EFFECT OF THERMAL EXPOSURE ON THE STRENGTH-TOUGHNESS BEHAVIOR OF ELEVATED TEMPERATURE SERVICE ALUMINUM ALLOYS

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ABSTRACT Aluminum alloys intended for application on supersonic transport aircraft must exhibit stable mechanical properties at temperatures greater than 90°C for very long service life. The current study examines the effect of long term thermal exposure on the tensile properties and fracture behavior of Al-Cu-Mg-Ag alloys, C415 and C416, and Al-Cu-Li-Mg-Ag alloys, RX818 and ML377, all of which are being considered for airframe structural application on supersonic aircraft. The strength-toughness behavior of these developmental alloys is compared with the properties of the Concorde alloy CM001 and the Russian alloy 1143. All four developmental alloys exhibited a superior combination of strength and toughness as compared to both CM001 and 1143 both prior to thermal exposure and after extended thermal exposure at temperatures up to 107°C.

Keywords: *Al-Cu-Mg-Ag alloys, Al-Cu-Li-Mg Ag alloys, tensile properties, fracture toughness, thermal exposure*

INTRODUCTION

Development of economically viable supersonic transport aircraft requires materials that exhibit stable mechanical properties for extended operating time at elevated service temperatures. Supersonic aircraft operating in the Mach 2 regime will experience airframe temperatures in excess of 90°C with anticipated service lifetime greater than 50,000 hours. Two Al-Cu-Mg-Ag alloys, C415 and C416, and two Al-Cu-Li-Mg-Ag alloys, RX818 and ML377, developed by Alcoa and Reynolds Metals Company, respectively, to meet these requirements were evaluated through a government and industry alloy development program [1,2]. Two Al-Cu-Mg alloys, the Concorde alloy CM001 and the Russian alloy 1143 were evaluated as baseline materials. The strength-toughness behavior of the developmental alloys is compared with properties for CM001 and 1143. During the alloy development program, tensile properties and fracture toughness were determined after isothermal exposures at 93°C, 107°C, and 135°C for various exposure times; properties were measured at -54°C, room temperature, 93°C, 107°C, and 135°C for both as received (T8) and thermally exposed conditions [3,4]. Fracture toughness was evaluated using J-integral test methodology involving compact tension specimens for each material condition and test temperature combination. In addition, fracture toughness was evaluated from R-curves determined using 56-cm wide middle crack tension (M(T)) panels for material in the T8 condition and after 5000 hours exposure at 93°C and 107°C to facilitate comparison of the developmental alloys with current airframe structural materials. Metallurgical analysis of the developmental alloys and fractographic examination of failed specimens were performed throughout the development program [1,5-8] to examine thermal stability of the microstructures and to correlate the observed data trends with fracture behavior. This paper examines the effect of long term (5,000 – 15,000 hour) thermal exposure on the room temperature tensile properties and fracture behavior of C415, C416, RX818, and ML377.

MATERIALS

Alloys C415, C416, RX818, and ML377 were produced through an alloy development program designed to improve the strength-toughness combination and thermal stability of aluminum alloys for elevated temperature service. The properties of the Concorde alloy CM001 and the Russian alloy 1143 were considered as a baseline. The actual chemical compositions of C415, C416, RX818, and ML377 are provided in Table 1 along with the nominal compositions for CM001 and 1143.

Table 1. Chemical Composition of Materials (in weight percent).

Material	Li	Cu	Mg	Ag	Zr	Mn	Fe	Si	Ti	Ni
C415		5.00	0.80	0.50	0.13	0.60	0.06	0.04		
C416		5.40	0.50	0.50	0.13	0.30	0.06	0.04		
ML377	0.95	3.50	0.39	0.42	0.12	0.30				
RX818	0.99	3.76	0.51	0.36	0.14					
CM001	--	2.5	1.5	--	--	--	1.0	0.22	0.09	1.12
1143	--	2.3	1.5	--	--	--	0.5	0.20	0.07	0.60

The alloys C415 and C416 are modifications of the Al-Cu-Mg alloy 2519, a material with relatively high strength and thermal stability due to thermally stable Cu-containing phases [9]. Additions of Ag and Mg in C415 and C416 promote formation of the Ω variant of Al_2Cu as a thermally stable strengthening phase [2]. The Cu/Mg ratio for C415 is 6, resulting in Ω being the dominant precipitate phase in the T8 condition and for C416 the Cu/Mg ratio is 10, resulting in the presence of both Ω and θ' phases [2,6]. The Cu and Mg contents and the Cu/Mg ratio of both alloys were selected to reduce the volume fraction of intermetallic phases to improve fracture toughness. Both C415 and C416 were cast in 550 kg ingots, processed to produce 2.3 mm thick sheet and aged to a T8 temper.

The alloys RX818 and ML377 are variants of the Al-Cu-Li-Mg-Ag Weldalite™ system, which is strengthened by fine distributions of thermally stable T_1 and θ' precipitates [2,8]. Ag and Mg were added in the RX818 chemistry to increase the volume fraction of strengthening precipitates, while Mn was added to ML377 as a substitute for Zr in dispersoid formation and to promote recrystallization [2]. Thermomechanical processing was adjusted for each chemistry to produce an unrecrystallized microstructure in RX818 and a recrystallized microstructure in ML377. Both RX818 and ML377 were cast in 4500 kg ingots, processed to 2.3 mm thick sheet using commercial processes, and aged to a T8 temper.

The alloy CM001 is an Al-Cu-Mg alloy that was developed from the British alloy RR58. RR58 is produced in France with the designation AU2GN; the U.S. equivalent alloy is 2618. The thermal stability of Al-Cu-Mg alloys is due to the presence of Al_9FeNi dispersoids [10]. The composition limits and thermomechanical processing of RR58 were modified to produce CM001 in product forms necessary for structural application on the Concorde [11,12]. The CM001 evaluated in this program was 4 mm thick Alclad sheet in the T6 condition.

The Russian alloy 1143 is an Al-Cu-Mg alloy that was developed from the Russian alloy AK4-1. AK4-1 is the material used on the TU144 [13,14], a Russian supersonic aircraft. The composition of AK4-1 is similar to RR58 and 2618. The composition of 1143 reflects reduction in Fe and Ni levels from AK4-1 to improve fracture toughness while retaining elevated temperature strength. The 1143 material evaluated in this program was 3.25 mm thick sheet in the T6 condition.

PROCEDURES

Tensile properties and fracture toughness were determined for each alloy in the T8 or T6 condition and after isothermal exposures at 93°C, 107°C, and 135°C for times up to 15,000 hours. Tensile and fracture properties were measured at room temperature for both as received (T8 or T6) and thermally exposed conditions. The materials were isothermally exposed as either machined specimens or as pieces of sheet from which specimens were machined after exposure. Properties were also evaluated for CM001 and 1143 in the T6 condition and after thermal exposure.

Tensile properties were measured using sub-size tensile specimens machined in the longitudinal (L) and transverse (T) orientations. Specimen thickness was the full sheet thickness for all alloys. All test procedures were in accordance with ASTM E8-96. For each test, material yield strength, ultimate tensile strength, percent elongation to failure, and elastic modulus were calculated.

Fracture toughness was evaluated for each alloy and thermal exposure combination for the LT orientation using compact tension (CT) specimens with W=50.8 mm. Specimen thickness was the full sheet thickness for all alloys. Fracture toughness was determined by the single specimen, J-integral method according to ASTM E1152-87. Physical crack length was determined by either the unloading compliance method or by the potential drop method. The resulting J-R curves were converted to K-R curves by Eq. 1,:

$$K_I \cong [J \times E]^{1/2} \quad (1)$$

where J is the J-integral value and E is Young's Modulus. Crack growth resistance, K_{J2mm}, defined as the R-curve value at 2 mm of physical crack extension, was evaluated for each alloy and exposure condition to assess thermal stability.

Additional tests were performed to estimate fracture toughness for the developmental alloys using middle crack tension (M(T)) specimens for material in the T8 condition and after 5,000 hours exposure at 93°C and 107°C. All procedures were in accordance with ASTM E561-94. The stress intensity factor, K, was calculated using Eq. 2,:

$$K= (P/BW)[\pi a \sec(\pi a/W)]^{1/2} \quad (2)$$

where P is load, B is thickness, W is width, and a is the half crack length. The M(T) specimens were tested in the LT orientation, and were nominally 56 cm wide and full sheet thickness. The specimens were fatigue precracked until the crack length, 2a, was approximately 20 cm long. Local out-of-plane buckling was restricted during the fracture test by using waffle pattern guide plates that covered 36 cm above and below the crack plane. Physical crack length was determined by unloading compliance and visual measurement. K_{app} was calculated using equation 2 based on maximum load ($P=P_{max}$) and the initial crack length in the fracture test.

RESULTS AND DISCUSSION

The strength-toughness behavior of C415, C416, ML377, and RX818 is illustrated in Figs. 1 and 2 for the T8 and selected post-thermal exposure conditions and compared with results for CM001 and 1143. Tensile and fracture results shown are for the L and LT orientations, respectively, and represent the average of duplicate specimens. All four developmental alloys show an improved strength-toughness combination compared to either baseline alloy in the pre-exposure conditions (T8/T6), reflecting higher strength and toughness than CM001, and higher strength than 1143 with comparable fracture toughness. Among the developmental alloys in the T8 condition, C415, C416, and ML377 exhibit comparable strength and toughness, with RX818 having 10-15% higher strength and 8-15% lower toughness.

Alloys C415, C416, and ML377 exhibit yield strength levels after exposure for 10,000 hours at 107°C that are equal to or greater than the T8 condition, although fracture toughness declined 12-18%. In addition, only C415 exhibits a significant toughness reduction with extension of the exposure time to 15,000 hours; C416 and ML377 are little changed (Fig. 1). No long term data is shown for RX818 because thermal exposures of this alloy were discontinued due to significant toughness degradation (>20%) observed after 3000 hours exposure [4]. The strength increase and toughness decrease (each approx. 10%) for 1143 after 10,000 hours exposure are similar to those of the developmental alloys. CM001, however, exhibits a small increase in toughness with no change in strength. The strength-toughness combination after thermal exposure remains higher for the developmental alloys compared to CM001 and reflects higher strength with comparable toughness compared to 1143. In general, the data trends described for the room temperature results after 107°C exposure are similar to trends observed after 93°C exposure and for post-exposure results measured at elevated and cryogenic temperatures [3,4].

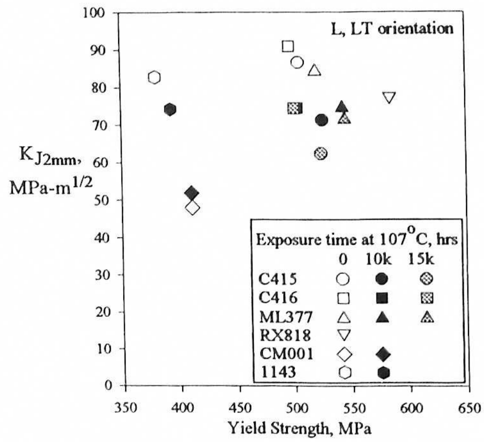


Fig. 1. Effect of 107°C thermal exposure on the yield strength and K_{J2mm} fracture toughness.

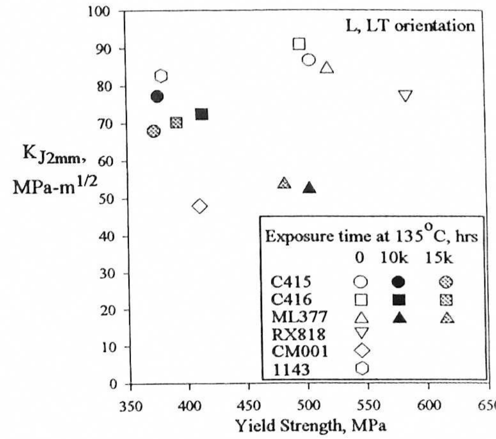


Fig. 2. Effect of 135°C thermal exposure on the yield strength and K_{J2mm} fracture toughness.

The effect of thermal exposure at 135°C is illustrated in Fig. 2 for the developmental alloys C415, C416, and ML377. Reductions in both strength and toughness of approximately 20% are observed for C415 and C416, after 15,000 hours exposure, resulting in a strength-toughness combination similar to 1143-T6. While the strength of ML377 is reduced by less than 10% after 15,000 hours exposure, the fracture toughness declines by approximately 35%, resulting in a strength-toughness combination that reflects higher strength with comparable toughness compared to that of CM001-T6.

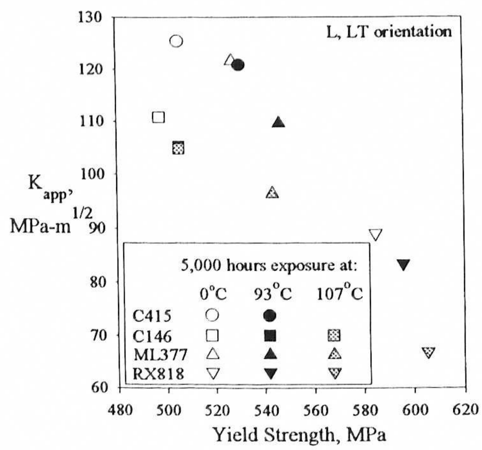
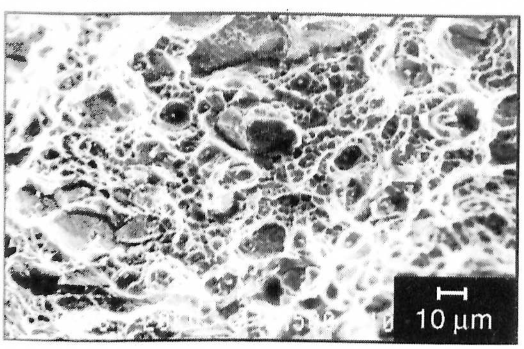


Fig. 3. Effect of thermal exposure on yield strength and K_{app} fracture toughness.

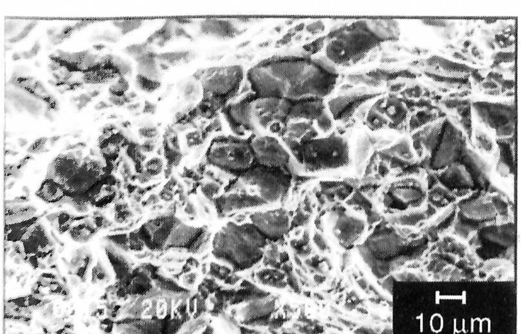
K_{app} fracture toughness determined from M(T) specimen testing for C415, C416, ML377, and RX818 is summarized in Fig. 3 for material in the T8 condition and after 5000 hours exposure at 93°C and 107°C. Fracture toughness results are for single specimens and tensile results are the average of duplicate specimens tested at each condition. All of the alloys exhibit increases in strength and decreases in toughness to varying degrees with thermal exposure. For C415 and C416, K_{app} is reduced by approximately 5% for both exposure temperatures while yield strength increased by no more than 5%. Alloys ML377 and RX818 show minimal increases in yield strength but progressive toughness degradation with increased exposure temperature, with losses in toughness increasing from 10% after 93°C exposure to more than 20% after 107°C exposure. The alloys C415, C416 and ML377 have 20-30% higher toughness and 10-20% lower strength than RX818 for the T8 condition and after 93°C exposure. The strength and toughness of ML377 after exposure for 5000 hours at 107°C are within 10% of the values for C416, but the changes for RX818 are more significant, reflecting 20% higher strength and 50% lower toughness than C416.

SEM fractography for all four developmental alloys indicated that the fracture mode was a mixture of transgranular microvoid coalescence (TG MVC) and intergranular (IG) fracture, with the relative proportions of each changing with thermal exposure condition. Representative fracture features observed for both C415 and C416 are shown in Fig. 4. Alloys C415 and C416 exhibit primarily TG MVC with very small amounts of IG fracture in the T8 condition, with the area fraction of IG fracture increasing to 20-30% with exposure at 93°C and 107°C, and to nearly 50% with 135°C exposure [5]. The equiaxed facets on the fracture surface shown in Fig. 4b reflect the underlying equiaxed, recrystallized microstructure of these alloys. TEM and DSC studies performed during the alloy development program determined that thermal exposure at 107°C resulted in dissolution of Ω in the matrix, coarsening of grain boundary Ω, and some precipitation of θ′ in C415 [6], while only mild coarsening of Ω was observed in C416. Coarsening of the grain boundary precipitates explains the increased occurrence of IG fracture with thermal exposure. C415 and C416 exhibit significant losses in both strength and toughness with exposure at 135°C. TEM examinations identified precipitation of S′ in C415 [6] and θ′ in C416 after 135°C thermal exposure, as well as coarsening of Ω on grain boundaries. The change in strengthening precipitate may explain the loss in strength, while grain boundary precipitate coarsening results in loss in toughness.

Alloys ML377 and RX818 exhibited mixed TG MVC and IG fracture in the T8 condition with

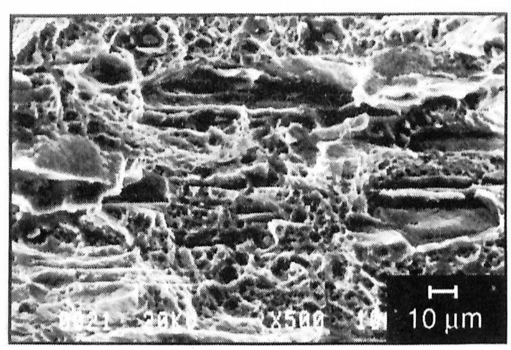


(a) C416; T8 condition.

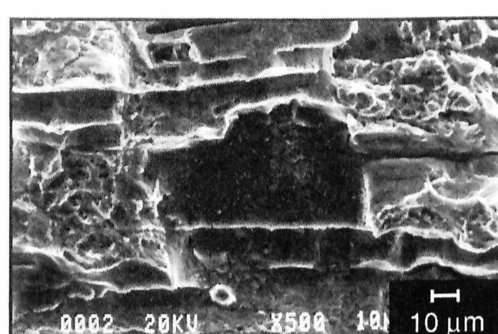


(b) C416 after 10,000 hours at 107°C.

Fig. 4. SEM fractographs for C416.



(a) ML377; T8 condition.



(b) C416 after 10,000 hours at 107°C.

Fig. 5. SEM fractographs for ML377.

increased IG fracture occurring after thermal exposure similar to C415 and C416. Representative fracture features observed for both ML377 and RX818 are shown in Fig. 5. The area fraction of IG was 30% for the T8 condition and increased to over 80% after 135°C exposure [1,5,7]. The elongated IG fracture facets shown in Fig. 5a reflect the pancake grain morphology of these alloys. The area fraction of IG fracture was higher in the T8 condition for RX818 and ML377 than for C415 and C416, suggesting that grain boundary precipitation was more substantial in these alloys prior to thermal exposure [5]. In other studies [1,7,8], TEM examination primarily identified coarsening of grain boundary T_1 with thermal exposure for RX818 and for ML377 at all exposure temperatures, with severe coarsening at 135°C. Precipitation of additional T_1 was also observed to a lesser degree with thermal exposure. Coarsening of grain boundary precipitates results in the losses in toughness, while the additional T_1 precipitation may explain the strength increases. Precipitation is preferential in ML377 and RX818 on selected grain boundaries aligned with the rolling direction [5]. The greater number of aligned grain boundaries in the unrecrystallized RX818 microstructure may explain its more rapid degradation in toughness as compared with ML377.

CONCLUSIONS

Results for all four developmental alloys indicated that tensile properties measured after 15,000 hours exposure at both 93°C and 107°C were equal to or greater than properties in the T8 condition, although results from the CT specimen tests indicated that fracture toughness declined approximately 20%. Reductions in toughness are explained by changes in fracture mode from transgranular microvoid coalescence to intergranular fracture, which is related to grain boundary precipitation and coarsening of grain boundary precipitates with thermal exposure. For both tensile and fracture toughness properties, there was little difference in the effect of exposure at 93°C and 107°C, but significant property degradation was observed for exposure at 135°C. All four developmental alloys exhibited significantly higher strength as compared to both CM001 and 1143 for the T8 condition and for thermal exposures up to 107°C, with comparable strength levels after 135°C exposure. Fracture toughness of C415 and C416 remained higher than that of CM001 for all thermal exposure conditions and was comparable to that of 1143 at the longest exposure times. ML377 retained toughness levels higher than CM001 and comparable to 1143 for thermal exposures up to 107°C.

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