

## EVALUATION OF CREEP BEHAVIOR OF EMERGING ALUMINUM ALLOYS FOR SUPERSONIC AIRCRAFT APPLICATIONS

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### ABSTRACT

The performance of four aluminum alloys was evaluated under moderately high stresses and warm temperatures to assess their creep behavior at supersonic service conditions. The alloys evaluated include two Al-Cu-Mg-Ag alloys and two Al-Cu-Li-Mg-Ag alloys. Isothermal creep tests were performed on these alloys at temperatures of 107°C and 135°C and at stresses from 210 MPa to 280 MPa. Test times ranged from 1000 hours to more than 10,000 hours. For comparison, similar tests were run on the Concorde alloy CM.001. A method for assessing the creep in this temperature and stress regime was developed to compare the four alloys and to extrapolate the data to anticipated lifetimes for supersonic aircraft. Over the temperatures, stresses, and times tested, all four developmental alloys remained in the primary creep regime and the creep curves were well described by the power law. In addition, over the range of test conditions considered, all four developmental alloys demonstrated superior creep resistance to the Concorde alloy. Transmission electron microscopy and existing microstructural data were used to review thermally-affected precipitates in specimens exposed to long-term isothermal exposure and creep as compared to specimens in the T8 condition to correlate the microstructural features with the creep behavior.

**keywords:** *creep, Al-Cu-Mg-Ag alloys, Al-Cu-Li-Mg-Ag alloys, microstructural analysis*

### 1. INTRODUCTION

Advanced aluminum alloys are prime candidates for potential structural applications on supersonic aircraft operating around Mach 2.0. In this operating regime, the temperatures are such that creep is of primary concern. The objective of this paper is to assess the creep properties of selected developmental alloys under potential service conditions anticipated for supersonic aircraft and to compare the performance of these alloys to that of the Concorde alloy. Four developmental aluminum alloys were tested to evaluate their creep performance and suitability for aircraft structural applications up to 135°C. Tests were conducted for times long enough to enable reliable extrapolation of short-term test data to lifetimes representative of service conditions. The alloys evaluated include two Al-Cu-Mg-Ag alloys and two Al-Cu-Li-Mg-Ag alloys which resulted from a program to develop aluminum alloys with improved strength/toughness combinations and good thermal stability [1]. The alloy used on the Concorde, CM.001, was also tested as a baseline.

The developmental alloys were creep tested at 210 and 280 MPa at 107°C, and at 210 MPa at 135°C. Test times ranged from 1000 to more than 10,000 hours, then the data was extrapolated to 50,000 hours for comparison with CM.001. The method for assessing the creep behavior to compare these four alloys and to extrapolate the data from relatively short term tests to longer lifetimes is discussed [2]. The response of the microstructures during thermal exposure and creep is discussed to develop a correlation between thermally-affected microstructural features and the creep behavior.

### 2. MATERIALS

Small additions of Ag to Al-Cu-Mg alloys with high Cu:Mg ratios significantly enhance strength and thermal stability [1]. C415 and C416 are Al-Cu-Mg-Ag alloys produced by Alcoa and based loosely upon the alloy chemistry of 2519 with the addition of Ag to promote the formation of  $\Omega$  ( $\text{Al}_2\text{Cu}$ ), which is a more effective strengthening precipitate and a more thermally stable precipitate than either  $\theta'$  ( $\text{Al}_2\text{Cu}$ ) or  $S'$  ( $\text{Al}_2\text{CuMg}$ ) [1].

Al-Li alloys have been shown to have high strengths and toughnesses, and excellent specific properties due to the addition of Li and the corresponding reduction in density. RX818 and ML377 are variants of the Weldalite™ (Al-Cu-Li-Mg-Ag) family of alloys produced by Reynolds Metals. The primary strengthening phases in both RX818 and ML377 are finely distributed T<sub>1</sub> (Al<sub>2</sub>CuLi) and S' [1,3]. RX818 was developed to have very high strength and toughness, and had an unrecrystallized microstructure. Mn was added to ML377 to encourage the formation of dispersoids and to promote a recrystallized microstructure [1,3].

In addition to these four developmental aluminum alloys, selective testing was performed on CM.001 for comparison. This alloy has several designations, including the Concorde alloy designation CM.001, the French designation AU2GN, the earlier British designation RR58, and the Aluminum Alloy designation 2618 [4]. CM.001 has relatively high Fe and Ni content to encourage the formation of a thermally stable microstructure, and a proprietary thermo-mechanical processing schedule which was developed specifically to meet the creep, fatigue, and strength goals of the Concorde design [5]. This alloy is strengthened by a fine dispersion of S' precipitates and Al<sub>3</sub>FeNi, an intermetallic phase which is virtually insoluble in aluminum [6].

The chemical compositions of the alloys tested are shown in Table 1; actual compositions are shown for the four developmental alloys, and nominal composition is shown for the CM.001. C415, C416, ML377, and RX818 were creep tested as 2.3 mm thick sheet in the T8 condition. The CM.001 tested in this program was procured as 3.2 mm thick alclad sheet in the T6 condition. The T6 condition was used for CM.001 on the Concorde because it was determined that stretching to achieve a T8 condition had a deleterious effect on the alloy's creep properties [5]. Equal amounts of material were ground from both sides of the CM.001 specimens to remove the alclad layer and to reduce the thickness of the sheet to 2.3 mm.

Table 1. Chemical compositions (wt %) of materials. (Balance = Aluminum)

| Alloy  | Cu  | Li   | Mg  | Mn  | Ag  | Zr   | Ti  | Ni  | Fe   | Si   |
|--------|-----|------|-----|-----|-----|------|-----|-----|------|------|
| C415   | 5.0 | --   | 0.8 | 0.6 | 0.5 | 0.13 | --  | --  | 0.06 | 0.04 |
| C416   | 5.4 | --   | 0.5 | 0.3 | 0.5 | 0.13 | --  | --  | 0.06 | 0.04 |
| ML377  | 3.5 | 0.95 | 0.4 | 0.3 | 0.4 | 0.12 | --  | --  | 0.05 | 0.05 |
| RX818  | 3.6 | 1.0  | 0.4 | --  | 0.4 | 0.14 | --  | --  | 0.07 | 0.04 |
| CM.001 | 2.3 | --   | 1.5 | --  | --  | --   | 0.1 | 1.1 | 1.1  | 0.2  |

### 3. PROCEDURES

Specimens were machined into a subsize tensile dogbone configuration (per ASTM E8) with a straight-sided gage section of 0.635 cm wide by 2.54 cm long. Specimens were instrumented with two thermocouples and back-to-back strain gages and installed in constant-load creep frames with wedge-pin grips. The specimens were loaded after heating to the test temperature. The temperature was maintained to within +/-2°C of the target temperature across the specimen and for the duration of the test. The first data point after the specimen was loaded was designated the starting point of the creep test. This initial elastic strain value was subtracted out of subsequent data so that the data reported is the creep strain. Tests were conducted at 210 and 280 MPa at 107°C and at 210 MPa at 135°C. Most tests were terminated after times between 1000 and 2000 hours, but selected tests were continued beyond 10,000 hours [2].

### 4. RESULTS

#### 4.1 Creep Data Analysis and Extrapolation

The creep behavior of the four developmental alloys at 210 and 107°C, 280 MPa and 107°C, and 210 MPa and 135°C is shown in Figs. 1, 2, and 3, respectively. The creep behavior of CM.001 is shown in Figs. 1 and 3. At times up to 1000 hours at 210 MPa and 107°C, RX818 exhibited the lowest creep strain, but the creep response of the four developmental alloys was tightly clustered

(Fig. 1). At longer times the creep response of ML377, C416, and C415 diverged only slightly. No long term data was obtained for RX818 because other experiments indicated that this alloy exhibited a significant loss of toughness and microstructural instability after long term isothermal exposure [7]. At 280 MPa and 107°C, the creep response of the developmental alloys remained clustered but showed somewhat more divergence than at the lower stress (Fig. 2). Near 1100 hours, ML377 had slightly lower creep strain than RX818 followed by C416 and C415. At 210 MPa and 135°C, the divergence of the creep curves for the developmental alloys is clearly evident (Fig. 3). Although the ranking of the alloys is the same as at 107°C, the creep resistance of the Li-bearing alloys is clearly better than that of the Al-Cu-Mg-Ag alloys at 135°C.

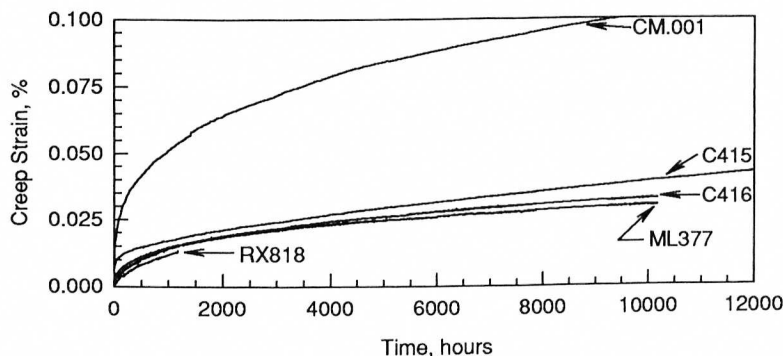


Figure 1. Creep curves at 210 MPa and 107°C.

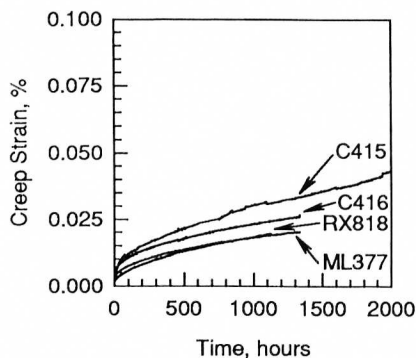


Figure 2. Creep curves at 280 MPa and 107°C.

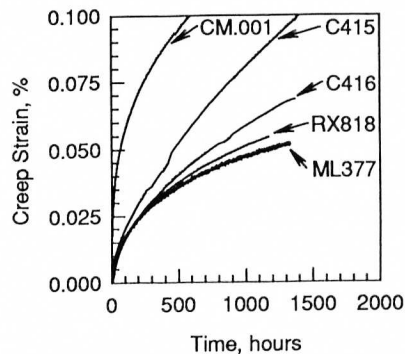


Figure 3. Creep curves at 210 MPa and 135°C.

All four developmental alloys demonstrate superior creep resistance to the CM.001 over all temperature and stress combinations tested. At 210 MPa and 107°C, the CM.001 creep strains were four to five times those of any of the four developmental alloys. Although the difference between the CM.001 and the four developmental alloys was somewhat less at 210 MPa and 135°C, the four developmental alloys still exhibited significantly lower creep strains than CM.001.

Further examination of the shapes of the creep curves shown in Figs. 1-3 reveals continuously decreasing creep rates at all of the temperature and stress combinations. Thus, the developmental alloys displayed primary creep behavior only, even after times exceeding 10,000 hours at 210 MPa and 107°C. Such behavior is readily described by the power law equation:

$$\varepsilon = a \cdot t^b \quad (1)$$

where  $\epsilon$  is the creep strain,  $t$  is the time, and  $a$  and  $b$  are empirically determined constants. The experimental data in this program were fitted to the power law and creep rates were calculated from the first derivative of the power law:

$$\dot{\epsilon} = a \cdot b \cdot t^{(b-1)} \quad (2)$$

where  $\dot{\epsilon}$  is the creep rate, and  $t$ ,  $a$ , and  $b$  are the same as in Eq. 1.

Eqs. 1 and 2 provide a means for extrapolating the experimental data to times representative of anticipated supersonic transport aircraft service life. Two extrapolation schemes were employed in an attempt to bound the problem of predicting the creep strain at very long service lives. In both schemes, the assumption was made that the material will not enter tertiary creep. In one scheme, the material was assumed to remain in primary creep over its entire exposure, and a power law extrapolation was performed using the parameters of the curve fitted to the experimental data. In the second scheme, the material was assumed to transition from primary creep to steady-state creep at 10,000 hours of creep exposure. In this case, the power law was used to calculate the strain at 10,000 hours using the parameters from the curve fitted to the experimental data, and linear extrapolation was performed from 10,000 to 50,000 hours using a constant creep rate, as calculated from Eq. 2 for a time of 10,000 hours. The arbitrary choice of 10,000 hours as the transition point from primary to secondary creep is believed conservative at 107°C because all experimental data at times this long (three alloys at 210 MPa and one alloy at 280 MPa) clearly show the materials to be in primary stage creep. This choice of transition point is probably less warranted at 135°C because the 210 MPa creep stress is 40-45% of the yield [8], and none of the tests at 135°C were continued to times this long. Extrapolation of the power law is less accurate when based upon shorter test times, but curve fitting to longer times refines the empirical power law parameters. Thus, extrapolations for cases where the experimental data is limited to significantly shorter time (1000 hours) will tend to over predict the actual strain.

Predicted creep strains after 50,000 hours exposure are shown in Table 2 for both extrapolation schemes. This time was chosen as representative of proposed supersonic transport service life and to be compatible with the 0.1% creep strain limit after 50,000 hours established for the Concorde [4]. At 210 MPa and 107°C, the predicted strains for ML377, C416, and C415 are 0.1% or less using both extrapolation schemes. At 280 MPa and 107°C, strains predicted for ML377 and C416 straddle 0.1% depending on the extrapolation scheme. However, the creep strains predicted for all the developmental alloys far exceeds the 0.1% limit at 210 MPa and 135°C. Thus, ML377, C416, and, to a lesser extent C415, would appear to have sufficient creep resistance at 107°C for application at reasonably high stresses. On the other hand, none of the developmental alloys would appear to be attractive for long term application at a temperature of 135°C.

Table 2. Predicted creep strains after 50,000 hours based on two extrapolation schemes.

| Alloy | 210 MPa / 107°C |              | 280 MPa / 107°C |              | 210 MPa / 135°C |              |
|-------|-----------------|--------------|-----------------|--------------|-----------------|--------------|
|       | Scheme 1*, %    | Scheme 2#, % | Scheme 1*, %    | Scheme 2#, % | Scheme 1*, %    | Scheme 2#, % |
| ML377 | 0.05            | 0.07         | 0.08            | 0.12         | 0.25            | 0.34         |
| RX818 | 0.12            | 0.15         | 0.14            | 0.18         | 0.32            | 0.43         |
| C416  | 0.06            | 0.08         | 0.09            | 0.13         | 0.54            | 0.71         |
| C415  | 0.07            | 0.10         | 0.18            | 0.25         | 0.94            | 1.2          |

\* power law to 50,000 hours

# power law to 10,000 hours, linear from 10,000 to 50,000 hours

#### 4.2 Microstructural Analysis

In the T8 condition, alloys C415 and C416 exhibit a finely dispersed  $\Omega$  phase and small amounts of  $\theta'$ . Because of its higher Cu and lower Mg content, C416 contains more  $\theta'$  but fewer  $\Omega$  precipitates than C415 in the T8 condition. Although  $\Omega$  is a more effective strengthening precipitate than either  $\theta'$  or S' and C415 has a higher amount of  $\Omega$  [1,3], the creep resistance exhibited by C416

is higher than that of C415. Other studies [9] have suggested that  $\Omega$  plates on  $\{111\}$  planes provide a more effective barrier to gliding dislocations than  $\theta'$  on  $\{100\}$  planes. However, TEM studies indicate that the creep behavior of C415 and C416 is influenced by structural changes during creep. Fig. 4 shows typical microstructures of these alloys before and after creep, with insets showing the corresponding selected area diffraction patterns. After nearly 1700 hours at 210 MPa and 135°C, the rod-shaped  $S'$  phase has been heterogeneously precipitated on dislocations at the expense of the  $\Omega$  phase in C415, but clusters of  $\theta'$  precipitated on dislocations with no reduction in the amount of  $\Omega$  in C416. Precipitation of the  $S'$  and  $\theta'$  phases was confirmed by energy dispersive x-ray analysis.

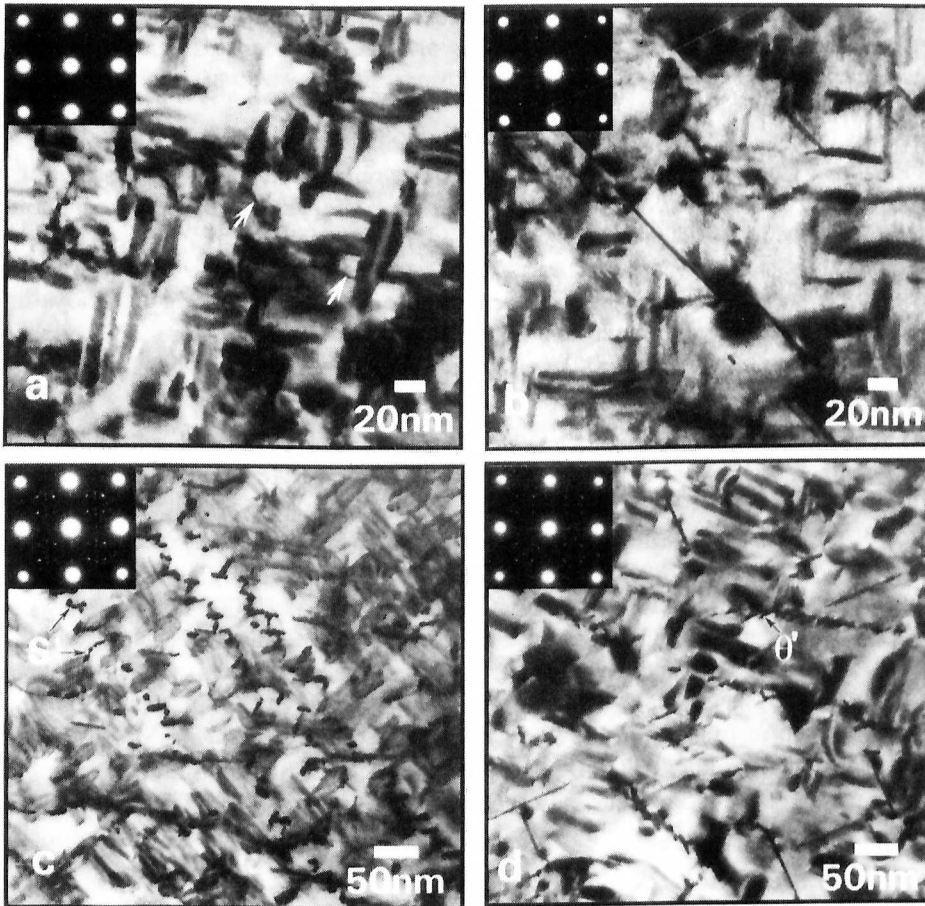


Figure 4. Bright-field electron micrographs and corresponding selected area diffraction patterns of (a) C415-T8, (b) C416-T8, (c) C415-T8 + 1700 hours at 210 MPa and 135°C, and (d) C416-T8 + 1700 hours at 210 MPa and 135°C. Electron beam is parallel to  $\langle 001 \rangle_{\alpha}$  zone axis.

Li and Shenoy [10] correlated TEM results with differential scanning calorimetry signatures relating to precipitation, dissolution, and coarsening reactions affecting the  $\Omega$ ,  $\theta'$ , and  $S'$  precipitates in C415 after long term isothermal exposures without stress. Their results show additional precipitation of  $\Omega$  and  $S'$  at shorter times (1000 hours) at both 107°C and 135°C. After longer times (3000 hours) at 135°C, appreciable  $\theta'$  precipitates and  $\Omega$  dissolves while  $S'$  is relatively unchanged. Whereas only mild coarsening of  $\Omega$  was observed after thermal exposure at 107°C for alloy C416. These results bolster the conclusion that C416 has better creep resistance than C415 because C416 retains more  $\Omega$ , but C415 precipitates  $S'$  and  $\theta'$  on dislocations in the matrix at the expense of  $\Omega$ .



For the Al-Li alloys RX818 and ML377, the primary strengthening precipitates consist of an extremely fine dispersion of  $T_1$  and  $S'$  in the T8 condition [1,3]. After 2000 hours exposure at 107°C and 135°C without stress, the size and volume fraction of  $T_1$  particles in RX818 increased significantly within the grains and on the grain boundaries, while the number density of  $T_1$  particles decreased, as compared to the T8 condition [1]. These effects are larger at 135°C than at 107°C, but the trends are similar at both temperatures. It has been shown [3] that the precipitation of  $T_1$  phase is similar between ML377 and RX818 after thermal exposure (1000 hours at 135°C). Creep data at the higher temperature and stress, and the slope of the creep curve and extrapolations at 210 MPa and 107°C indicate that RX818 has inferior creep resistance to ML377. This difference in creep behavior is attributed to the presence of the Mn dispersoids and the recrystallized microstructure in the ML377. Besides encouraging recrystallization, these dispersoids may also assist the creep resistance of ML377 by pinning dislocations.

It is expected that microstructural changes due to thermal exposure in both the Al-Cu-Mg-Ag and the Al-Li alloys would be similar with and without stress. Coarsening or dissolution of the primary strengthening precipitates, particularly in the matrix, is likely to cause a reduction in the creep resistance of these materials over time. Since all four developmental alloys have been shown to exhibit precipitation and coarsening of their primary strengthening phases after thermal exposure, the superior creep performance of the Al-Li alloys as compared to the Al-Cu-Mg-Ag alloys indicates that  $T_1$  is apparently a better strengthening phase for creep than  $\Omega$ . More detailed analytical microscopy is required to fully understand the creep behaviors observed.

## 5. CONCLUSIONS

Over the temperature and stress combinations tested, all four developmental alloys remain in primary creep for the duration of the tests performed, and demonstrate superior creep resistance to CM.001. The creep data up to 10,000 hours are well described using the power law, and subsequent extrapolations of the data at 210 MPa and 107°C indicate that the predicted creep strains will not exceed 0.1% in 50,000 hours for C415, C416, and ML377. However, at 135°C, all of these developmental alloys are predicted to far exceed 0.1% creep strain in 50,000 hours. ML377 exhibits the best creep resistance of the four developmental alloys, followed by RX818, C416, and C415, over the conditions tested. The superior creep performance of the Al-Li alloys as compared to that of the Al-Cu-Mg-Ag alloys indicates that  $T_1$  is apparently a better strengthening phase for creep than  $\Omega$ . The improved creep resistance of ML377 over RX818 is attributed to the Mn dispersoids and recrystallized microstructure in ML377. Due to precipitation of  $S'$  and  $\theta'$  on dislocations in the matrix at the expense of  $\Omega$ , C415 exhibits lower creep resistance than C416, which retains more  $\Omega$ .

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