THERMAL STABILITY AND CREEP BEHAVIOUR OF FOUR AL ALLOYS

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Abstract

The thermal stability and the creep behaviour of four aluminum alloys in sheet form (2048, 2650, 6013 and 8090) have been investigated. These alloys correspond to three various systems of hardening precipitation. When thermally exposed at 423 K, the 8090 alloy exhibits a strong microstructural evolution leading to dramatic variations of tensile properties. Accelerated creep tests have shown that in the tertiary stage, these four Al alloys behave in various ways. A tentative analysis of creep curves allows to suggest that the development of intergranular cavitation may vary from one alloy to another. Preliminary examinations of cavitation formed under severe creep conditions support this view.

Introduction

The work performed in the course of the Concorde programme has highlighted that resistance to creep is a criterion of prime importance for the choice of a fuselage metallic material [1]. Since the service life of the next supersonic transport aircraft is planned to be more than 4 times longer than the life of Concorde [2], this property has to be reinvestigated in the case of 2XXX Al alloys; the 2650 and the 2048 alloys have been developped from the 2618A and the 2024 alloys respectively, at the end of the Concorde programme, in order to offer an improved compromise between creep resistance and damage tolerance. Moreover, it is worth examining also some Al alloys which have been developped more recently: in this respect, the 6013 and 8090 are interesting materials because they correspond to two other systems of hardening precipitation.

This paper is devoted to a first examination of the creep behaviour of these alloys during accelerated creep tests. The influence of a prolonged exposure at 423 K on tensile properties as well as on microstructure is also reported.

Experimental

The alloys 2650 (nominally Al - 2.5 Cu - 1.5 Mg - Mn - Fe - Ni) and 2048 (nominally Al -

3.5 Cu - 1.5 Mg -Mn) were delivered by Rhénalu-Péchiney in the T8 temper; the 6013 alloy (nominally Al - 1 Mg - 1 Si - 1 Cu - Mn) was delivered by Alcoa in the T6 temper and the 8090 alloy (nominally Al - 2 Li - 1 Cu - 1 Mg) was delivered by Alcan in an underaged condition. All these alloys were in sheet form with sheet thickness ranging from 1.6 to 1.8 mm. They exhibit a recrystallised grain structure.

For the assessment of thermal stability, prolonged exposure at 423 K was performed in an oil bath where temperature was controlled to within ±1 K. Microstructural examinations were carried out by means of optical microscopy, scanning electron microscopy and transmission electron microscopy (TEM). Thin foils for TEM examinations were obtained in an usual way: they were thinned by double-jet electropolishing in a 2/3 vol. methanol - 1/3 vol. nitric acid solution. These foils were observed in a Philips CM 20 electron microscope operating at 200 kV.

Flat specimens for tensile testing were machined with the load axis parallel to the sheet long-transverse direction. Their thickness was 1.45 mm. An extensometer with a 20 mm. gauge length was used. Cross-head velocity was kept constant at 0.1 mm per minute. Creep flat specimens (gauge length: 56.8 mm, thickness: 1.4 mm) were machined with the load axis parallel to the rolling direction or to the long-transverse direction. Creep testing was performed on dead load machines. Temperature control was obtained by two thermocouples tightly attached to the specimen and the temperature was recorded all allong the test. Whatever the duration of the tests, temperature variations were less than ± 2 K. An extensometer fixed to the heads of the specimen allowed the elongation to be measured and recorded during the test. Test conditions were varied as follows: 448 K and 250 or 150 MPa, 423 K and 250, 220 or 200 MPa, 403 K and 250 MPa.

Thermal stability

Tensile properties were first determined at 293 K before and after a 1 000 or 5 000 hours thermal exposure at 423 K. The 0.2% proof stress and the elongation to rupture are plotted versus the exposure time in the Figures 1 and 2.

As illustrated by Figure 1, the 0.2% proof stress of the 8090 alloy undergoes large variations. A steep increase is observed after a 1 000 hours exposure. This increase may continue on for longer exposure times but it is exhausted before 5 000 h. This result is not surprising since this alloy was primarily underaged at the same temperature (423 K). TEM observations clearly showed that prolonged exposure at 423 K has allowed some coarsening of the δ' (Al_3Li) phase and the precipitation of the S' (Al_2CuMg) phase which was absent in the as-received material. This precipitation is likely to be mainly responsible for the strong increase of the proof stress [3]. As this alloy hardens, its elongation to rupture decreases (Figure 2). This effect is associated with a change of the rupture mode from transgranular to intergranular due to the precipitation of T2 particles within grain boundaries.

The 2650 and 2048 alloys exhibit a weak softening. These alloys which are hardened by a S'

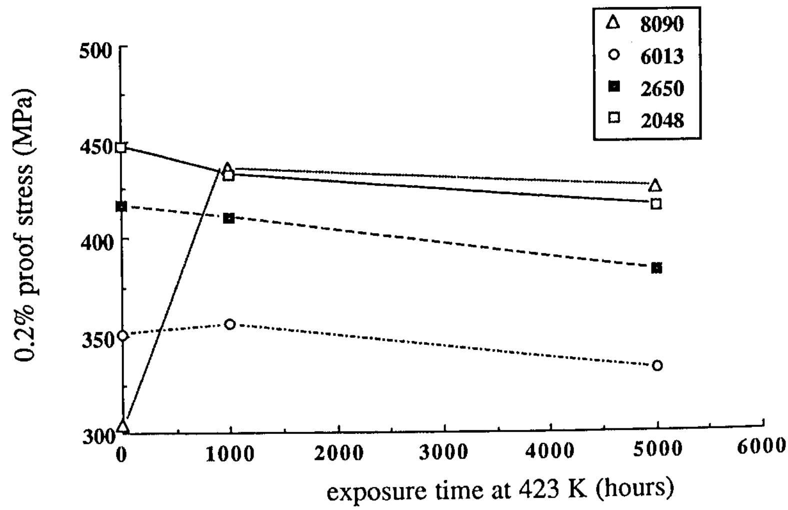

Figure 1. Tensile testing at 293 K: plot of 0.2% proof stress versus exposure time at 423 K. Lines are drawn for better understanding.

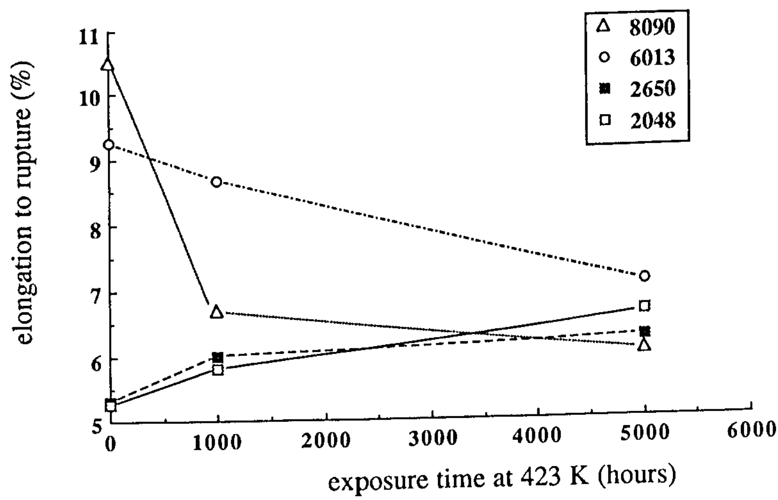

Figure 2. Tensile testing at 293 K: plot of elongation to rupture versus exposure time at 423 K. Lines are drawn for better understanding.

precipitation only, were artificially aged to peak strength; the prolonged exposure then consisted in an overaging. It can be noted that the coarsening kinetics seems to be rather low at 423 K. A weak increase in ductility corresponds to this softening (Figure 2).

The 0.2% proof stress of the 6013 alloy appears to be weakly sensitive to exposure times up to 5 000 hours. In the T6 temper, this alloy contains two types of hardening precipitates which have been proposed to be precursors of Mg_2Si-β and of the quaternary phase called Q or λ [4]. After a 5 000 hours exposure at 423 K, this dual precipitation was still observed but more precise examinations would be useful for a better description of this microstructure. In addition, a coarse precipitation of Q has occurred within some grain boundaries.

Additional tensile testing was performed at 423 K before and after thermal exposure at the same temperature. As far as the 0.2% proof stress is concerned, the same trends as those described before have been observed.

Creep behaviour

Creep curves obtained for all the various sets of conditions (σ and T) were compared with a particular attention paid to their shape in the tertiary stage. No ranking of the resistance to creep of the alloys will be proposed here since they are not comparable in a straightforward way: indeed, some microstructural parameters such as grain size are quite different.

As illustrated by the Figure 3 showing creep curves obtained at 423 K and 250 MPa, two main remarks can be done:

- the 8090 alloy always exhibits the lowest minimum strain rate. This may be related to the precipitation of S' during the creep test. This precipitation, which preferentially takes place at dislocations, possibly hinders movement of dislocations thus resulting in a lower minimum strain rate. This point needs further investigation.
- in the tertiary stage, creep curves differ from one alloy family to another. The 2048 and 2650 alloys spend a major part of their creep life in the tertiary regime and the elongation obtained in this regime is the highest for these 2 alloys. At the opposite, the 6013 alloy and principally the 8090 alloy show a far more limited strain-time trajectory. Although the creep conditions experienced in this study were severe, the understanding of these various behaviours is likely to be relevant for further modelling of the creep deformation.

Creep damage tolerance

The development of the tertiary regime can also be examined through a creep damage tolerance parameter such as λ: this parameter which was first suggested by Ashby and Dyson [5] is equal to $\varepsilon_r / \dot{\varepsilon}_{min} t_r$, where ε_r is the strain to rupture, t_r the time to rupture and $\dot{\varepsilon}_{min}$ the minimum strain rate. This parameter is more suited for alloys exhibiting a limited primary regime. In our case, a simple parameter λ' can be proposed in order to keep the same physical meaning as λ. Its calculation requires the primary regime to be suppressed: the best way consists in translating the zero of the deformation axis to the intersection of this axis with the

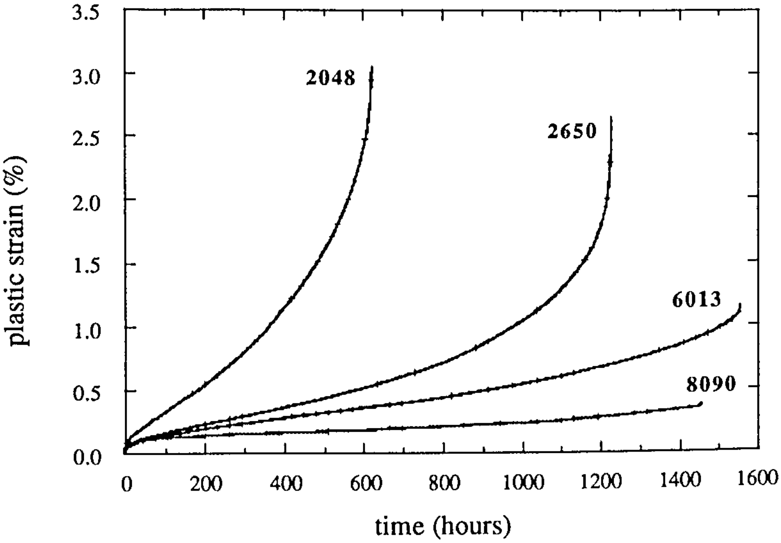

Figure 3. Creep curves obtained at 423 K and 250 MPa.

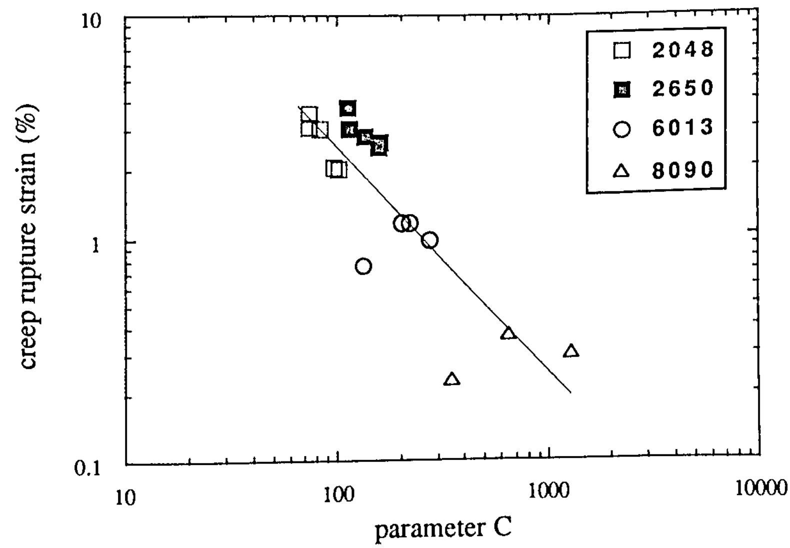

Figure 4. Plot of creep rupture strain versus parameter C. Data from all the creep tests performed at 423 K and 250, 220 or 200 MPa are used here.

minimum slope tangent to the creep curve. In this new coordinate system, the parameter λ' is calculated by the same formula as λ.

The hierarchy of λ' values is found to be consistent with the previous conclusions: at 423 K and for initial stresses ranging from 200 MPa to 250 MPa, λ' is roughly equal to 3 for the 2650 alloy, 2 for the 2048 alloy , 1.5 for the 6013 alloy and finally 1.3 for the 8090 alloy. At 403 K / 250 MPa and 448 K / 250 MPa this hierarchy is the same.

Examination of the log strain rate versus strain curves

The examination of the log strain rate versus strain curves gives an interesting result: in the major part of the tertiary regime, the strain rate varies exponentially with strain. In the log strain rate-strain curve, the slope of the corresponding straight line is referred as C.

Such a behaviour has been already observed [6] in Ni base superalloys. The parameter C has been considered to be the sum of three different contributions to damage [6, 7]: the intrinsic softening related to instabilities of the dislocation substructure, the loss of external section during a constant load test and the reduction of internal section due to intergranular cavitation. Only the last term depends on the fracture strain: it decreases with increasing fracture strain. It should be emphasized that this description does not take into account thermal effects such as transformation or simply coarsening of the hardening precipitates. Nevertheless, if the experimental values of C (for creep tests performed at 423 K) are plotted as a function of the fracture strains (Figure 4) it appears that, in such a plot, differences between the four alloys are clearly evidenced. This suggests that the various behaviours observed in the tertiary creep are related in some way to various developments of cavitation within these alloys.

A preliminary study of creep cavitation

First examinations of cavitation were performed on failed creep specimens in the neighbourhood of the fracture surface. Conditions selected for these tests were severe: 453 K and 230 MPa. Two alloys were compared, the 2650 and the 6013 alloys. The major result is that the geometry of intergranular cavities is not the same in these alloys (Figures 5 and 6). While in the 2650 alloy these cavities keep spherical, in the 6013 alloy they are crack-shaped. This seems to be the reason why fracture surfaces show different features. In the 2650 alloy, linking of the cavities is a ductile process and the fracture surface consists in a mixture of dimples and shear ligaments. The dimples correspond to the cavities and their size is roughly equal to the intergranular facet size. On the 6013 fracture surface, intergranular facets are clearly visible. Shear ligaments are also observed but their areal density is smaller than in the case of the 2650 alloy.

Conclusion

The four Al alloys under investigation can be distinguished by various degrees of thermal stability during thermal exposure at 423 K. The microstructural evolution of the 8090 alloy

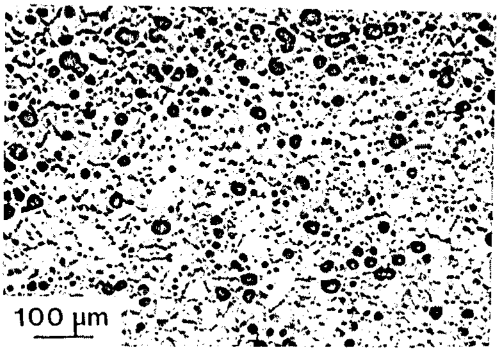

Figure 5. Cavitation observed near the fracture surface of the 2650 alloy (creep conditions: 453 K / 230 MPa).

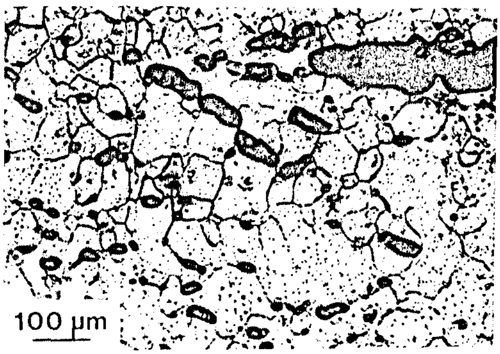

Figure 6. Cavitation observed near the fracture surface of the 6013 alloy (creep conditions: 453 K / 230 MPa).

leads to dramatic variations of its tensile properties such as 0.2 % proof stress and elongation to rupture. On the contrary, the 2650, 2048 and 6013 alloys show a rather low sensitivity to a prolonged exposure at 423 K.

Accelerated creep tests have shown that in the tertiary stage, the four alloys behave in various ways. In the major part of this stage, the strain rate varies exponentially with strain. This dependence allows a damage parameter to be derived. Experimental values of this parameter for the four alloys can be correlated with their creep rupture strains. This suggests that the development of intergranular cavitation may change from one alloy to another. Preliminary examinations of cavitation formed under severe creep conditions support this view.

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