Room-temperature low-cycle fatigue behaviour of Ti–27Al–15Nb alloy

P N SINGH, C RAMACHANDRA, B K SINGH and V SINGH

Department of Metallurgical Engineering, Banaras Hindu University, Varanasi 221 005, India

MS received 4 May 1994; revised 25 January 1995

Abstract. Low-cycle fatigue (LCF) behaviour of the alloy Ti–27Al–15Nb, in (α_2 + B2) heat-treated condition was studied in total axial strain control mode at different total strain amplitudes ($\Delta\epsilon_t/2$) from ± 0.65 to $\pm 1.0\%$ and room temperature. While there was little hardening of the material at the lowest strain amplitude ($\Delta\epsilon_t/2: \pm 0.65\%$), pronounced hardening was observed at the higher strain amplitudes ($\Delta\epsilon_t/2 \geq 0.83\%$). The cyclic stress, at high strain amplitudes, continuously increased from the beginning till fracture of the specimens. The LCF resistance of the material was found to be low and this was due to its poor ductility at room temperature. Dual slope was observed in the Coffin–Manson plot, with less slope of the upper segment than that of the lower one, as observed in several alloys. The fracture behaviour pointed to brittleness and faceted features were observed.

Keywords. Titanium aluminide; low-cycle fatigue; cyclic hardening; fatigue life; dual slope; C–M plot.

1. Introduction

Titanium aluminides possess attractive properties like high creep resistance (MacAndrew and Kessler 1956; Blackburn and Smith 1981), high temperature-specific strength and sufficient oxidation resistance (MacAndrew and Kessler 1956) and retain high static strength and stiffness at elevated temperatures (Lipsitt *et al* 1975). These characteristics make them potential candidate materials for applications in gas turbine engines and airframe structures in advanced aircraft. The use of stoichiometric Ti_3Al and TiAl alloys is limited due to their poor ductility at ambient temperature (Sastry and Lipsitt 1977a, 1980; Lipsitt *et al* 1980; Baker and George 1992). However, the addition of Nb to Ti_3Al has been found to improve its ductility through stabilization of another phase, B2 (based on the high-temperature bcc allotrope of Ti), and thus rendering the single phase Ti_3Al to a dual phase (α_2 + B2) system. Investigations have been carried out on the influence of morphology of the constituent phases α_2 and B2 on tensile properties and fracture behaviour of Ti_3Al –Nb alloys (Gogia *et al* 1992; Kim *et al* 1993). Reports are also available on creep behaviour (Mediratta and Lipsitt 1980; Hayes 1989; Mishra and Banerjee 1990), fracture toughness and fatigue crack growth (Soboyejo 1992) of the titanium aluminides. Sastry and Lipsitt (1977b) studied the cyclic deformation behaviour of Ti_3Al over a range of temperature and analysed the process of cyclic hardening. However, little work has been carried out on low-cycle fatigue (LCF) behaviour of Ti_3Al –Nb alloys. The present work is concerned with characterization of LCF behaviour of the Ti–27Al–15Nb alloy in the α_2 + B2 heat-treated condition at room temperature.

2. Experimental

Hot rolled rods of 15 mm dia of the alloy Ti–27Al–15Nb were procured from the Defence Metallurgical Research Laboratory (DMRL), Hyderabad. Blanks of 12 mm dia were

vacuum-encapsulated in silica tubes with titanium getter and solution-treated in the $\alpha_2 + \text{B2}$ phase field at $1060 \pm 5^\circ\text{C}$ for 4 h and cooled at a rate of $\approx 288^\circ\text{C/h}$. The specimen for optical microscopy was prepared by mechanical polishing and etching with a solution of 10% HF + 5% HNO_3 + 85% H_2O (by volume). Tensile specimens with double shoulder having gauge length and gauge diameter 11 and 3.25 mm respectively were machined from the heat-treated blanks and tested using Instron universal testing machine. LCF specimens with threaded ends, having gauge length and gauge diameter 15 and 5.5 mm respectively and shoulder radii of 20 mm, were machined from the heat-treated blanks. The gauge section was mechanically polished up to 4/0 grade of emery paper. LCF tests were conducted in tension–compression ($R = -1$), under total axial strain control, in the range of total strain amplitude ($\Delta\epsilon_t/2$) from $\pm 0.65\%$ to $\pm 1.0\%$ at constant frequency of 0.2 Hz, using a computer-controlled servohydraulic MTS test facility. The variation of stress amplitude ($\Delta\sigma/2$) and inelastic strain amplitude ($\Delta\epsilon_p/2$) in tension and compression with number of cycles was recorded. Fracture characteristics were examined using a JEOL 840A scanning electron microscope.

3. Results and discussion

The optical microstructure of the alloy in the $\alpha_2 + \text{B2}$ heat-treated condition shows primary α_2 phase isolated in the B2 matrix (figure 1). The microstructure is seen more clearly at higher magnification obtainable in the scanning electron micrograph shown in figure 2. The material in the $\alpha_2 + \text{B2}$ treated condition was found to have yield strength 609 MPa, fracture strength 759 MPa and elongation 3.1%. The LCF data at different total strain amplitudes ($\Delta\epsilon_t/2$) in terms of the plastic strain amplitude ($\Delta\epsilon_p/2$), elastic strain amplitude ($\Delta\epsilon_e/2$) and cyclic stress amplitude ($\Delta\sigma/2$) at first cycle (N_1), half life ($N_f/2$) and the last cycle to failure (N_f) along with the

Figure 1. Optical micrograph showing the microstructure, consisting of isolated α_2 and transformed B2 phases.

Figure 2. Scanning electron micrograph showing α_2 and transformed B2 phases.

Table 1. Low-cycle fatigue data for the alloy Ti-27Al-15Nb at room temperature.

$\Delta\epsilon_i/2$ (%)	$\Delta\epsilon_p/2$ (%) at			$\Delta\epsilon_e/2$ (%) at			$\Delta\sigma/2$ (MPa) at			Cycles to failure
	N_1	$N_f/2$	N_f	N_1	$N_f/2$	N_f	N_1	$N_f/2$	N_f	N_f
0.83	0.242	0.106	0.100	0.588	0.724	0.730	640.2	771.6	777.8	1724
0.84	0.250	0.175	0.160	0.590	0.665	0.680	643.9	746.9	766.6	268
0.85	0.258	0.200	0.201	0.592	0.645	0.649	646.8	739.0	762.6	54
0.88	0.280	0.250	0.216	0.600	0.630	0.664	650.0	731.7	760.1	32
1.00	0.347	0.330	0.321	0.653	0.670	0.679	653.5	664.5	675.1	4

number of cycles to failure (N_f) are presented in table 1. It may be seen from table 1 that while $\Delta\epsilon_p/2$ continuously decreases, ($\Delta\epsilon_e/2$) and ($\Delta\sigma/2$) steadily increase from N_1 to N_f and thus there is continuous hardening of the material during cyclic loading. It may, however, be noted that changes in these parameters ($\Delta\epsilon_p/2$, $\Delta\epsilon_e/2$ and $\Delta\sigma/2$) are relatively more during the first half of the life than those in the later half. The cyclic hardening behaviour at different strain amplitudes may be seen from the stress response curves shown in figure 3. There is little hardening at the lowest strain amplitude ($\Delta\epsilon_i/2: \pm 0.65\%$). However, it is relevant to mention that this test was not continued till fracture but was interrupted after 10632 cycles. On the other hand there is pronounced hardening at the higher strain amplitudes ($\Delta\epsilon_i/2 \geq 0.83\%$). The cyclic hardening at higher strain amplitudes shown in figure 3 is similar to that observed by Sastry and Lipsitt (1977b) in Ti_3Al below $700^\circ C$. They have analysed the process of cyclic hardening in Ti_3Al and related the observed behaviour to pronounced planarity of slip and the interaction of a dislocations with dislocation

Figure 3. Cyclic stress response curves showing variation of stress amplitude with number of cycles.

debris of $\frac{1}{3}\langle 11\bar{2}3 \rangle$ Burgers vector. The extent and, in particular, the rate of cyclic hardening in the alloy Ti–27Al–15Nb (figure 3) is relatively lower than that in Ti_3Al (Sastry and Lipsitt 1977b) and may perhaps be related to different dislocation behaviour in the α_2 and B2 phases of the alloy Ti–27Al–15Nb in contrast to that in the single phase α_2 in Ti_3Al .

Cyclic hardening in the alloy Ti–27Al–15Nb in the present investigation is also evident from the cyclic and monotonic stress–strain curves (figure 4). Stress and strain values of the cyclic curve correspond to the cycle close to the half life ($N_f/2$). The drop in the cyclic stress at the highest strain amplitude ($\Delta\epsilon_t/2: \pm 1.0\%$) may be due to early crack initiation in the specimen tested at the highest strain amplitude to cause its failure in four cycles.

The variation of fatigue life with different components of the strain amplitude is shown in figure 5. It may be noted that the values of $\Delta\epsilon_p/2$ and $\Delta\epsilon_e/2$ used in figure 5 correspond to the cycle close to $N_f/2$. The variation of reversals to failure ($2N_f$) with $\Delta\epsilon_t/2$ is linear. However, there is mild tendency for dual slope in the elastic strain plot. While the slope of the segment corresponding to high-strain region is negative, that of the low-strain region is positive (figure 5, table 2). A very weak tendency of bilinearity in the elastic strain plot has earlier been reported in a high-temperature ($\alpha + \beta$)-type titanium alloy Ti–6.6Al–3.3Mo–1.9Zr–0.3Si (Varma *et al* 1993); however, the slopes of the segments, corresponding to high- as well as low-strain region, were negative. Bilinearity in the elastic strain plot in the present investigation may be related to the

differences in cyclic work hardening behaviour of the material at low- and high-strain amplitudes. The Coffin–Manson (C–M) plot ($\Delta\epsilon_p/2$ vs $2N_f$) displays a distinct bilinear behaviour (figure 5). The slopes of the segments I and II are -0.132 and -0.219 respectively. These slopes are quite different from the usual slope of ≈ -0.5 observed

Figure 4. Cyclic and monotonic stress-strain curves.

Figure 5. Variation of number of reversals to failure ($2N_f$) with elastic, plastic and total strain amplitudes.

Table 2. Low-cycle fatigue parameters for alloy Ti-27Al-15Nb at room temperature.

LCF parameters	Segment I	Segment II
ϵ'_f	0.00434	0.00664
c	-0.132	-0.219
σ'_f	766.7	528.58
b	-0.0319	0.047

in many alloys. Bilinearity in C-M plot is reported in a number of alloys such as dual-phase steel (Mediratta *et al* 1986), Al-Li alloy (Eswara Prasad *et al* 1989), Nimonic PE16 (Singh *et al* 1991) and VT9 (Singh *et al* 1994). The occurrence of bilinearity in the C-M plot has been attributed to different factors like change in the deformation behaviour through work hardening characteristics (Mediratta *et al* 1986), variations in the fracture morphology with increasing strain amplitude (Coffin 1971; Eswara Prasad *et al* 1989), environmentally assisted fatigue degradation (Coffin 1972; Sanders and Starke 1976), and change in the microstructural mode of deformation from low to high strain amplitude (Singh V *et al* 1991; Singh *et al* 1994). Recently Radhakrishnan (1992), has analysed the bilinear behaviour of the C-M plot in many ferrous and non-ferrous alloys and rationalized that the metallic materials, with cyclic stress range corresponding to one million cycles below the cyclic yield strength range of the material, are likely to exhibit a bilinear C-M relationship. However, the prediction of dual slope in aluminide in the present investigation cannot be made on the above lines in view of the fatigue life of the specimens being far less than one million cycles. As seen from figure 5, the segment I is less steeper than segment II, hence estimation of fatigue life from extrapolation of either of the segments would lead to overestimation of fatigue life. Thus it is essential to evaluate fatigue life of such materials, exhibiting dual slope in C-M plot of the type shown in figure 5, over a wide range of strain amplitudes. It may be seen from figure 5 that the level of the plastic strain amplitude plot is much lower than that of the elastic one even in the region of high strain amplitude and hence the LCF resistance of this material appears to be quite low. The poor LCF resistance could also be visualized in terms of extremely low value of the fatigue ductility coefficient (ϵ'_f) (table 2). The transition fatigue life, estimated from extrapolation of even the segment II of the elastic and plastic (C-M) plots is quite low (< 2 cycles) (figure 5). The poor LCF resistance of this material may thus be attributed to its extremely low ductility at room temperature. On the other hand the high-cycle fatigue resistance appears to be high, as reflected by high value of the fatigue strength coefficient (σ'_f) and low value of fatigue strength exponent (b) (table 2). The fracture characteristics of this material, tested in fatigue, are observed to be considerably different from

Figure 6. Scanning electron micrographs showing fracture characteristics of the LCF tested sample ($\Delta\epsilon_t/2 = 0.83\%$, $N_f = 1724$ cycles). (a) Fracture surface at low magnification, (b) fatigue striations in peripheral region and (c) cleavage-type fracture in central region of the fracture surface.

those of the ductile alloys. The fracture surface does not show flat region, typical of fatigue failure, at low magnification (figure 6a). Fatigue striations are observed at higher magnification in some circumferential regions, to a very limited depth from the surface (figure 6b). The fracture surface in general shows cleavage type of failure (figure 6c).

4. Conclusions

The alloy Ti–27Al–15Nb in α_2 + B2 heat-treated condition exhibits little hardening at low strain amplitude ($\Delta\epsilon_t/2$: $\pm 0.65\%$), but pronounced hardening at high-strain amplitudes ($\Delta\epsilon_t/2 \geq \pm 0.83\%$), starting from the first cycle till the failure, during axial strain cycling ($R = -1$), at room temperature. The C–M plot shows a distinct bilinear behaviour. The LCF resistance of this material is very low and it is due to its poor ductility at RT. However, fatigue life in the region of low total strain amplitude where there is dominance of the elastic strain component appears to be highly promising.

Acknowledgements

The authors are grateful to Dr D Banerjee, Associate Director, and Dr A K Gogia, Deputy Director, Defence Metallurgical Research Laboratory (DMRL), Hyderabad, for supply of the material and keen interest in the work. Financial assistance from the ARDB (Project 642), Government of India, is gratefully acknowledged.

References

- BlackBurn M J and Smith M P 1981 US Patent No. 4 294 615
- Baker I and George E P 1992 *Metals and Materials* June 318
- Coffin L F 1971 *J. Mater. Sci.* **6** 388
- Coffin L F 1972 *Metall. Trans. A* **3A** 1777
- Eswara Prasad N, Malakondaiah G, Raju K N and Rama Rao P 1989 *Proc. Advances in Fracture Research ICF7 Houston, USA* 21103
- Gogia A K, Nandy T K and Banerjee D 1992 *Mater. Sci. Eng.* **A159** 73
- Hayes R W 1989 *Scr. Metall.* **23** 1931[†]
- Kim J Nack, Kim J Y and Cho W S 1993 *Metall. Trans. A* **24A** 1785
- Lipsitt H A, Schechtman D and Schafrik R E 1975 *Metall. Trans. A* **8A** 1991
- Lipsitt H A, Schechtman D and Schafrik R E 1980 *Metall. Trans. A* **11A** 1369
- MacAndrew J B and Kessler H D 1956 *J. Metals* **8** 1348
- Mediratta M G and Lipsitt H A 1980 *J. Mater. Sci.* **15** 2985
- Mediratta S R, Ramaswamy V and Rama Rao P 1986 *Scr. Metall.* **20** 555
- Mishra R S and Banerjee D 1990 *Mater. Sci. Eng.* **A130** 151
- Radhakrishnan V M 1992 *Int. J. Fatigue* **14** 305
- Sanders T H and Starke E A 1976 *Metall. Trans. A* **7A** 1407
- Sastry S M L and Lipsitt H A 1977a *Metall. Trans. A* **8A** 1543
- Sastry S M L and Lipsitt H A 1977b *Acta Metall.* **25** 1279
- Sastry S M L and Lipsitt H A 1980 in *Titanium 80 Science and Technology* (eds) H Kimura and O Izumi (Warrendale, PA: TMS AIME) p. 1231
- Singh P N, Ramachandra C and Singh V 1994 *Low Cycle Fatigue Behaviour of Ti–6.5Al–3.3Mo–1.6Zr–0.3Si at RT*. Presented at 5th Annual General Meeting Feb. 1994 MRSI Hyderabad (to be published)
- Singh V, Sundararaman M, Chen W and Wahi R P 1991 *Metall. Trans. A* **22A** 499
- Soboyejo W O 1992 *Metall. Trans. A* **23A** 1737
- Varma Vijay Kumar, Saxena Vikas Kumar and Srinivas M 1993 *Scr. Metall. Mater.* **29** 673