An Overview of High-Performance Aircraft Structural Al Alloy-AA7085

T. Ram Prabhu¹

Received: 28 January 2015/Revised: 15 March 2015/Published online: 3 May 2015
© The Chinese Society for Metals and Springer-Verlag Berlin Heidelberg 2015

Abstract AA7085 Al alloy is a new-generation, high-strength, quench-insensitive alloy that features in airframe applications such as massive bulkhead and wing spars structures in Airbus A380 and Joint Strike Fighter aircrafts. This alloy has a great potential in replacing currently used high-strength AA7xxx alloys in many international and national military aircraft programs [Hawk, Su-30, MiG series, light combat aircraft (Tejas), Hindustan turbo/jet trainer, light combat helicopter, advanced light helicopter, and fifth-generation fighter aircrafts]. An effort has been made to highlight the processing, heat treatment, properties and applications of the AA7085 Al alloy from the existing literature on the AA7085 alloy. The potential applications of this alloy are presented and compared with the currently used AA7075 Al alloy. A typical process cycle involved in the manufacturing of wrought AA7085 alloy products is described in detail. Particularly, the precipitation mechanisms involved in the different temper conditions are critically reviewed with the supporting microstructures. The mechanical properties with a particular emphasis on tensile, fatigue, and fracture toughness properties are discussed in detail based on the analysis of experimental data. The effects of temper conditions on the stress corrosion cracking and exfoliation corrosion have also been reviewed critically.

KEY WORDS: Al alloys; Heat treatment; Material processing; Mechanical properties; Corrosion

1 Introduction

Al and Ti alloys and polymer matrix composites constitute more than 95% of the materials in a typical aircraft structures. Among the lightweight structural materials, Al alloys take a larger share than any other materials owing to the established mechanical and damage tolerant properties and matured processing technology and inspection techniques [1, 2]. For example, the massive airframe structure of Airbus A380 civil aircraft uses Al alloys more than 60%.

More than 70% of materials used in Boeing aircraft (747, 757, 767, 777 series) primary structures are Al alloys [3].

In contrast, the attractive properties of high specific strength, stiffness, and fatigue properties and the recent advancement in the composite processing technology make the polymer matrix composites (PMCs) a potential alternative to lightweight Al alloys. To give an example, Boeing 787 extended the composite utilization to 50% from 11% utilization in an earlier series of Boeing aircrafts (757, 767, 777) [3]. However, high processing cost, low impact resistance, and mechanical properties variation with environments are still big challenges to the composite materials.

Competition between PMCs and Al alloys drives the Al alloy producers to design new alloys with improved strength, toughness, fatigue, damage tolerant and corrosion properties, and structural performance. Currently, three major groups of Al alloys [Al–Cu (2014, 2024, 2219,

Available online at <http://link.springer.com/journal/40195>

✉ T. Ram Prabhu
ramprabhu.t@gmail.com

¹ CEMILAC, Defence R&D Organization, Bengaluru 560093, Karnataka, India

2618), Al–Zn (7075, 7076, 7010, 7050, 7039), Al–Mg–Zn (6061, 6066)] are commonly used in the aircraft structural applications. Among these, Al–Zn alloys are particularly interesting in the load-bearing structural applications because of superior strength to weight ratio, machinability, and fracture toughness properties [4]. Particularly, the 7075 T6 alloy, which composition (wt%) is 5–6 Zn, 2.1–2.9 Mg, 1.2–2 Cu, 0.2 Ti, 0.3 Mn, 0.4 Si, 0.5 Fe and balance of Al, is a primary choice for the aircraft stress critical applications such as landing gear, wing spars, stringers, horizontal/vertical stabilizer, fuselage frames, bulkheads, engine ribs [2, 5]. However, the stress corrosion cracking (SCC) resistance and the quench sensitivity are still major concerns in this alloy [6]. To improve the SCC resistance, the alloy is usually overaged (T7x). However, the overaging reduces the strength to 80%–85% of peak-aged (T6) condition strength values. The quench sensitivity does not allow large and thick parts to be produced from 7075 alloy due to the wide variation in the mechanical properties across the section.

Efforts have been devoted towards the development of novel high-strength, quench-insensitive Al alloys with improved corrosion resistance particularly the SCC resistance. One of the preferable substitutes for AA7075 Al alloy is the AA7085 Al alloy. This alloy was originally developed by Alcoa in 2003 [7]. It has many attractive properties such as high strength, high fracture toughness, high fatigue strength and SCC resistance, low quench sensitivity, superior ballistic performance compared to other 7xxx series alloys [7–9]. Most importantly, the variation of mechanical properties with section thicknesses is not significant, and particularly, there is no change in the strength properties from 1 to 300 mm thickness [7]. This saves the weight of the aircrafts by reducing the number of parts in the assembly [10]. Also, low quench sensitivity property allows the use of lower cooling rate quenching medium that helps in reducing the residual stress in the product and provides the possibility for press quenching [11]. In press quenching, the product is solution heat treated (SHT) during the extrusion press operation by optimizing the product temperature at the exit side. It eliminates the separate SHT operation in the heat treatment cycle and thus helps in significant cost savings by reducing energy consumption and manufacturing cost. Composition selection is a key to obtain the above properties. It is a well-known fact that the Mg/Zn ratio influences the strength–toughness balance. Higher Mg results in higher strength and lower toughness. Higher Cu, Fe, and Si content impairs the corrosion resistance and damage tolerance (e.g. fatigue strength, fracture toughness) properties by forming deleterious intermetallics such as $\text{Al}_7\text{Cu}_2\text{Fe}$ and Mg_2Si phases [10]. The quench sensitivity of the alloy can be minimized by reducing the solutes

saturation concentration above the maximum solubility limit. Based on these facts, the composition of AA7085 is designed to have low Mg/Zn ratio (high Zn) and low Cu, Fe, Si, and Mg contents compared to that of the other 7xxx series Al alloy. The presence of small amount of Zr content forms fine dispersoids of Al_3Zr that helps in grain refinement. Stringent control of impurities (Fe and Si) to <0.14% is maintained to improve the corrosion resistance and toughness of the alloy.

Currently, Airbus A380 airliner uses this alloy in its largest wing spar structure that is ever made before [5]. Recently, Alcoa has designed a single-piece AA7085 bulkheads forging (152.4 mm thick) to replace a multi-piece bulkheads assembly for the Joint Strike Fighter. This replacement helped in reducing significant weight and material volume with more than 20% cost savings over the machined multi-piece assembly [12]. AA7085 alloy has been widely recognized in ballistic and blast protection applications such as armour plates and panels, underbody blast protection kits, armour roof/side panel window frames, wheel applications, tank hatches, turrets. Information about the ballistic performance of this alloy is available elsewhere [13–15] and outside the scope of this review.

Alcoa has a patent for this alloy and is a monopoly in supplying AA7085 alloy. The development of this alloy by other manufacturers is not found in the international market. Information on AA7085 in the open literature is limited and scattered. None of the reviews on recent developments on aerospace Al alloys have focused on the AA7085 alloy [5, 10, 16].

In the light of these, this review work has been undertaken to cover all aspects of AA7085 for aircraft applications, including processing, heat treatment, and physical, mechanical, and corrosion properties.

2 Material Processing

The AA7085 Al alloy is a wrought alloy. A typical chemical composition (wt%) of the AA7085 alloy is 7–8 Zn, 1.2–1.8 Mg, 1.3–2 Cu, 0.08–0.15 Zr, 0.08 Fe, 0.06 Si, and balance of Al [17]. A typical process sequence for manufacturing wrought products is casting, homogenization, hot forming (forging, extrusion, rolling, and so on), solution treatment, quenching, stress relieving by cold deformation (optional), and ageing treatment [18].

The AA7085 alloy is usually prepared through ingot metallurgy. The pure Al, Mg, Zn, and master alloys (Al–Zr, Al–Cu) are melted at 700–740 °C in a typical oil fired or induction furnace in the normal air atmosphere. Information about melting practice of AA7085 is not available in any specification or open literature. Particularly, there is no

information available on the type of flux, melting environment, alloy mixing and degassing techniques, and skimming practice for ingot melting in an industrial scale. These technologies are currently under proprietorship. As 7xxx series alloys are highly sensitive to the impurities level, the use of conventional salt fluxes has a problem of Na and Cl pick up during melting resulting in the segregation problem after solidification. Also, the increase in Fe level in the melt causes the formation of constituent particles (Fe_3Si) during solidification [19]. The formation of Fe_3Si and the segregation of Na and Cl in the grain boundary severely affect the properties such as fracture toughness and fatigue strength of the alloy. The tight control of composition within the specification is essential to obtain the proper combination of paradoxical properties such as strength and SCC resistance, and strength and toughness. Therefore, the research should be directed to standardize the melting practices of the AA7085 alloy. The preferred ingot casting process for the AA7085 Al alloy is either the direct chill (DC) or the electromagnetic casting (EMC) [11]. Even the solidification rate of the industrial-scale DC casting is much faster than the solidification rate required to attain the equilibrium according to the phase diagram, and the AA7085 alloy invariably forms the cored dendritic structure in the as-cast ingots. Further, the cooling parameters in the DC casting decide the cell size in the ingots. The strength of the alloy decreases with increasing cell size. Therefore, proper selection of cooling parameters is essential to obtain the refined microstructure in the AA7085 alloy. Based on the above facts, it is recommended that the research should be directed to standardize the cooling parameters used in the casting of AA7085 alloy.

Cast ingots should be homogenized to eliminate the solute concentration-related defects (coring, microsegregation) and to redistribute the solutes uniformly before the forming operations. Heating rate, temperature, and time are the important parameters in the homogenization treatment. The homogenization of a AA7085 ingots is typically carried out below 500 °C in two or more increasing temperature steps because most of the eutectic phases (the melting point of the S phase (Al_2MgCu) and the M phase (MgZn_2) is 489 and 479 °C, respectively) melt below about 490 °C [20]. Also, it helps in the formation of coherent Al_3Zr dispersoids that assist in grain refinement and inhibit recrystallization during the forming operations [18, 21–23]. It is noted that the control of the degree of recrystallization by dispersoids helps in improving the fracture toughness of the alloy [18]. The heating rate should be carefully selected for the homogenization treatment so that the properties deterioration due to the local overheating/melting is avoided. The soaking time varies between 3 and 50 h depending on the thickness [11].

The homogenized ingots are the input materials to produce the wrought products (forging, extrusion, rolling). For any forming operations, the deformation characteristics of the alloy should be known to select the appropriate process parameters (forming temperature, time, flow stress, strain rate) so that the desired microstructure and properties can be obtained in the final product. The flow stress characteristics of the alloy were extensively studied by Chen et al. [24]. The flow stress results of the AA7085 alloy fit well with the following hyperbolic sine function:

$$\dot{\epsilon} = A \sinh(\alpha\sigma)^n \exp(Q/RT), \quad (1)$$

where $\dot{\epsilon}$ is the strain rate, A is the constant, σ is the flow stress, Q is the activation energy for deformation (J/mol), R is the universal gas constant, T is the temperature (K), and α is the ratio of the slope of the $\ln \dot{\epsilon}$ versus σ to that of the $\ln \dot{\epsilon}$ versus $\ln Q$ plots at different temperatures.

Figure 1 shows the typical flow stress curves of the AA7085 alloy for both the homogenized and solution-treated conditions, which are reproduced from Ref. [24]. It is found that the flow stress increases to peak value before attaining steady or flow softening state. The flow stress has dependency on both the deformation temperature and the strain rate. The peak flow stress reduces with increasing deformation temperature and/or decreasing strain rate for both the homogenized and the solution-treated samples. The dislocation density decreases with decreasing strain rate. The transformation of low-angle to high-angle grain boundaries also occurs rapidly. With increasing temperature, the precipitate size increases, the volume fraction of precipitates decreases, the subgrains are formed by dislocation annihilation during the dynamic recovery process, and later, the stress-free set of grains is nucleated and grown from the deformed grains during the dynamic recrystallization. Thus, the precipitate coarsening, minimization of the dislocation density, and the dynamic recovery and recrystallization are mainly responsible for the reduction in the flow stress with increasing temperature and decreasing strain rate [25]. Compared to the temperature effects, the strain rate effects on the flow stress are predominant at above 400 °C. As shown in Fig. 1a–c, the flow stress reduces from 120 to 30 MPa when the strain rate changes from 1 to 0.0001 s^{-1} at 400 °C in a homogenized 7085 Al alloy. Similar behaviour is observed for the solution-treated alloy, as shown in Fig. 1d–f. The effects of homogenization treatment and solution treatment on the peak flow stress of the alloy are also different [24, 26]. A typical microstructure of homogenized and solution-treated AA7085 alloy is shown in Fig. 2 [24]. The homogenized structure has relatively coarser grain, precipitate size, and lesser amount of recrystallization than the solution-treated structure due to the precipitate coarsening and grain growth effects. Within the ranges of strain rate

Fig. 1 Flow stress curves of the homogenized AA7085 Al alloy at 350 °C **a**, 400 °C **b**, 450 °C **c** and the solution-treated AA7085 Al alloy at 350 °C **d**, 400 °C **e**, 450 °C **f** under strain rates of 0.0001–1 s⁻¹ [24]

Fig. 2 Typical microstructure of AA7085 Al alloy: **a** homogenized at 450 °C for 24 h and 470 °C for 30 h, **b** solution-treated at 470 °C 1 h [24]

(0.0001 s⁻¹) and deformation temperature (300–450 °C), the homogenized alloy showed lower peak flow stress and flow softening behaviour than the solution-treated alloy. Many factors are attributed to the low flow stress of the homogenized alloy: (1) the relatively low hot deformation activation energy, for homogenized alloy is 187.4 kJ/mol, while for solution-treated alloy is 318.9 kJ/mol; (2) less residual solute retained in the solid solution resulting in low back stress for dislocation motion; (3) coarse grain size; (4) less dynamic precipitation; (5) precipitate coarsening. The optimal deformation temperature and the strain rate for hot working of the AA7085 are found to be 450 °C and 0.0001 s⁻¹, respectively [24].

The comparison of hot deformation activation energy between the AA7075 and the AA7085 alloys showed that the homogenized AA7085 has higher activation energy than the overaged AA7075 (143–156 kJ/mol), whereas in the solutionized condition, the values are very close for them [27, 28]. This shows that the AA7085 alloy has lesser flow stress than the AA7075 alloy at a given temperature and strain rate.

The wrought products are usually heat-treated (solutionizing and ageing) to various temper conditions (discussed in the next section) to obtain the required strength, toughness, and corrosion resistance properties combination. Before the heat treatment, the product is inspected for

surface defects using the fluorescent penetrant test (Mil STD 1907). Then, the product is heat-treated and thoroughly inspected for surface defects by fluorescent penetrant test (Mil STD 1907) and for internal defects by ultrasonic testing (AMS 2630B). The formed AA7085 product is usually cleaned by acid etching to remove oil, grease, and other residues from the surface. It is observed that the acid etching results in surface pitting on the formed 7xxx Al alloy products. The formation of pits is associated with the presence of non-uniform and high residual stress in the material. These stresses are induced during the quenching operation in the heat treatment. This problem has not been discussed in the open literature. Therefore, the research should be directed towards understanding the effects of residual stress on the surface pitting of the formed products and finding the remedial measures to avoid the surface pitting. After cleaning operations, the part is machined, anodized, painted, and installed on the final assembly of the aircraft.

The defect tolerance level [diameter of flat-bottom hole (FBH)] requirement for the flight or mission critical aerospace components is usually 1.2 mm according to the AMS 2630 class AA standard. It is therefore recommended that the future research should be directed to standardize the melting and extrusion practices in an industrial scale to obtain the high-quality input stock material for finished component forming operations such as forging, rolling, and sheet forming.

3 Heat Treatment

AA7085 Al alloy is amenable to various temper conditions. Particularly, T6, T74, two steps high- and low-temperature ageing (HLA), retrogression and reageing (RRA), and dual RRA (DRRA) are most widely used. Typical heat treatment cycles of various temper conditions are listed in Table 1.

The precipitation sequence of the AA7085 Al alloy is G.P zones (coherent) $\rightarrow \eta'$ (semicoherent, MgZn_2) $\rightarrow \eta$ (incoherent) [26, 29, 30].

A typical microstructure of an AA7085 alloy for different heat treatments is presented in Fig. 3. The nature, size, and distribution of the precipitates vary with the different temper conditions which are explained below. T6 is a peak temper condition in which the 5–9-nm fine semicoherent η' precipitates are uniformly distributed inside the grains, and the grain boundaries are completely covered with the same precipitates (continuous precipitation), as shown in Fig. 3a.

The fine η' precipitates are responsible for high strength. However, the presence of continuous precipitates around the grain boundary creates the galvanic couple resulting in poor corrosion resistance, especially the resistance to SCC. Also, the copper content of the grain boundary precipitates (GBPs) is lesser due to low ageing temperature (120 °C) [31]. The low copper content further increases the potential difference between the matrix and GBPs, which results in poor SCC resistance.

The T74 temper is an overaged condition that provides excellent corrosion resistance, especially the resistance to SCC with the sacrifice on the strength of 10–15% reduction. In this condition, the relatively coarse size (40–65 nm) is found for the precipitates ($\eta' + \eta$) inside the grains, and the same precipitates are sparsely distributed around the grain boundary, as shown in Fig. 3b. These coarse precipitates reduce the strength of the alloy by reducing the total volume fraction of the precipitates and easy dislocation by-pass by Orowan mechanisms.

HLA, RRA, and DRRA are new temper conditions specifically designed to optimize the strength and corrosion resistance (SCC) properties [32–35]. Out of these, RRA and DRRA are claimed to retain the strength similar to a T6 peak-aged condition with excellent SCC resistance similar to a T74 overaged condition [32, 34, 36]. In the

Table 1 Typical heat treatment cycles, tensile, toughness, and corrosion properties of the AA7085 alloy subjected to different temper and ageing conditions [7, 26, 32, 39]

Temper conditions	Ageing cycle	I_{SSRT}	Yield strength (MPa)	Tensile strength (MPa)	Elongation (%)	Max. exfoliation corrosion depth (μm)	Unit initiation energy (N/mm)
T6	120 °C, 24 h	0.6	530	582	13	78	148
T74	110 °C/6 h + 160 °C/10 h	0.8	457	525	10	52	182
RRA	120 °C/24 h + 180 °C/0.5 h + 120 °C/24 h	0.95	528	578	11	60	169
DRRA	120 °C/24 h + 180 °C/0.5 h + 120 °C/12 h + 180 °C/0.5 h + 120 °C/24 h	0.95	528	578	12	53	184
HLA	200 °C/5 min + 120 °C/24 h	–	430	510	13	130	–

Fig. 3 Typical transmission electron micrographs of AA7085 Al alloy subjected to T6 **a**, T74 **b**, RRA **c**, DRRA **d**, HLA **e** temper conditions [26]

RRA treatment, the G.P zones and η' precipitates are partially redissolved in the retrogression stage to enhance the total volume fraction of η' during the reageing stage, thus enhancing the strength without affecting the SCC resistance. This effect is stronger for DRRA treatment. While comparing the nature of precipitation of the RRA or DRRA treatments with other tempers, the precipitates inside the grains (very fine-size η' , homogenous distribution) are similar to the T6 condition, whereas the GBPs (coarse-size η' , discontinuous, sparse distribution) are similar to the T74 condition with enriched copper content, as shown in Fig. 3c, d [37, 38]. In the case of HLA temper condition, the size of precipitates inside the grains (8–10 nm) and the grain boundary (23–32 nm) are in between T6 and T74 conditions, as shown in Fig. 3e.

The quench delay time is an important factor in the above heat treatment operations. This parameter has not been discussed anywhere in the open literature. As the super saturation of alloying elements reduces with the increasing quench delay time, the availability of solutes for precipitates formation reduces. Thus, the strength is reduced significantly. Therefore, efforts should be directed to identify the maximum quench delay time for the 7085 alloy with respect to the section thickness and the solution treatment temperature for above heat treatment temper conditions.

The beneficial effects of RRA and DRRA are shown only in the laboratory scale. The establishment of these heat treatment technologies in industrial scales is still an open problem. It is recommended that efforts should be focused on standardizing these technologies in future.

The relation between the precipitate characteristics (size, amount, interparticle spacing, and distribution inside and along the grains) and the corrosion and mechanical properties will be discussed in detail in the subsequent sections.

4 Mechanical and Physical Properties

The typical physical and mechanical properties of the AA7085 for various tempers are given in Tables 1 and 2 [26]. From Table 1, it is observed that the strength increases in the following order: HLA < T74 < DRRA $\approx$ RRA $\approx$ T6. The variation in strength among different temper conditions is attributed to the way in which the matrix and GBPs are distributed after ageing treatments. Also, the amount, size, and the nature of the precipitates vary with the ageing treatments. In AA7085 Al alloy, the G.P zones and η' precipitates decide the strength of the alloy. In T6 temper, the very fine and homogeneously distributed G.P zones are responsible for high strength, whereas in DRRA or RRA temper, the dense reprecipitation of η' precipitate during reageing helps in increasing the strength close to the T6 temper condition. In T74 temper, the η' precipitate coarsening, the sparse distribution of precipitates (η'), and less volume fraction of precipitates reduce the strength of the alloy. In the case of HLA temper, initial high-temperature ageing (200 °C) converts all G.P zones into coarse η' and η precipitates that forbid the fine η' precipitates formation. The subsequent low-temperature ageing retains most of the coarse

Table 2 Physical properties of the AA7085-T7452 die forging [7]

Property	Typical value
Density (g/cm ³)	2.85
Electrical conductivity (%IACS)	24–41.5
Melting point (°C)	552–635
Specific heat at 100 °C (J/(g °C))	0.884
Thermal conductivity at 100 °C (W/(mK))	159.5
Coefficient of thermal expansion between 20 and 100 °C (10 ^{−6} /°C)	24.7

precipitates resulting in substantial reduction in the strength of the alloy [35]. More information about the precipitation mechanisms for different temper conditions was given in the previous section.

Chen et al. [32] studied the effects of a corrosive medium (3% NaCl + 0.5% H₂O₂ solution) and air on the tensile properties of the AA7085 alloy. It was found that the alloy loses its strength, toughness, and elongation properties when exposed to the corrosive medium, as shown in Fig. 4. The strength degradation is caused by the attack of corrosive medium on the grain boundary. The corroded grain boundaries are the potential sites for crack nucleation under tensile loading resulting in significant reduction in the strength of the alloy. The loss of ductility indicates that the alloy is prone to SCC (Table 3).

Table 4 gives the tensile properties of AA7085 T7415 for various section thicknesses and sample directions. The effects of section thickness on the strength properties are insignificant. However, there is a slight decrease in the ductility with increasing section thickness. The sample

Table 3 Mechanical properties of the AA7085-T7452 die forging for the section thicknesses of 25.1–300 mm, and the properties are measured in the samples of longitudinal orientation [17]

Property	Typical value
Yield strength (MPa)	448–462
Ultimate tensile strength (MPa)	496–503
Elastic modulus (GPa)	69.6
Elongation (%)	7–10
Shear strength (MPa)	282
Shear modulus (GPa)	26.9
Poisson ratio	0.33
Fracture toughness (MPa m ^{1/2})	137–214
Fatigue strength of a smooth specimen for $R = +0.1$ condition for the life of 10 ⁷ cyc (MPa)	289–344
Fatigue strength of an open hole ($K_t = 2.3$) specimen for $R = +0.1$ condition for the life of 10 ⁷ cyc (MPa)	165–193

Table 4 Effects of section thickness and sample orientation on tensile properties of the AA7085-T7452 die forging [17]

Section thickness (mm)	Direction	Yield strength (MPa)	Tensile strength (MPa)	Elongation (%)
25.4–50.8	L	448	496	10
	LT	427	482	8
	ST	407	482	5
102–152	L	448	496	9
	LT	427	482	7
	ST	407	482	4
254–305	L	448	496	7
	LT	427	482	4
	ST	407	482	3

Fig. 4 Effect of air and corrosive medium (3% NaCl + 0.5% H₂O₂) on tensile properties of the AA7085 Al alloy under different temper conditions (T6, T74, and DRRA) [39]

direction influences the tensile properties of the alloy significantly. The sample extracted from the longitudinal direction (the direction of working) shows higher strength properties than that from the transverse direction. The typical stress versus strain curve of an AA7085 T7452 forging for tensile loading condition is shown in Fig. 5. The curve has smooth linear elastic and flat plastic regions. The sample orientation influences the tensile strength and ductility of the alloy. The short transverse direction sample shows lower tensile strength and ductility than other two directions samples in the tensile testing.

The effect of quenching rate on the tensile properties of the AA7085 T6 alloy was studied by Chen et al. [39]. They concluded that the quenching rate has significant influence on the yield and tensile strength properties of the alloy. For instance, the yield strength of about 35 MPa was reduced when the quenching medium was changed from cold water

Fig. 5 Typical stress–strain curves of AA7085 Al alloy [7]

(150 °C/s) to oil (50 °C/s). It was also found that elongation is not significantly affected by the quenching rate.

The thermal exposure effects on the room-temperature tensile properties and the electrical conductivity of the AA7085 T7452 die forgings were studied by Jabra et al. [6]. It was found that the room-temperature tensile strength is stable up to 2 h at 180 °C exposure.

The increasing exposure temperature or time affects the strength significantly. For instance, the room-temperature tensile strength becomes half when the alloy is exposed to 10 h at 230 °C, respectively. At higher temperatures, the strength decreases rapidly within the 0.1 h of exposure and then stabilized up to 1000 h. The electrical conductivity of the alloy increases with the exposure temperature and time. For instance, the alloy exposed to 230 °C shows the room-temperature conductivity in the range of 41–45% IACS for the exposure times of 0.1–1000 h.

The plain strain fracture toughness (K_{Ic}) of an AA7085-T7452 alloy ($>137 \text{ MPa m}^{1/2}$) is much higher than that of an AA7075 T651 alloy ($27 \text{ MPa m}^{1/2}$) [5]. The K_{Ic} of an AA7085-T7X is sensitive to the ageing time, the plate thickness, and the sample orientation. The study by Shuey et al. [9] showed that the K_{Ic} increases with increasing ageing time. The alloy exhibits strong anisotropy in the fracture toughness properties. The K_{Ic} increases in the following order of specimen orientations: short transverse < long transverse < longitudinal. The orientation dependence of K_{Ic} values is attributed to the anisotropy of grain boundaries and size and amount of precipitate constituents. Further, the critical strain per unit plate thickness computed from the data of K_{Ic} and tensile properties for various plate thicknesses shows that the strength and toughness are independent of the plate thickness. This property is unique to the AA7085 alloy in the 7xxx series. The effect of temper condition on the fracture toughness of

AA7085 alloy is found to be significant. A study by Chen et al. [32] showed that the fracture toughness of the AA7085 increases in the following order: T6 < RRA < T74 < DRRA, as given in Table 1. The difference between the effects of T74 and DRRA tempers is that T74 improves the toughness at the expense of strength by $\sim 13.6\%$, whereas DRRA improves both the strength and the toughness.

Chen et al. [32] investigated the fracture surface morphology of the AA7085 for different tempers. The fractographs of different temper conditions of an AA7085 alloy are shown in Fig. 6. The fracture surface of the tensile tested AA7085 T74 is mostly transgranular with dimples. Voids or cracks initiate at matrix constituents and propagate by internal necking between voids at matrix constituents resulting in the transgranular failure [40]. However, the fracture surface appearance varies with different temper conditions. This variation is attributed to the strength difference between the matrix and grain boundary regions is generally related to the size, amount, and distribution of the precipitates and the precipitate-free zone width [41]. The amount of intergranular fracture and the size of the dimples increase from T74 to T6 conditions. In T6 condition, the increase in matrix strength by fine precipitates and continuous precipitates along the grain boundaries promotes the intergranular fracture by easy crack propagation. The RRA and DRRA tempered alloys show predominantly transgranular features with little amount of intergranular fracture because of sparsely distributed coarse GBPs. The fracture morphology also varies with the quenching rate in the T6 treatment. Chen et al. [39] found that the alloy quenched in cold water shows predominately intergranular fracture features. The fracture surface changes to the combination of trans- and intergranular fracture when quenched in oil medium.

The strain concentration behaviour of an AA7085 T7415 alloy was extensively studied by Menzemer et al. [42] for two configurations, open hole and filled hole with fasteners, with different sample widths and thicknesses. It was found that the alloy is sensitive to the strain concentration. The filled-hole fastener configuration shows higher strain concentration factor than the open-hole condition. The strain concentration factor increases with increase in the thickness and decrease in the width of the sample by changing the stress distribution around the hole. A typical fracture surface of an AA7085 T7415 alloy that is deformed by tensile loading consists of mixed mode (ductile–brittle) features, which is similar to that shown in Fig. 6a. Careful examination of the fracture surface at higher magnifications reveals that dimples are distributed along the high-angle grain boundaries. These dimples are formed by coalescence of voids present at GBPs and coarse

second-phase particles ($\text{Al}_7\text{Cu}_2\text{Fe}$). These are preferable sites of crack nucleation at high stress levels. The crack usually initiates at the dimples present at the grain boundary triple junctions and propagates along the grain boundaries resulting in the brittle intergranular fracture.

The high-cycle fatigue properties of an AA7085 alloy were investigated by Luong and Hill [43]. Fatigue tests were conducted in a stress ratio of $R = +0.1$ and frequency of 6–7 Hz for a range of constant stress conditions. The S–N plot of the AA7085 alloy is shown in Fig. 7. The results of fatigue tests show that the alloy has a life of 1×10^6 cycles for maximum applied stress of 240 MPa. The fatigue strength may be further improved by the laser peening or the combination of laser peening and anodizing treatment. Both these treatments increase the fatigue limit to 320 MPa for the life of 1×10^6 cycles. This improvement is attributed to many factors: (1) a higher and deeper level of residual compressive stress; (2) the low level of surface cold working; and (3) the improved surface finish with reduced surface waviness. The axial fatigue strength of the AA7085 alloy is much better than that of AA7075 alloy. While comparing with the AA7050 alloy, the AA7085 alloy has 10%–20% fatigue strength improvement in both the high-cycle (10^7 cyc) and low-cycle (10^5 cyc)

regimes for the conditions of $R = +0.1$, frequency of 6–7 Hz, and $K_t = 1$ [7]. The fractograph of fatigue-failed sample (as shown in Fig. 8) shows clearly the fatigue crack

Fig. 7 Typical S–N curves of as machined and laser peening-treated AA7085 Al alloy ($R = +0.1$, $K_t = 1$, RH >90%) [43]

Fig. 6 Fracture surface morphologies of AA7085 Al alloy under conditions of T6 **a**, T74 **b**, RRA **c**, HLA **d** temper conditions [32]; high-magnification images of fracture surface of AA7085 Al alloy show the cracking along the grain boundary triple points, dimples, and voids **e**, cracking around coarse intermetallic particles **f** [42]

Fig. 8 Fracture surface of an as machined AA7085 Al alloy, and surface crack initiation site is clearly observed [43]

initiated surface location and beach marks and striations features [43].

Laser peening and low plasticity burnishing techniques are reported to improve both the fretting and joint fatigue strength of AA7085 alloy in the typical aerospace joint assemblies [44]. These techniques provide a hard and rough surface with sufficient magnitude of compressive residual stress to a considerable depth that inhibits fatigue crack nucleation and propagation. Also, it was found that the anodizing techniques alone based on chromic acid, phosphoric acid, sulphuric/boric acid, and chromium conversion coating do not improve much the fretting/joint fatigue performance of the AA7085 alloy.

As most of the aerospace structural components are joined by welding, the weldability of this alloy is a very important property to be known. The welding of an AA7085 alloy has not been researched yet. Hence, it is recommended that the research should be directed to explore the weldability of the alloy.

5 Corrosion

5.1 General Corrosion

The electrochemical response of an AA7085 alloy for different temper conditions was investigated by Chen et al. [26]. The open circuit potential (OCP) values for various tempers are arranged in the following increasing order: HLA (−0.903 V) < T6 (−0.779 V) < RRA (−0.776 V) < DRRA (−0.758 V) < T74 (−0.756 V). It was found that the OCP depends on the precipitates composition. The heat treatment practices decide the composition of the precipitates. In AA7085 alloy, the elements except Cu are not sensitive to the heat treatment conditions. Therefore, OCP values are related to the Cu content in the precipitate. The higher positive corrosion potential of the DRRA and T74 tempers is due to the diffusion of copper from the solid solution to the GBPs. The potential difference between the

matrix and the grain boundary is reduced significantly due to the equalization of the copper and thus improves the corrosion resistance of the alloy in T74 and DRRA tempers [45].

The quenching rate in the T6 temper condition also influences the OCP value of the alloy. Chen et al. [39] studied the effect of quenching rate on the corrosion potential of an AA7085 T6 alloy. It was found that the potential increases with increasing cooling rate from 1 to 150 °C/s. Increasing the cooling rate enriches the copper content in the precipitate and therefore enhances the OCP, in other words the electrochemical reactivity of the GBPs.

5.2 Stress corrosion cracking (SCC)

The SCC susceptibility of an AA7085 alloy was tested by the slow strain rate ($<10^{-6} \text{ s}^{-1}$) testing (SSRT) in air and in 3% NaCl + 0.5% H_2O_2 corrosive solution [39]. The samples tested under the corrosive conditions showed much lower ultimate tensile strength and elongation than that tested in air. The properties degradation indicates that the alloy is susceptible to SCC. Also, the rate of quenching in the heat treatment cycle has influence on the properties degradation. The quenching rate intermediate between the cold water and the air was found to show relatively lesser properties degradation or, in other words, the better SCC resistance [39]. TEM studies showed that the intermediate quenching rate (oil quenching, 50 °C/s) provides the slightly coarse, discontinuous GBPs with no precipitate-free zone that result in improvement in the SCC resistance. Chen et al. [32] studied the SCC resistance of an AA7085 alloy under different temper conditions. SCC resistance was related to the ISSRT value using the following relation:

$$I_{\text{SSRT}} = I_{\text{cor}}/I_{\text{air}}, \quad (2)$$

where I represents the loss of elongation in the slow strain rate tensile testing. Cor and air denote the corrosive solution and the air medium, respectively.

SCC resistance increases with increasing I_{SSRT} values. Results of I_{SSRT} for various temper conditions are given in Table 1. From the results, it was found that the overageing (T74) and double DRRA treatments provide the highest and nearly equivalent SCC resistance, and the peak temper (T6) condition provide the least SCC resistance. The SCC resistance of the RRA treatment lies intermediate between T6 and T74 (or DRRA) conditions.

Moran et al. [46] studied the SCC behaviour of an AA7085-T76 alloy tested for different stress levels (179, 241, 276 MPa) according to the ASTM G47/G44. They have tested 280 samples with varying thicknesses ranging from 101.6 to 203.2 mm. The results are presented in Fig. 9. As shown in Fig. 9, the SCC resistance increases

with decreasing applied stress. Other studies showed that the forged samples fail between 63 and 120 days when they were tested at 241 MPa stress in a 3.5% NaCl solution medium according to the ASTM G44 standard [7]. While compared with the AA7075-T6 alloy (<68 MPa), the AA7085 alloy has much higher SCC threshold stress value (>179 MPa). The electrical conductivity measurement on the AA7085 T7452 die forging shows that the alloy possesses the conductivity ranging from 41.2% to 42.5% IACS [7].

Fig. 9 Effect of stress on the stress corrosion cracking behaviour of the AA7085 Al alloy [46]

5.3 Exfoliation Corrosion

Exfoliation corrosion is a form of intergranular corrosion in which the surface layer exfoliates when the alloy is exposed to a corrosive medium. The wrought 7xxx sheets and plates are highly susceptible to this form of corrosion when the product has aligned, elongated grain structures parallel to the surface [47]. This structure increases the potential difference between the grain interior and the grain boundary. Besides, it modifies the precipitate size and distribution in the grain and the grain boundary, and also increases the width of the precipitate-free zone resulting in attacks on the grain boundaries. Corrosion proceeds along the grain boundary and spreads parallel to the surface. The volume of the corrosion product is greater than that of unaffected parent metal that forces the corrosion product to exfoliate from the surface.

Exfoliation corrosion morphology of the AA7085 alloy for various temper conditions is shown in Fig. 10 [26]. The surface of the T6 tempered sample shows the severe exfoliation/peel-off, whereas the sample of T74 or DRRA treated shows no visual features of exfoliation. From these observations, it can be concluded that both the DRRA and T74 temper conditions provide the best exfoliation corrosion resistance, whereas the HLA and T6 temper conditions are highly susceptible to exfoliation corrosion. The best exfoliation corrosion resistance of the DRRA/T74 tempered AA7085 alloy is attributed to two factors: (1) the high copper content in the GBPs; (2) the sparse distribution

Fig. 10 Surface morphology of an AA7085 Al alloy affected by exfoliation corrosion under different temper conditions: **a** T6, **b** T74, **c** RRA, **d** DRRA, **e** HLA [26]

of coarse η' precipitates in the grain boundary with very thin precipitate-free zone [26, 31, 48]. The poor exfoliation corrosion resistance of the T6 temper is due to the low Cu content in the GBPs and easy crack propagation in the continuous network of GBPs [49]. In the case of HLA temper, although the precipitates are coarse and discontinuous in the grain boundary, the precipitates are devoid of copper content due to the high-temperature ageing that inhibits the diffusion of copper from the matrix to the grain boundary. Thus, the potential difference between the matrix and grain boundaries increased resulting in higher susceptibility to the exfoliation corrosion. The degree of exfoliation corrosion attack, in other words the maximum corrosion depth, as given in Table 1, for various tempers, was found in the following ascending order: T74, DRRA < RRA < T6 < HLA [26].

The exfoliation corrosion resistance of an AA7085-T76 alloy was rated P (i.e. no exfoliation) according to the ASTM G 34 even after 4 years of seacoast exposure [46, 50]. In contrast, the AA7075-T6 alloy exhibits severe exfoliation (rated EC/ED) after 4 years of exposure. This shows that the AA7085 alloy has much better exfoliation corrosion resistance than the AA7075 alloy.

6 Concluding Remarks

The purpose of this work is to highlight the processing and properties and aerospace applications of the AA7085 Al alloy from the recent literatures. A significant amount of experimental data from various literatures have collated and analysed to draw conclusions as follows:

1. AA7085 alloy is a potential replacement for AA707X or AA705X series alloys in the airframe applications such as bulkheads, wing spar, landing gears, fuselage frames.
2. The preferred ingot preparation route for the alloy is the DC or EMC.
3. The typical homogenization temperature of the alloy should be just below 500 °C so that all the eutectics melt and the coherent Al_3Zr dispersoids can easily form to assist in grain refinement and to inhibit the recrystallization during forming operations.
4. The optimal deformation temperature and strain rate for hot working of the AA7085 are found to be 450 °C and 0.0001 s^{-1} , respectively.
5. Homogenous distribution of very fine-size copper-enriched η' precipitates inside the grains and the sparse distribution of coarse-size copper-enriched η' precipitates along the grain boundaries are the preferred microstructure to obtain the best SCC resistance without sacrificing the strength properties of the alloy.

6. Double retrogression and reageing (DRRA) is the most preferable temper condition to optimize the SCC resistance and the strength of the alloy.
7. The mechanical properties of the alloy are highly insensitive to the thickness up to 305 mm and sensitive to the quenching rate, sample orientation, strain concentration, and temperature beyond 180 °C. The strength with respect to temper treatment has the following increasing sequence: HLA < T74 < RRA $\approx$ T6 $\approx$ DRRA. The resistance to SCC with respect to temper treatment has the following increasing sequence: HLA < T6 < RRA $\approx$ T74 $\approx$ DRRA.

Aircraft design and qualification testing are a costlier affair. Any new alloy development takes the time to find an entry to the aircraft. The substitution of a new alloy to an old alloy demands the design revalidation and repetition of functional testing. Also, most of the aerospace components are generally designed with an adequate factor of safety margin of 1.3–1.5. These factors make the designer to be hesitant to try the new alloy unless there are significant improvements in the properties of the alloy. As AA7085 alloy has significant advantages in terms of low quench sensitivity, superior SCC and exfoliation corrosion resistance, excellent ballistic performance, static and dynamic strength properties, it will definitely replace most of the high-strength 7xxx alloys in the structural components of the future aircrafts.

References

- [1] I.J. Polmear, *Light Alloys: Metallurgy of the Light Metals*, 2nd edn. (Edward Arnold Publishers, London, 1989), pp. 29–235
- [2] T. Warner, Mater. Sci. Forum **519–521**, 1271 (2006)
- [3] A.S. Warren, Mater. Forum **28**, 24 (2004)
- [4] A. Heinz, A. Haszler, C. Keidel, S. Moldenhauer, R. Benedictus, W.S. Miller, Mater. Sci. Eng. A **280**, 102 (2000)
- [5] T. Dursun, C. Soutis, Mater. Des. **56**, 862 (2014)
- [6] J. Jabra, M. Romios, J. Lai, E. Lee, M. Setiawan, E.W. Lee, J. Witters, N. Abourialy, J.R. Ogren, R. Clark, T. Oppenheim, W.E. Frazier, O.S. Es-Said, J. Mater. Eng. Perform. **15**, 601 (2006)
- [7] L. Mueller, L. Suffredini, D. Bush, R. Sawtell, P. Brouwer, *ALCOA 7085 Forgings: 7th Generation Structural Solutions*, (Alcoa Center Publisher, USA, 2004), www.files.engineering.com/download.aspx-ALCOA-7085-Forging. Accessed 10 Jan 2004
- [8] D. Chakrabarti, J. Liu, R. Sawtell, G. Venema, Mater. Sci. Forum **28**, 969 (2004)
- [9] R.T. Shuey, F. Barlat, M.E. Karabin, D. Chakrabarti, Metall. Mater. Trans. A **40**, 365 (2009)
- [10] E.A. Starke, J.T. Staley, Prog. Aerosp. Sci. **32**, 131 (1996)
- [11] European Patent Specification EP 2, 288, 738 B1, 12 Feb 2014
- [12] Alcoa Inc. Web Catalogue: AA7085 (2010), <http://www.alcoa.com/>. Accessed 16 Jan 2010
- [13] D. Gallardy, Ballistic Evaluation of 7085 Aluminium, (Army Research Laboratory Publisher, 2012), <http://www.dtic.mil/cgi-bin/GetTRDoc?AD=ADA562407>. Accessed 16 Jan 2012

- [14] W. Gooch, *Unpublished Data on 7085*, (U.S. Army Research Laboratory Publisher, 2009), <http://www.dtic.mil>. Accessed 16 Jul 2009
- [15] MIL-DTL-32375 (MR), Specification, Armor Plate Aluminium Alloys: 7085 (Unweldable applique) (2012)
- [16] W. Cassada, J. Liu, J. Staley, *Adv. Mater. Process.* **160**, 27 (2002)
- [17] SAE AMS 4403 Specification, Aluminum Alloy Die Forgings: 7.5Zn–1.6Cu–1.5Mg–0.12Zr (7085-T7452) (Solution heat treated, compression stress-relieved, and overaged), 2006
- [18] P.A. Rometsch, Y. Zhang, S. Knight, *Trans. Nonferrous Met. Soc. China* **24**, 2003 (2014)
- [19] F.C. Campbell, *Light Weight Materials*, 1st edn. (ASM International, Materials Park, Ohio, 2012), pp. 33–119
- [20] D.K. Xu, N. Birbilis, R.A. Rometsch, *Mater. Sci. Eng. A* **534**, 234 (2012)
- [21] M. Conserva, D.E. Russo, O. Caloni, *J. Metall. Trans. A* **2**, 1227 (1971)
- [22] X. Lu, E. Guo, P. Rometsch, L. Wang, *Trans. Nonferrous Met. Soc. China* **22**, 2645 (2012)
- [23] J.D. Robson, P.B. Prangnell, *Acta Mater.* **49**, 599 (2001)
- [24] S. Chen, K. Chen, G. Peng, X. Chen, X. Ceng, *J. Alloys Compd.* **537**, 338 (2012)
- [25] G.Y. Lin, Z.F. Zhang, H. Zhang, D.S. Peng, J. Zhou, *Acta Metall. Sin. (Engl. Lett.)* **21**, 109 (2008)
- [26] S. Chen, K. Chen, G. Peng, L. Jia, P. Dong, *Mater. Des.* **35**, 93 (2012)
- [27] E. Cerri, E. Evangelista, A. Forcellese, H. McQueen, *Mater. Sci. Eng. A* **197**, 181 (1995)
- [28] T. Sheppard, A. Jackson, *Mater. Sci. Technol.* **13**, 203 (1997)
- [29] K. Stiller, P. Warren, V. Hansen, J. Angenete, J. Gjnnes, *Mater. Sci. Eng. A* **270**, 55 (1999)
- [30] X. Yang, J. Liu, J. Chen, C. Wan, L. Fang, P. Liu, C. Wu, *Acta Metall. Sin. (Engl. Lett.)* **27**, 1070 (2014)
- [31] T. Marlaud, A. Deschamps, F. Bley, W. Lefebvre, B. Baroux, *Acta Mater.* **58**, 248 (2010)
- [32] S. Chen, K. Chen, P. Dong, S. Ye, L. Huang, *Trans. Nonferrous Met. Soc. China* **24**, 2320 (2014)
- [33] B.M. Cina, U.S. Patent 3,856,584, 24 Dec 1974
- [34] G. Peng, K. Chen, S. Chen, H. Fang, *Mater. Sci. Eng. A* **528**, 4014 (2011)
- [35] D. Wang, D. Ni, Z. Ma, *Mater. Sci. Eng. A* **494**, 360 (2008)
- [36] M. Talianker, B. Cina, *Metall. Mater. Trans. A* **20**, 2087 (1989)
- [37] N. Danh, K. Rajan, W. Wallace, *Metall. Mater. Trans. A* **14**, 1843 (1983)
- [38] F. Viana, A. Pinto, H. Santos, A. Lopes, *J. Mater. Process. Technol.* **92**, 54 (1999)
- [39] S. Chen, K. Chen, G. Peng, L. Xin, X. Chen, *Trans. Nonferrous Met. Soc. China* **22**, 47 (2012)
- [40] M.E. Karabin, F. Barlat, R.T. Shuey, *Metall. Mater. Trans. A* **40**, 354 (2009)
- [41] G.M. Ludtka, D.E. Laughlin, *Metall. Mater. Trans. A* **13**, 411 (1982)
- [42] C.C. Menzemer, D.F. Lam, T.S. Srivatsan, *J. Mater. Eng. Perform.* **19**, 705 (2010)
- [43] H. Luong, M.R. Hill, *Mater. Sci. Eng. A* **477**, 208 (2008)
- [44] C. Giummarra, H.R. Zonker, *ICAF 2005 Proceedings, Hamburg, Germany*, 8–10 June 2005
- [45] B. Sarkar, M. Marek, E.A. Starke, *Metall. Mater. Trans. A* **12**, 1939 (1981)
- [46] J.P. Moran, R.J. Rioja, E.L. Colvinm, in *Proceedings of the Light Metals Technology Conference, Saint-Saveur, Québec, Canada*, 24–26 Sep 2007
- [47] J. Wloka, T. Hack, S. Virtanen, *Corros. Sci.* **49**, 1437 (2007)
- [48] Q. Meng, G. Frankel, *J. Electrochem. Soc.* **151**, B271 (2004)
- [49] R.K. Viswanadham, T.S. Sun, J.A.S. Green, *Metall. Mater. Trans. A* **11**, 85 (1980)
- [50] ASTM G34-01 Specification, Standard test Method for Exfoliation Corrosion Susceptibility in 2XXX and 7XXX Series Aluminium Alloys (EXCO Test), 2007