Creep Behaviour of a γ Ti – Al Intermetallic Alloy

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Abstract: *The creep behaviour of a 1 mm thick sheet of a Ti – 46.5 Al – 4(Cr, Nb, Ta, B) (at%) in the 600 – 800º C has been studied. Specimens were machined in both the longitudinal and long transverse orientation of the sheet. No failure was produced, even after 2000 hours and under stresses that represented 85% of the yield strength, in the specimens at 600º C and no sign of cavities was found in their whole gauge length. A significantly better creep performance of transverse specimens was observed in the whole temperature range. At higher temperatures specimens failed after relatively short period times and in longitudinal specimens tested at 800º C and the higher stress level no real steady state creep was observed. Larson and Miller parameter has been used to fit experimental results and good correlations were found.*

INTRODUCTION

Titanium and its alloys began to be used in the early 50s due to their excellent properties of strength to weight ratio, higher than other available candidate materials, suitable toughness, excellent corrosion resistance and good fatigue properties which made them attractive for aeronautical applications. However, the state of the art at the end of the eighties limited their application of the most advanced titanium alloys to around 600º C because of strength and oxidation considerations (1). Even if a large effort has been dedicated since then to overcome these limitations no significant advance has been achieved in the performance of these alloys, and the need of new materials for higher temperatures is evident. A new class of material, the titanium aluminides, received a large attention in the nineties and are considered suitable candidates for their use in the next generation of aero engines, space launcher vehicles with re-entry capabilities and space platforms.

These new materials are not solid solutionable alloys but intermetallic compounds having a fairly fixed composition. There are two classes of titanium aluminides. The first ones, termed α_2 alloys, has the composition Ti_3Al, and possess useful structural properties up to 700º C. The second ones, the so called γ titanium aluminides, are based on TiAl and are aimed for even higher temperatures. The last ones as binary alloys, with aluminium contents of 48% show low elongation to fracture at room temperature, inadequate creep strength and poor oxidation resistance. In order to avoid these disadvantages the γ titanium aluminide is alloyed with approximately 2 at% of chromium, manganese or vanadium, added to increase room temperature ductility and/or 2-4% at% of niobium, tantalum, molybdenum or tungsten and, in more recent developments additions of up to 1 at% silicon or boron, which are all them added improve the high temperatures properties of creep strength and corrosion resistance (2). Moreover, boron is intentionally added to these alloys to refine their grain size.

γ titanium aluminides can be heat treated to obtain four types of microstructures; near gamma, duplex, nearly lamellar and fully lamellar. The near gamma microstructures are obtained by an annealing heat treatment at temperatures just above the eutectoid temperature. The duplex microstructure is typically the finest one and is produced after annealing treatments at temperatures where gamma and alpha phases are approximately equal by volume. Heat treatments at temperatures near the alpha transus produce nearly fully lamellar microstructures. Finally, three types of fully lamellar structures have been identified, depending on the heat treating conditions but all them are formed by alternating plates of alpha and gamma, having the usual crystallographic orientation relationships (3). The alpha phase lamellae that remain during cooling subsequently transform to an ordered α_2 phase at temperatures below the eutectoid temperature (4). Maximum ductility at room temperature is achieved with the duplex microstructure but fully lamellar microstructures are preferred to obtain the best creep resistance (5).

These improved materials exhibit very attractive properties that can be summarized as: 15% lighter than titanium alloys and half of the density of nickel base superalloys, higher specific strength and higher specific Young modulus than these materials in the whole application range and good oxidation and burn resistance. These properties make them suitable for the thermal protection systems and hot structures of spacecraft but before that these materials could be introduced in space vehicles an evaluation of their creep performance is needed.

The aim of this paper is to assess the creep behaviour of a 1 mm thin sheet of a Ti – 46.5 Al – 4 (Cr, Nb, Ta, B) (at%) alloy in the 600 – 800º C temperature range.

EXPERIMENTAL PROCEDURE

Material chosen for this study consisted in a 1 mm thin sheet of a Ti – 46.5 Al – 4 (Cr, Nb, Ta, B) (at%) alloy. This material was vacuum melted, hot isostatic pressed to close cast porosity, homogenized in the alpha phase field to minimize microsegregations, hot rolled and heat treated slightly above the alpha transus to obtain a fully lamellar microstructure.

Creep tests were carried out on specimens machined in the longitudinal and long transverse orientations of this sheet. These tests were performed in air, in the 600 – 800º C temperature range and using various stress levels, according to ASTM E-139 standard (6). Some tests were duplicated to evaluate the scatter in the results. Strain was continuously recorded during the tests by means of an extensometer fixed to the specimen and strain rate calculate to determine the minimum (steady state) creep rate. These tests were performed until failure or were stopped when a previously established time was reached. This time was initially fixed in 100 hours although it was later increased to longer times (up to 250 and even to 2000 hours) to analyze the creep damage induced in the material at the different temperatures and stress levels.

After failure one half of each broken specimen was used for a fractographic study carried out by scanning electron microscopy on their fracture surfaces. Moreover, metallographic samples were obtained from longitudinal sections of the other halves, including their fracture surfaces. These samples were observed in the optical and scanning electron microscope to relate the microstructural facets with the failure operating mechanism. Finally, with the aim of detecting the presence of creep cavities and evaluating the level of damage induced in those specimens that passed the test without failure, a thorough examination of the whole gauge length by optical and scanning electron microscopy was performed.

RESULTS AND DISCUSSION

Table 1 summarizes the results obtained in the creep tests performed at the different temperatures and stress levels.

TABLE 1: Creep tests results

T ºC	Stress (MPa)	Orientation	Time to failure (h)	T ºC	Stress (MPa)	Orientation	Time to failure (h)
800	275	Longitudinal	4.4	800	275	Transverse	10.2
800	275	Longitudinal	6.6	800	275	Transverse	14.7
800	250	Longitudinal	23.8	800	250	Transverse	24.7
800	250	Longitudinal	5.5	800	250	Transverse	18.6
800	200	Longitudinal	19.0	800	200	Transverse	102.2
800	200	Longitudinal	59.5	800	200	Transverse	136.5
775	275	Longitudinal	9.7	775	275	Transverse	57.4
775	250	Longitudinal	32.6	775	250	Transverse	98.7
750	275	Longitudinal	33.0	750	275	Transverse	158.4
750	250	Longitudinal	52.3	750	250	Transverse	228.6
700	275	Longitudinal	97.0	700	275	Transverse	2298
700	275	Longitudinal	227.2	700	275	Transverse	> 250
700	250	Longitudinal	> 250	700	250	Transverse	> 250
700	200	Longitudinal	> 250	700	200	Transverse	> 100
600	275	Longitudinal	> 2000	600	275	Transverse	> 2000
600	275	Longitudinal	> 2000	600	275	Transverse	> 2000
600	250	Longitudinal	> 100	600	250	Transverse	> 100
600	200	Longitudinal	> 100	600	200	Transverse	> 100

It can be easily observed by comparison of the results at the same row that although there is a certain scatter in the results, transverse specimens exhibit a better creep performance than longitudinal ones tested in the same stress and temperature conditions. The same trend has been also reported in other recent works (7 – 8) where the mechanical properties of a similar alloy with primary annealed or fully lamellar microstructures were investigated. This anisotropy was attributed to the existence of a modified cube texture in the primary annealed material, where the long axes of the tetragonal unit cells are aligned in the transverse direction and the short axes parallel to the rolling direction (7). However, in a fully lamellar material, very similar to that studied in the present work, a different explanation was given. It was speculated that a preferred lamellae orientation against the sheet plane and rolling direction occurred and anisotropy in the properties was associated with these microstructural features (8). Nevertheless, in the present study, no clear differences were detected in the microstructure of the longitudinal and transverse sample. Consequently, further analysis is needed before a conclusion about the origin of this anisotropy could be reached.

No failure was observed in those specimens tested at 600º C, even after 2000 hours and under stress levels that represent 85% of the yield strength at this temperature. Moreover, metallographic analysis of these specimens did not reveal any sign of creep cavities in the whole gauge length. The creep performance of the alloy at this temperature is significantly superior to those exhibited by the titanium alloys which can hardly operate in these conditions and can be considered to substitute the heavy nickel base alloys in some components.

Most of the specimens tested at higher temperatures failed during the test, mainly those machined in the longitudinal direction. As expected a decrease in the failure time and higher secondary creep rates with an increase in the testing temperature and/or applied stress level was observed. The values of the logarithm of the time to failure versus the inverse of the testing temperature were plotted and a regression a line drawn according to the method proposed by Larson and Miller. A good correlation between tests results and predictions exists as can be seen in figures 1 and 2 that exhibit these plots for the longitudinal and transverse specimens, respectively. It must be pointed out, nevertheless, that the value of the constant in the obtained for the longitudinal specimens (around 12) was lower than those reported for most metallic materials. This low value could be associated with the inclusion in the graph of the data corresponding to 800º C tests, where failures were produced after very short times and a certain influence of oxidation on the failure could be claimed. The value of this constant for the transverse ones (around 20) is in better agreement with the expectations and supports the hypothesis that very short failure times introduced a bias in the real value of the longitudinal ones.

Metallographic examination of these specimens revealed the presence of voids not only near the fracture surface but also relatively far away from it. Voids tend to elongate in the transverse direction to the loading axis. This is characteristic of creep cavities and support the hypothesis that they are real creep cavities and no voids produced by the plastic deformation. Scanning electron microscope examination of the fracture surfaces exhibited some localized flat zones but most of them showed an intergranular morphology with alternating lamellae of α_2 and γ phases.

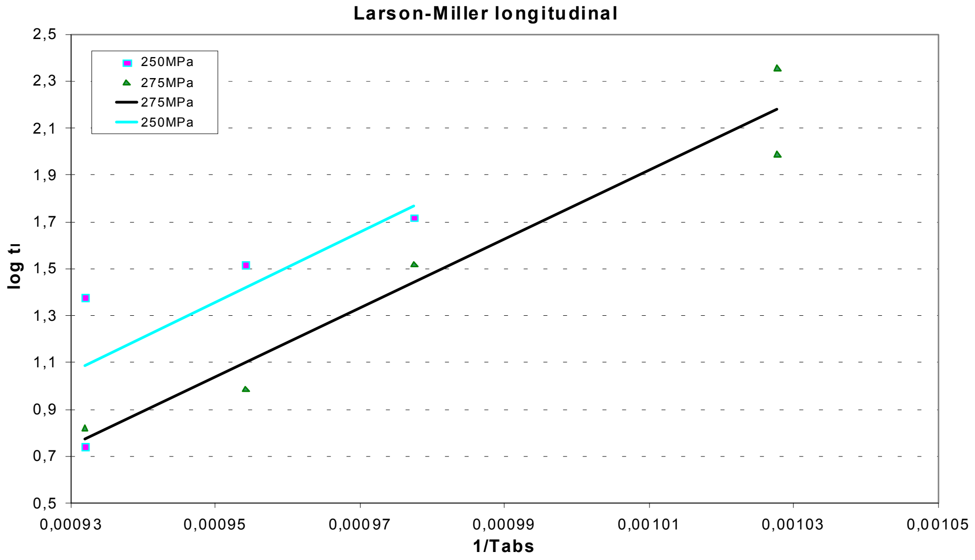

Figure 1: Plot of the logarithm of the time to failure versus the inverse of the testing temperature (longitudinal specimens)

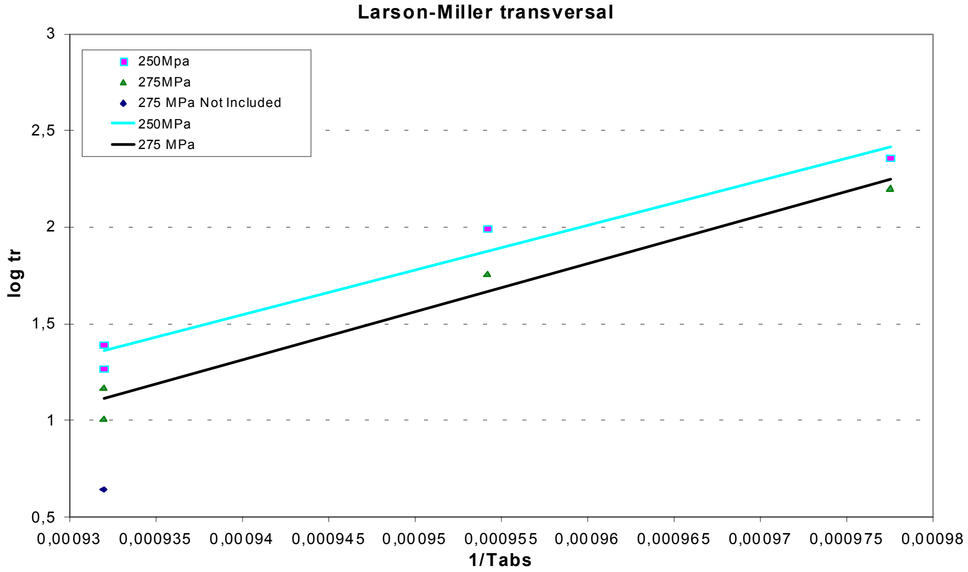

Figure 2: Plot of the logarithm of the time to failure versus the inverse of the testing temperature (transverse specimens)

Longitudinal specimens tested at 800º C failed after a very short period of time, even when tested under the lowest stress level. Transverse ones lasted longer times but hardly pass 100 hours. Moreover, in those tested at the highest stress levels no real steady stress creep was observed. Fractographic examination of these specimens revealed that the whole fracture surface was covered by a thick oxide layer. The morphology of these fracture surfaces seems very similar to that reported by other authors (9), where was pointed out that the oxidation resistance of unimplanted was a concern above 700º C and particularly poor above 830º C. Energy dispersive spectrometry of this layer is formed by aluminium and, mainly, titanium oxides. The formation of an intermixed scale of titanium and aluminium oxides was reported in another paper having limited its protective effect only to about 750 – 800º C (10), that is the temperature used for these tests.

CONCLUSIONS

a. Specimens tested at temperatures below 700º C did not failed even after 2000 hours and no damage was detected in their whole gauge length. The creep performance of the alloy in the range of temperatures between 600 and 700º C is clearly superior to titanium alloys which can hardly operate in these conditions.

b. Transverse specimens exhibited longer creep lives than longitudinal ones in the whole studied range of temperatures and stresses. This anisotropy could be associated with the crystallographic texture of the sheet or with some microstructural features (preferred orientation of the lamellae) but further research is needed before a conclusion could be reached.

c. Good correlations were found between experimental results and the predictions obtained using the method proposed by Larson and Miller although a value of the constant lower than expected was obtained for the longitudinal specimens.

d. Specimens tested at 800º C failed after a short period of time and no real steady state creep was observed in some tests. Fractographic examination reveal that fracture surfaces are covered with a layer of oxide pointing towards a significant contribution of oxidation in the failure process.

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