻¹ using 5 kW of laser beam power. Zacharia *et al.*²⁷⁸ carried out an investigation to deter-

mine the weldability of alloys 2090 and 2091 (Al-Li-Cu-Mg alloys) by EB welding on 12.5 mm thick plates. It was observed that these alloys are susceptible to fusion zone cracking. All the welds contained about 2-5% porosity in the fusion zone, but this is not considered to be excessive for EB welding of aluminium alloys. Results of hot ductility tests indicated that these alloys are not susceptible to HAZ cracking, although there was some evidence of liquation cracks in the HAZ extending from the fusion zone to the unaffected base metal. However, Martukanitz et al.279,280 conducted GTA, GMA, and EB welding studies on alloy 2090-T8E41 using alloy 4047, 4145, 2319, 4043, and 5356 fillers. Either mechanical milling of 0.08-0.23 mm from the surface or chemical milling of 0.04-0.15 mm from the surface in a 5 wt-%NaOH aqueous solution at 49°C, followed by desmutting in a chromic-sulphuric acid solution, was carried out before welding. Only alloy 5356 filler displayed sufficiently severe hot cracking susceptibility to be considered beyond the limit of commercial weldability. Sunwoo and Morris^{281,282} conducted EB welding studies on alloy 2090 and no hot cracking was reported.

Marsico and Kossowsky²⁸³ performed CW CO₂ laser beam welding of Al–Li–Cu alloy 2090 (1.25 mm thick plates). Although some porosity (about 2.3%) was observed in the fusion zone, no solidification cracking was detected. They reported that the alloy can be readily laser welded and the laser weldability of alloy 2090 is claimed to be better than that of most conventional aluminium alloys because of the lower thermal conductivity caused by lithium addition. Molian and Srivatsan^{284,285} also demonstrated that alloy 2090 (1.6 mm thick plates) can readily be welded by CW CO₂ laser (with a power level of 1.6 kW). They observed no cracking and only a few pores.

Weldalite 049 has also been welded by EB, GTA, GMA, and variable polarity plasma arc (VPPA) welding processes. No propensity to hot cracking has been observed in highly constrained weldments and experienced welders claim that the alloy welds similarly to alloy 2219.

Kramer *et al.*²⁴⁴ performed a study to compare the hot cracking susceptibility of Weldalite 049 containing different copper levels, alloy 2090, and conventional alloys 2219 and 2014. The Weldalite 049 variants contained nominal 6-2, 5-4, and 5-0 wt-%Cu contents, over which peak strength does not vary,²⁴⁵ but ductility and toughness increase with decreasing copper content. Alloy 2014 exhibited the greatest weld zone hot cracking susceptibility, followed by alloy 2090. The three Weldalite 049 variants displayed lower hot cracking susceptibility than alloy 2090, with susceptibility

generally decreasing with increasing copper content. Alloy 2219 had the lowest weld zone hot cracking susceptibility of the five alloys, while the four Al–Li alloys exhibited greater resistance to HAZ cracking than both alloys 2219 and 2014.

Webster and Bennett²⁸⁶ investigated the weldability of newly developed high strength-toughness XT alloys (with proprietary compositions) by the varestraint test and compared the results with those of commercial Al-Li alloys, including 8090 and 2090. It was reported that the TX alloys display a susceptibility to hot tearing lower than those of alloys 8090 and 2090 and comparable to the best conventional aluminium alloys.

Porosity. The porosity in aluminium weldments is mainly caused by hydrogen gas entrapped during solidification, which has appreciable solubility in molten aluminium and limited solubility in the solid state. Lithium containing aluminium alloys display a higher propensity for weld metal porosity,^{247,287} which has been observed in Al–Li power beam welds by many researchers.^{277,283,288–290} The observed high tendency to form porosity in Al–Li alloys has generally been attributed to higher initial amounts of hydrogen in the weld pool as a result of the increased hydrogen solubility in Al–Li alloys.^{247,291}

In a review of Soviet literature²⁹² it was reported that weld zone porosity was considered to be a key problem in welding alloy 01420 by many researchers.²⁹³⁻²⁹⁵ The high porosity was attributed to the surface reactivity of the alloy with ambient moisture resulting in the formation of lithium compounds such as LiO₂, LiOH, Li₂CO₃, and LiN. It was, however, also reported that a marked drop in porosity in alloy 01420 weldments could be achieved by removing 0.05 mm from the joint surface before welding. Pickens et al.²⁹⁶ studied the weldability of the Al-5Mg-2Li-0.1Zr 01420 variant using GTA welding with parent or alloy 5356 filler. Chemical milling of 0.25 mm in a 30%NaOH (6.5M) aqueous solution at 50-60°C, followed by rinsing in a 30%HNO₃ aqueous solution (4.8M), reduced porosity to acceptable levels. In addition, machining 0.13 mm from the surface weldment before welding was equally successful in reducing porosity. Namba and Sano²⁶⁹ also observed weld zone porosity in GTA weldments of Al-4.7Mg-(0.3-1.3)Li alloy, although it was reduced by trailing inert gas shielding, backing inert gas shielding, and scarfing of 0.2 mm from the surface before welding.

Gnanamuthu and Moores²⁷⁷ also reported that chemical milling with 5%NaOH and 10%HNO₃ before CW CO₂ laser welding decreased the level of porosity in alloy 8090 weldments (2.7 mm thick plates). In more recent work, Whitaker *et al.*²⁹⁰ also studied the autogenous laser weldability of alloy 8090 (1–4 mm sheet). Although porosity existed predominantly along the weld centreline, it was reported that the Al–Li based alloy 8090 can be welded using a CW CO₂ laser at speeds in the range 20–120 mm s⁻¹ with power of 1:5–3:8 kW.

Pre-weld joint preparation requires special care to avoid porosity. Thorough removal of mill scale at the weld joint, by means of dry machining or chemical milling, is required. Common practice in the welding of all Al–Li alloys to minimise porosity is to dry machine ~ 0.25 mm from mating surfaces as well as top and bottom plate surfaces, 25–50 mm from the centreline, on both sides of the joint. It is also considered good practice to weld within a few hours of machining and to degrease immediately before welding. It has also been shown that low levels of oxygen impurity (less than 300 ppm) must be maintained to produce acceptable welds in alloy 2090.²⁹⁷ In addition, when making through thickness welds, as in a keyhole mode, rearside shielding must be used to prevent excessive oxidation and thus porosity.



5 Hardness profiles across HAZ of GTA welds on 3.2 mm thick 2219–T87, 5456–H116, and 6061–T6 produced using constant heat input²⁴⁷

Katayama *et al.*²⁸⁹ evaluated the weldability of alloy 2090 using a CW CO₂ laser. They reported that the welding tendency of alloy 2090 was good, but some porosity in the upper part of fusion zones was observed. No mechanical data were reported. Skillingberg²⁷⁴ also reported that alloy 2090 weldments generally displayed higher porosity levels than 8090 weldments, particularly the 2090–5356 GMA weldment. However, Molian and Srivatsan^{284,285} reported that only a few pores occurred at the weld/base material interface of CW CO₂ laser beam weldments of alloy 2090 (1.6 mm thick plate) and the degree of porosity of the laser weld seemed to be unaffected by the depth of penetration.

Base metal HAZ degradation. For heat treatable aluminium alloys, base metal degradation in the HAZ is another problem encountered in arc welding. The HAZ degradation is distinguished by dissolution (2xxx alloys) or growth of precipitates (6xxx alloys).^{247,298–300} Although different in nature, the processes are all diffusion controlled and thermally activated.³⁰¹

A common method of determining the width and extent of the HAZ is to measure the hardness across it. Figure 5 shows hardness profiles for GTA welds of two common heat treatable alloys, 2219-T87 and 6061-T6, and, for comparison, for the work hardenable alloy 5456-H116.247 In the case of alloy 2219-T87, dissolution of precipitates occurs when temperature exceeds the solvus lines of respective precipitates. At positions close to the fusion zone, higher temperatures are experienced and greater dissolution of the strengthening phase occurs. Consequently, a continual decrease in strength from base metal towards the weld metal is typical of Al-Cu alloys (Fig. 5).247 For alloy 6061-T6, transformation of precipitates proceeds at the greatest rate when temperature is between 290 and 425°C.^{247,298,300,302} The minimum hardness value for alloy 6061-T6 in Fig. 5 represents the position in the HAZ which experiences temperatures within this range. This minimum is also referred to as the overaged zone for these alloys. At higher temperatures, which are experienced close to the fusion zone, particles are dissolved into solid solution and on cooling may precipitate as strengthening phases. This contributes to the slight increase in hardness that typically occurs near the fusion zone for 6xxx series alloys.

The extent of base metal degradation is determined by the welding process and parameters.^{247,303} Conventional arc welding processes involve the application of 10^3-10^4 W cm⁻² arc intensity and slow weld speeds



6 Hardness traverse of EB welded alloy 6061 joints

 $(<15 \text{ mm s}^{-1})$ which lead to excessive heat input into the base metal, thus resulting in a coarse weld microstructure and a wide HAZ. Weld metal degradation in the HAZ (HAZ degradation) of heat treatable alloys is therefore of prime concern in arc welding. In contrast to non-heat treatable alloys, heat treatable alloys can fortunately be post-weld heat treated to recover the strength of the HAZ.

The HAZ degradation is not expected to be a major problem with power beam weld methods such as EB and LB welding, which involve low heat input. These processes are promising for achieving higher strength in precipitation hardenable alloy weldments. However, loss of strengthening elements during welding may occur in the fusion zone. Cieslak and Fuerschbach²⁶⁵ reported magnesium evaporation during both continuous and pulsed Nd-YAG laser welding of alloy 6061, which is a function of welding parameters (particularly travel speed for continuous laser welds). The hardness of alloy 6061 welds was affected by the magnesium vaporisation such that base metal strengths could not be achieved by subsequent heat treatment to the T6 condition owing to a reduced ability of the alloy to precipitation harden as a result of a lower magnesium concentration. The reduced strength in the fusion zone caused by loss of strengthening elements can, however, be restored by using filler metal.

Figure 6 shows the hardness traverse of an EB welded alloy 6061 joint. In contrast to conventional fusion welding processes where the lowest hardness values are exhibited by the overaged region in the HAZ (Fig. 5), the lowest hardness values are obtained in the weld metal of power beam welds (Fig. 6). The extent of the weld zone, i.e. the fusion zone and HAZ, produced by power beam processes is much smaller than that produced by conventional fusion welding procedures. The local hardness minimum in the overaged region of the HAZ is also present in power beam welds, but it lies between the hardness values of the base metal and the fusion zone, which exhibits the absolute hardness minimum. This problem can be overcome by using proper welding parameters and filler wire.

Zacharia *et al.*²⁷⁸ observed softening in the HAZ of alloy 2090 EB welds (12.5 mm thick plate), but after post-weld aging (at 190°C for 16 h) the HAZ strength was restored to the same level as the base metal. No such softened zone was observed in the HAZ of alloy 2091 EB welds owing to the alloy's ability to age naturally at room temperature after welding. However, lithium depletion^{269,296} as well as loss of other elements in the fusion zone during welding may result in softening of the weldments in the weld region, rendering the fusion zone the weakest part. Marsico and Kossowsky,²⁸³ for example, reported that alloy 2090 weldments manufactured using CO₂ laser beam welding on 1.25 mm thick plate were not particularly strong, with the best transverse weldment strength being about 217 MPa.

Weld properties

When comparing weld properties of Al-Li alloys, the welding procedure used, autogenous welding or welding with filler, should be taken into account since these two procedures lead to different properties owing to different weld compositions.

Joint strength. As pointed out above, metallurgical transformations in the HAZ of heat treatable alloys during arc welding lead to degradation of strength in this region. Postweld solution treating and aging provides the greatest improvement in joint strength, but this practice involves use of water quenching which may result in distortion. Post-weld aging, which is carried out at lower temperatures, provides moderate recovery of joint strength and does not require water quenching.

An alternative method for increasing the weld strength of heat treatable alloys (i.e. alloys 2024, 2219, 6061, and 7020) is to weld them in the solution heat treated temper (T4) condition and age them after welding. To accomplish this effectively, a welding procedure that keeps the heat input relatively low and short in duration, such as EB or LB welding, should be employed.^{302,304,305}

The tensile properties of many wrought Al–Li alloys are exceptionally high compared with those of commercial aluminium alloys. However, the strength of the welded joints in these alloys is limited by the strength of the solidified weld metal. Even though joint strengths are higher than those obtained in conventional aluminium alloys, joint efficiencies of the order of 50% are to be expected for high strength Al–Li alloys.

Al-Mg-Li alloys. Kamada et al.³⁰⁶ reported that Al-4Mg-1Li alloy is fusion weldable with alloy 5356 filler, but joint properties were not given. Namba and Sano²⁶⁹ investigated the weldability of Al-4·7Mg-(0·3-1·3)Li by GTA and EB welding processes and reported that the lithium content of the weld bead was lower than that of the parent material and suggested that by using a lithium containing filler the lithium content of the weld zone is restored, thus increasing its strength.

Additionally, Saida and Matsumoto³⁰⁷ performed GMA welding of Al–Mg–Li alloys for a fusion reactor vessel application using alloy 5356 filler and reported that the weldments displayed high tensile strengths (up to 295 MPa, which is equal to that of the best Al–Mg alloys) and excellent elongations.

Alloy 01420. Mironenko et al.^{293,308} carried out a detailed investigation to evaluate the fusion weldability of alloy 01420 and found it to have 'good weldability'. They also found that filler materials containing various amounts of manganese, titanium, zirconium, or chromium provided stronger joints than those produced by the parent filler material. Fridlyander³⁰⁹ reported joint efficiencies (i.e. strength of the weld divided by strength of the parent material) of 70% without PWHT and as high as 99.5% after PWHT, i.e. resolutionising, air cooling, and artificial aging. Joint efficiencies of 80% for GTA welding were obtained for welds that received no PWHT using an Al-6·3Mg-0·5Mn-0·2Zr filler alloy. It was also reported that with re-solution heat treatment, i.e. quenching and artificial aging, joint efficiencies of 100% were obtained.²⁹² Pickens et al.²⁹⁶ also reported that alloy 01420 weldments

Pickens *et al.*²⁹⁶ also reported that alloy 01420 weldments exhibited tensile strengths as high as 272 MPa without PWHT and as high as 363 MPa with heat treatment. The joint efficiencies of up to 85% obtained were slightly lower than those reported in the Soviet literature. Softening of the weld bead with respect to the base materials was attributed to the effects of lithium depletion in the weld bead when using base material filler coupled with microstructural coarsening. Alloy 2090. Martukanitz *et al.*^{279,280} determined joint properties of alloy 2090 GTA and EB weldments. They reported that re-solution heat treating, quenching, and artificial aging of the GTA weldments produced a tensile strength of 386 MPa, but fracture was brittle with no measurable ductility. Autogenous EB weldments (as welded) displayed a tensile strength as high as 322 MPa and the same PWHT increased weldment tensile strength to 413 MPa, but no elongation was measured.

Skillingberg²⁷⁴ manufactured alloy 2090 GMA weldments using both parent and alloy 5356 filler and 8090 weldments using 4043 filler. It was reported that the strengths of the alloy 2090 weldments were lower than those of the 8090 weldments. The highest as welded strength for 2090 was 253 MPa, obtained using 5356 filler, whereas the highest for 8090 was 301 MPa, obtained using 4043 filler material.

Gaw³¹⁰ also reported a welding study on alloys 2090 and 8090, although the details of this study are included in a US Government report dissemination of which is restricted. As welded tensile properties of the weldments that were made using the thinner sheet were significantly stronger than those reported by Martukanitz *et al.*^{279,280} or Skillingberg.²⁷⁴ However, the weldments made using thicker materials were not as strong.

Sunwoo and Morris²⁸¹ also determined joint strengths of alloy 2090 GTA and EB weldments. Joints in the as welded condition exhibited low strength levels (less than 200 MPa) but good ductility (greater than 4% elongation) as a result of dissolution of strengthening precipitates. The highest strength levels of EB and GTA weldments reported were obtained by aging at 160°C for 32 h with 75% joint efficiency and at 190°C for 16 h with 65% joint efficiency respectively. It was also reported that aging at higher temperatures (230°C) leads to coarsening of precipitates and intermetallic constituents, deterioration of weldment strengths, and improvement of elongations. The best overall weldment properties are obtained in the solution treated and aged conditions, owing to a homogeneous distribution of strengthening precipitates.

In addition, Sunwoo and Morris²⁸² performed a study to characterise precipitation development in EB and GTA fusion zones of alloy 2090 and its effect on weldment properties. They reported that both EB and GTA fusion zones lack the strength and ductility of the base metal because of the low volume fraction and inhomogeneity of the strengthening precipitates. They also pointed out that the weld strength is primarily determined by the distribution of δ' -Al₂Li in the as welded condition. In the underaged condition, the primary strengthener is the δ' phase and even in the overaged condition the volume fraction of T₁-Al₂CuLi is still too low for it to be an effective strengthener. The EB welds display better overall properties than the GTA welds; joint efficiencies after aging at 170°C for 32 h (designated T8) were 75% and 55% respectively.

Marsico and Kossowsky²⁸⁸ also performed a laser weldability study on Al–Li alloy 2090. However, the autogenous weldments were not particularly strong and the best transverse weldment strength was about 217 MPa.

More recently, Molian and Srivatsan^{284,285} welded Al-Li-Cu alloy 2090 (1.6 mm thick plates) with a CW CO₂ laser. They obtained tensile joint efficiencies of 55% without any surface preparation work on the as received material. After mechanical milling of 0.127 mm from both surfaces, the tensile joint efficiencies increased to as high as 83%. These authors also reported that the hardness levels of the laser welds were consistent and unaffected by the presence of various microstructural features resulting from welding. They also compared joint efficiencies of EB and LB weldments with those from conventional processes and reported that LB welding provides the best results, owing to the small HAZ and fine structures in the weld zone. Alloy 8090. Edwards and Stoneham²⁷¹ fabricated alloy 8090–T6 autogenous EB weldments and GTA weldments using parent and Al–5Mg fillers. The highest alloy 8090 joint strengths (in GTA weldments), 311 MPa tensile strength in the as welded condition and 367 MPa after T6 temper of the weldment, were obtained using parent filler. The results are consistent with those of Gittos,²⁷⁰ but no mechanical data on the EB weldments were reported.

Wilner³¹¹ also investigated the weldability of alloy 8090, but the applications, welding parameters, and weldment properties investigated are proprietary. Nevertheless, he indicated that a weldment tensile strength design specification of 288 MPa is attainable with proper welding procedures using an Al-Mg filler, zirconium modified NG-61. The composition of the filler was not disclosed.

Skillingberg²⁷⁴ has produced GTA and GMA weldments on alloy 8090 using alloy 1100, 4043, 5356, and parent fillers. The highest as welded tensile strength of 301 MPa was that of the weldment made with alloy 4043 filler, whereas re-solution heat treatment, quenching, and artificial aging of the weldments made with parent filler produced a highest tensile strength of 447 MPa, which was 85% of the parent properties. The toughness of weldments was also assessed by Charpy impact testing. Toughness data were scattered, but the alloy 5356 filler produced the highest values.

Although it appears that alloy 8090 is susceptible to hot cracking, quality weldments can be made with proper care. The alloy produces high weldment strengths with alloy 4043, 5356, 5556, and parent fillers.

Weldalite 049. Weldalite 049 has been welded using the GTA process with ER 2319 and parent filler wire.²⁴⁵ Tensile

strengths of these weldments were reported to be as high as those of the strongest of welded conventional aluminium alloys, e.g. 310 MPa for alloy 7039 and 275 MPa for 2219. Crossweld tensile strengths of more than 340 MPa could be attained in the as welded condition.

A preliminary EB welding study by Pickens³¹² was performed in conjunction with welds at TWI, Cambridge, UK. It was reported that the as welded yield strengths were extremely high (e.g. 417 MPa) and that weldment tensile strengths were in the range 430–434 MPa.³¹³ After re-solution heat treatment, quenching, and artificial aging, an extremely high apparent yield strength of 527 MPa was obtained.

It was also reported by Pickens³¹² that both manual and automatic GTA weldments have been made on 10.2×0.6 or 10.2×0.95 cm extruded plate using alloy 2319 filler. After some weld parameter development, weldments having extremely low porosity and high strength were obtained. These strengths are significantly higher than those typically obtained with alloy 2219 using 2319 filler, the parent–filler combination the alloy was designed to replace. Commercial alloy 2219 VPPA welded with 2319 filler is the standard weldment used to fabricate the external fuel tank of the Space Shuttle. Shah *et al.*³¹⁴ and Hackett and MacFarlane³¹⁵ reported successful joining of Weldalite 049 by a VPPA process using alloy 2319 filler. The Weldalite 049–2319 weld was found to be stronger than the 2219–2319 weld.

Table 1 gives mean tensile properties for alloy 2090, alloy 8090, and Weldalite 049 arc weldments and for some conventional aluminium alloy arc weldments. Table 2 gives a comparison of various arc weldment and EB weldment

Table 1 Mean tensile properties of alloy 2090, alloy 8090, Weldalite 049, and some commercial Al alloy arc weldments²⁴⁷

	Filler	Walding	Weld			Ultimate tensile Yield Elongati strength, strength, ——— est temperature MPa MPa 25 mm	on, %		
Base metal	metal	process	mm	Heat treatment	Test temperature		strengtn, MPa	25 mm	50 mm
2219	2319	VPPA*	9.5	As welded	Room temperature	273	140	7.9	4.6
2219	049	VPPA*	9.5	As welded	Room temperature	283	154	7.1	4.7
2219	049	VPPA*	5.8	As welded	Room temperature	325	161	9.0	5.0
2219-T851	2319			As welded	Room temperature	300	185		5.0
2090	2319	VPPA*	13	As welded	Room temperature	252	156	8.6	4.7
2090	049	VPPA*	6.5	As welded	Room temperature	285	147	7.1	3.8
2090-T81	2319	GMA	13	As welded	Room temperature	232	204†		5.2
2090T4	2319	GMA	13	Post-weld aged	Room temperature	258	†±		0
2090–T81	2319	GMA	13	Post-weld solution treated and aged	Room temperature	386	†‡		0
7017–T6	Al-5Mg		•••	Naturally aged for 30 days	Room temperature	340	220		8.0
8090T6	Al	•••		As welded	Room temperature	165	137		5.0
8090-T6	Al-5Si			As welded	Room temperature	205	165		3.0
8090-T6	Al-5Mg			As welded	Room temperature	228	176		4·0
8090–T6	Al-5Mg		•••	As welded + T6 temper	Room temperature	302	245		4·0
8090T6	Al-5Mg (+Zr)		•••	As welded	Room temperature	235	183	•••	4 ∙0
8090-T6	8090			As welded	Room temperature	310	285		2.0
8090–T6	8090		•••	As welded + T6 temper	Room temperature	367	315		4·0
049	2319	VPPA*	9.5	As welded	Room temperature	274	248	1.5	1.0
049	049	VPPA*	9.5	As welded	Room temperature	315	249	1.5	1.5
049	049	VPPA*	9.5	Naturally aged for 800 h	Room temperature	372	290	3.0	
049	049	VPPA§	9.5	As welded	Room temperature	372	290	3.0	
049	049	VPPA§	9.5	As welded	175°C	287	188	5.4	
049	049	VPPA§	9.5	As welded	−195°C	413	360	1.9	
049	049	VPPA§	9.5	As welded	-253°C	505	427	1.7	

* Square butt weldments, all fractures occurred in the HAZ.

†Failure occurred through the weld fusion zone or at the interface.

‡ Failure occurred before reaching 0.2% offset, indicating nil elongation.

§ Weldments of extruded plate, 100×9.5 mm.

properties for some Al–Li alloys, taken from numerous sources. Alloy 2094 is found to have the highest joint strength (372 MPa ultimate strength after VPAA welding and 434 MPa ultimate strength after EB welding), which corresponds to a 38-61% improvement in joint efficiency over alloy 2219 weldments with 269 MPa ultimate strength.²⁷⁵

Ductility. Welds in heat treatable alloys do not generally exhibit as high ductility as those in non-heat treatable alloys. Moreover, PWHT usually decreases weld ductility further.

If the weld metal has significantly lower strength than the base metal (undermatching), most of the strain will be concentrated in the weld metal in a transverse tensile test. This is particularly the case for high strength aluminium alloys. Typically, joint ductility measured over a 50 mm gauge length will be 2-4% even though the actual weld metal ductility approaches 10-12%. Tensile elongations of Al-Li alloy arc weldments are given in Table 1.

No tensile ductility data have been reported for power beam welds of heat treatable alloys, although it is expected that tensile ductility will be low, especially for high strength aluminium alloys. However, this does not mean that the weld metal is intolerably brittle. The low ductility is a result of the very small thickness of the weld metal which undergoes plastic deformation in the transverse tensile test. The level of ductility achieved in such tests is strongly dependent on the size of the 'gauge length' used in the specimen. Acceptable tensile ductilities can be obtained provided that the strength of the joints can be improved.

General comments on fusion welding of Al–Li alloys

There has been great progress in the joining of Al–Li alloys. However, since Al–Li alloys are being considered for use in welded structures, further research efforts must be directed towards assessing the weldability of these alloys under a variety of conditions. Current fusion welding processes involve application of 10^3-10^4 W cm⁻² arc intensity and slow weld speeds of < 15 mm s⁻¹, which allow excessive heat input into the Al–Li alloys resulting in:

(i) a coarse weld microstructure

(ii) a wide HAZ

(iii) lithium depletion in the weld zone.

The above can lead to lower strength of welded joints and low resistance to stress corrosion cracking.

This problem can be overcome if the weld pool is solidified rapidly, thereby promoting a fine weld microstructure, minimal loss of elements from the weld pool, and

 Table 2
 Transverse properties for arc and EB welded Al-Li weldments in as welded condition²⁴⁷

Alloy and temper condition	Filler metal	Welding process	Yield strength, MPa	Ultimate tensile strength, MPa	Ref.
01420-Тб	5356	GTA	103	241	296
2090-Т8	2319	GTA	165	248	316
2090-T8	2319	VPPA	165	283	275
2090-Т8	2319	GMA	207	234	279
2090-T8		EB		324	279
8090-T8	5356	GTA	179	310	316
8090-T8	5356	VPPA	193	296	316
8090T8	5356	GMA	186	276	274
8090-T6		EB		317	317
2094T8	2319	GTA		352	318
2094–T8	2319	VPPA	269	372	319
2094–T8	2319	GMA		283	318
2094-Т8	•••	EB		434	320

a narrow HAZ. The high power beam processes, such as EB or LB welding, can be successfully used for this purpose and already some success has been reported in the literature.

High power beam welding processes have been successfully used to join Al–Li alloys. Electron beam welding has already been used to join alloys 8090, 2090, and 2094. Typically, moderate improvements in strength compared with arc weld properties (Table 2) can be achieved as a result of a finer microstructure, a narrower fusion zone, and better response to aging. Similarly, promising results for LB welded alloy 2090, particularly after PWHT, have been reported.

Diffusion bonding of Al-Li alloys

Diffusion bonding has been successfully used to join titanium and its alloys. However, diffusion bonding of aluminium alloys is rather difficult owing to the tenacious oxide layer which acts as a barrier to diffusion. Most early studies on aluminium diffusion bonding^{321–323} concentrated on the relationships between bond strength and bonding variables such as bonding temperature and time, applied pressure, and surface finish. There has been relatively little progress in understanding the physical and chemical phenomena occurring at and around the bond line.

When diffusion bonding aluminium alloys, the native oxide prevents metal to metal contact and hence the formation of the mechanically sound joint. The oxide therefore needs to be removed, fractured, or disrupted during bonding.

Two bonding methods have nevertheless emerged. The first involves surface modification, such as deposition of a silver or zinc coating on a nominally oxide free aluminium surface by ion plating (the oxide film is removed by ion bombardment before coating)³²⁴⁻³²⁶ and the second uses large scale deformation in the bonded zone to fracture the tenacious layer of alumina on the aluminium surface.^{327,328} The second method is cheaper than the first one. Another approach to eliminate the oxide layer is the use of a clad layer which reacts with the aluminium to form a liquid phase capable of dissolving it.³²⁹

With regard to chemical phenomena, it has been shown³³⁰ that magnesium and lithium, which are present as alloying elements in the Al–Li–Cu alloy 8090, are important to the bonding process because they chemically modify the tenacious alumina layer present on the aluminium surface. Maddrell and Wallach³³¹ investigated diffusion welding of alloy 8090 (Al–9Li–0.5Cu–0.8Mg (at.-%)). The bonding pressure was 2.5 MPa, which was not adequate to substantially disrupt the tenacious oxide layer. Consequently, the bond line remained affected by a semicontinuous layer of oxide, leading to poor mechanical properties.

Dunford and Partridge³²⁴ obtained solution treated bond strengths of 110 MPa in the Al-Li alloy 8090, the oxide layer of which had been removed by sputtering and the clean surface coated with a thin layer of silver. Bonding was carried out at a pressure of 110-130 MPa and a temperature of 280-300°C. In later work, Edwards et al.³³² studied the diffusion bonding of Al-Li alloy 8090 clad with pure aluminium or coated with silver. They reported that there was no need to coat with silver provided that sufficient pressure was applied. The highest lap shear strength in the as solution treated condition of 100-110 MPa was obtained in the alloy clad with pure aluminium by applying a pressure of 50 MPa at 200°C and bonding at 500°C. However, these bonds failed in a brittle manner by bending or peeling. The authors also reported that a clad layer of 0.025 mm or less is necessary to obtain bonds capable of being age hardened.

Urena and Dunkerton³³³ also studied the diffusion bonding of alloy 8090 using pure aluminium (15 μ m thick) or silver interlayers (12.5 μ m thick). They reported that the bonding procedure with the pure aluminium interlayer provides higher integrity joints than joints with silver interlayers because of the absence of intermetallic compound layers. Lap shear strengths nominally of 80 MPa were reported for as bonded joints with a pure aluminium interlayer for conditions of 530° C, 8-10 MPa, and 60 min. Post-bond solution and aging treatments on the pure aluminium interlayer joints led to lap shear strengths in the range 80-100 MPa.

Junai *et al.*³³⁴ successfully demonstrated diffusion bonding of alloy 8090 with 1.5 mm thick sheet. No oxide layer was visible in the interface microstructures of the specimens, which were diffusion bonded using pressures of 6.3 and 8.7 MPa. It was also difficult to locate the interface in these microstructures, indicating that a pressure of 6.3 MPa was adequate to break and disperse the oxide layer present at the joint interface. A peel strength of 90 MPa was obtained for the diffusion bonded joints of alloy 8090 when bonded at a pressure of 6.3 MPa for 60 min at 530°C.

Ricks et al.³³⁵ investigated the effect of a roll clad zinc based alloy interlayer on the diffusion bonding of alloy 8090. The aim of using the interlayer was to produce a TLP which solidifies by diffusion, thus inducing bonding. Either atmospheric or inert gas conditions were used and bonding times of up to 8 h were used with bonding pressures of ~5 MPa at 500–540°C to minimise distortion. Bond shear strengths of 100–120 MPa were achieved, although it was noted that considerable care should be taken at the surface cleaning stage to achieve these properties. Hot tear strengths in excess of 4.5 MPa at temperatures consistent with superplastic deformation of the alloy were also reported, indicating that the superplastic forming–diffusion bonding (SPF–DB) process is feasible.

General comments on diffusion bonding of Al–Li alloys

Some success has already been achieved in diffusion bonding of Al-Li alloys. However, most of the bonding methods present problems when bonding Al-Li alloys, particularly when a superplastic forming (SPF) process is to be applied after bonding and the joint strengths obtained are low. Superplastic forming after bonding restricts the bonding temperature which can be used; ideally the temperature should be lower than or equal to the superplastic temperature for the alloy. Furthermore, the bonding time must be restricted to avoid grain growth in the parent sheet which could affect the SPF properties. The potential of Al-Li alloys for weight reductions in aerospace applications will be further enhanced if a viable SPF-DB process can be developed.

CONCLUSIONS

Use of advanced materials in the aerospace, petrochemical, automotive, and power generation industries will inevitably be limited if they can not be joined effectively to themselves or other alloys.

In principle, solid state bonding and all fusion welding processes are of potential interest for joining of advanced materials. However, some specific requirements of the different application areas make one particular joining process more suitable than the others. The alloys, their weldability, welding metallurgy, and joining processes, which represent state of the art advanced materials and processes, have been covered in this review. Attention was given to material types (such as intermetallic alloys) and joining processes (such as diffusion bonding and laser welding) in which the GKSS Research Center has a current interest.

This review has demonstrated that a variety of the advanced materials of great industrial interest can be welded or bonded with current joining processes. However, further work to optimise joining procedures and to understand the metallurgical changes in the fusion zone for improving weldability still represents a considerable challenge. There is also a need to develop appropriate mechanical testing procedures, taking into account the unique features of the joints, to assess the joint properties and demonstrate the consistency of the joint quality.

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