INFLUENCE OF BENDING FATIGUE ON RESIDUAL MECHANICAL PROPERTIES OF A γ-TiAl INTERMETALLIC ALLOY

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ABSTRACT

The aim of this study is mainly the investigation of the influence of bending fatigue on residual mechanical properties of a near fully lamellar (NFL) γ-TiAl intermetallic material. For this purpose three Ti-47Al-2Cr-2Nb-1B ingots were prepared by Vacuum Induction Melting (VIM) in a ceramic crucible, permanent mould casting and heat treatment without hot-isostatic-pressing (HIP) process. In order to obtain three different NFL microstructures metallic mould was preheated at three different temperatures. Four-point bending strength, bending strain to fracture, fracture toughness, hardness and low-cycle bending fatigue strength $(\sigma_{max})_L$ of near fully lamellar Ti-47Al-2Cr-2Nb-1B with different colony sizes and interlamellar spacings were measured on un-HIP'ed test bars. The S-N curves are fairly flat with a low value of stress life exponent; some previous experiments revealed that the shape of the S-N curve of FL γ-TiAl in the low-cycle fatigue range can be quite different from that in the high-cycle fatigue regime. The effects of low-cycle fatigue, 2×10^5 alternate bending cycles (R=0.1), at high stress intensity levels, i.e. the fatigue limit, on mechanical behaviour of the three different microstructures were evaluated. The data reported and discussed in this paper show that bending fatigue "training" enhances the mechanical behaviour of this material, mainly through the appearance of a plastic deformation during the bending test. According to experimental evidence, the higher the maximum applied stress, normalized by the ultimate flexural strength during fatigue cycling, the larger the improvement of mechanical properties of NFL microstructures The effects of the microstructures on bending deformation, fracture behaviour and fatigue resistance were discussed as well.

1 INTRODUCTION

Fully lamellar gamma titanium aluminides are under extensive research because of their low density, adequate high temperature properties and oxidation resistance. In order to promote their structural applications, the mechanical fatigue behaviour has to be studied in details.

Fully Lamellar (FL) and Duplex (DP) are the microstructural morphologies responsible for the most interesting mechanical behaviour of gamma titanium aluminides: the fully lamellar microstructure shows the highest fracture toughness and fatigue crack growth resistance, whereas the duplex one displays larger tensile properties and better resistance to fatigue crack nucleation. Kim [1].

The mechanical properties of FL γ-TiAl strongly depend on composition and microstructure. It has been shown that the grain size or colony size influences fracture toughness and tensile ductility of FL materials in opposite manners, resulting in an inverse relationship between fracture toughness and ductility. Kim [1,2] Chan[3,4]. Fatigue life and fatigue crack propagation resistance increase with the lamellar colony size, whereas finer lamellar microstructure is more resistant to crack nucleation. Dowling [5].

Specific interest was given to the microstructural features influencing mechanical strength and ductility, fracture toughness and fatigue resistance of NFL γ-TiAl

A serious barrier to the industrial diffusion of this class of materials comes from its high cost, also due to HIP treatment usually performed to reduce porosity. A proper understanding of the behaviour of this material under mechanical fatigue in the presence of microdefects may well promote its structural employment in un-HIP'ed conditions, resulting in decreased production costs. Following this philosophy the defect tolerance of titanium aluminides is under investigation. Lerch [6].

In this work HIP treatment was not utilized and the integrity of all the samples was checked by radiographic testing.

2 EXPERIMENTAL PROCEDURES

Three ingots, 20x60x90 mm, of nominal composition (at. %) Ti-47Al-2Cr-2Nb-1B with three different NFL microstructures were cast in a metallic mould after induction melting of the pure elements (99.9%) in a zirconia crucible coated with plasma-sprayed yttria.

The aim was to obtain three different NFL microstructures, therefore the copper mould was preheated to 300°C, 500°C and 600°C to cast ingots n°1, n°2 and n°3 respectively.

The three ingots were heat treated below the eutectoidic temperature for obtaining stress relief and chemical homogenization.

Small bars (3x8x50 mm) were cut from each ingot by spark erosion for mechanical testing. Samples were polished using SiC papers.

As reported in the introduction the ingots were not submitted to HIP treatment. The samples, taken from the sound part of the ingot, i.e. only rejecting the area close to the shrinkage cone, were selected by radiographic examinations performed according to ASTM 1742-95 standard.

Four-point bending strength (modulus of rupture, MOR) was measured utilizing a servohydraulic machine equipped with a load cell of 5KN using a crosshead rate of 5 mm/min.

Table 1:Chemical composition and microstructural parameters of Ti-47Al-2Cr-2Nb-1B ingots after heat treatment.

Ingot	Microstructure	Colony size [μm]	Interlamellar spacing [nm]	Size of equiassic γ grains [μm]
n°1	NFL	200-500	250	5-10
n°2	NFL	500-1000	400	5-25
n°3	NFL	700-1200	500	5-30

Chemical comp. (wt%)	Ti	Al	Cr	Nb	B	O	H	N	Zr+Y
n°1	Balance	32.71	2.83	4.57	0.26	0.097	0.001	0.002	<0.1
n°2	Balance	32.75	2.78	4.59	0.27	0.096	0.001	0.002	<0.1
n°3	Balance	32.81	2.81	4.49	0.26	0.099	0.001	0.002	<0.1

Fracture toughness was evaluated by three-point bending tests of Single Edge Precracked Beam (SEPB) specimens (3×4×25 mm^3) according to ASTM E 399 standards. Hardness (HV) was measured by a Vickers indentor using a 10 kg maximum load. All the reported values represent the average of at least five tests.

Fatigue experiments were carried out in four-point cyclic bending mode with spans of 40 and 20 mm, applying a 10 Hz frequency sinusoidal wave and a stress ratio $R=\sigma_{max}/\sigma_{min}= 0.1$. Fatigue tests were performed using a servohydraulic machine with a load cell of 5 KN and specimens of 3×8×50 mm; low cycle bending fatigue strength $(\sigma_{max})_L$ was statistically determined at 2×10^5 cycles with 80% reliability, applying the "stair case" method.

In order to examine the effects of low cycle fatigue at high stress levels on mechanical behaviour of NFL γ-TiAl, the residual flexural strength, fracture toughness and hardness of the three different microstructures were measured on those specimens which had survived fatigue testing.

3 RESULTS

The NFL as-cast microstructures of the three ingots are different in lamellar colony or Grain Size (GS) and interlamellar spacing, as well as in the amount and size of equiaxed γ phase as reported in Table 1.

Mechanical properties of the heat-treated material are reported in Table 2. Four-point bending strength was first measured. None of the tested specimens shows evidence of diffuse plastic deformation.

Table 2:Mechanical properties after heat treatment

	Ingot n°1		**Ingot n°2**		**Ingot n°3**	
	Before fatigue	**After fatigue***	**Before fatigue**	**After fatigue***	**Before fatigue**	**After fatigue***
HV (100 N)- *as cast*	400	—	380	—	370	—
HV (100 N)- *after heat treatment*	390	390	335	335	320	320
MOR [MPa]	625	700	480	575	420	560
Crosshead displ. to failure [mm]	0.37	0.52	0.34	0.48	0.3	0.46
K_{Ic} [MPa m ½]	32±0,5	33±0,5	19±0,5	21±0,5	13±0,5	16±0,5
$(\sigma_{max})_L$	440		425		375	
$(\sigma_{max})_L$ /MOR	0.70		0.88		0.89	
MOR*/MOR	1.12		1.20		1.33	

* mechanical properties were measured after 2×10^5 four-point bending cycles at $(\sigma_{max})_L$ with R=0.1 and f=10Hz; $(\sigma_{max})_L$ is the four-point alternate bending fatigue strength after 2×10^5 cycles with R=0.1 and f=10Hz

Fracture toughness tests show an average K_{Ic} value of 32±0,5 MPa $m^{1/2}$ for ingot n°1, whereas the lowest K_{Ic} value was displayed by the specimens with microstructure n°3 (K_{Ic} ≅13±0,5 MPa $m^{1/2}$). Transverse optical micrographs of cracks and SEM fracture surfaces are shown in Fig. 1. The SEM micrograph reported in Fig. 1(a) shows evidence of interlamellar fracture, whereas Fig. 1(b) testifies translamellar fracture and many intergranular secondary cracks are also evident. Fig. 1(b) refers to the ingot with microstructure n°3 showing the presence, in hard orientation in relation to the crack plane, of debonded lamellae with small cleavage facets. Fracture surfaces of the fine microstructure of ingot n°1, Fig. 1(c), appear rougher due to crack deflection activation.

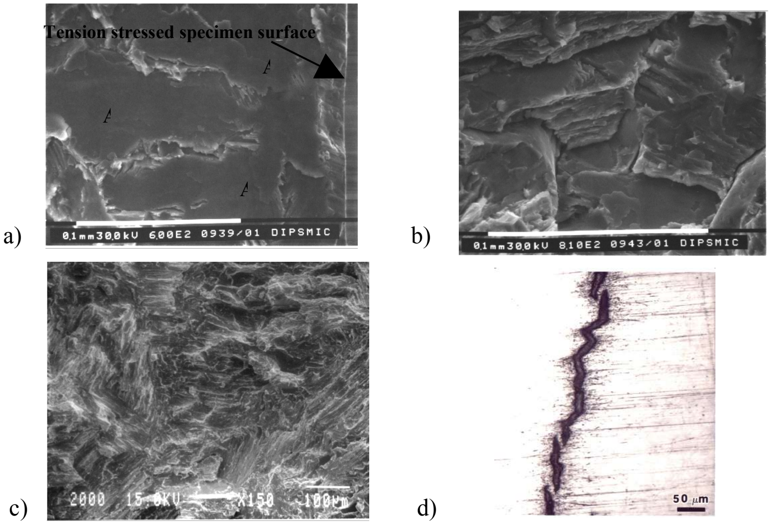

Fig. 1: SEM micrographs (a) and(b) refer to RT four point bending fracture surfaces of microstructures investigated showing: (a) interlamellar clivage fracture and (b) mostly translamellar fracture with secondary cracks between colony boundaries in specimens from ingot n°3; (c) reports translamellar fracture and rough fracture surface of specimens with microstructure n°1; (d) shows bridging by crack-wake ligaments.

As shown in Table 2, ingot n°1 displays a low-cycle fatigue strength $(\sigma_{max})_L$ of about 440 MPa; hence, its specimens which survived fatigue testing were cycled in bending mode at a $(\sigma_{max})_L$/MOR ratio equal to 0.70. Ingots n°2 and n°3 display lower fatigue strength and their $(\sigma_{max})_L$/MOR ratio is almost equal to 0.90 as confirmed by Chan [3] and Kim [2]. Thus, even though fatigue strength of ingot n°1 is the highest, the specimens with microstructures n°2 and n°3 reach higher (σ_{max})L/MOR ratios than the finer microstructure n°1. Fatigue testing shows that alternate bending lifetime is greatly reduced at stress levels just above $(\sigma_{max})_L$; in fact, the S-N curves, Fig. 2, are fairly flat with a low value of stress life exponent. During previous experiments Chan [4] revealed that the shape of the S-N curve of FL γ-TiAl in the low-cycle fatigue range (LCF, $N<10^6$) can be quite different from that in the high-cycle fatigue (HCF) regime ($>10^6$). According to fatigue resistance data reported by Chan [7], the elevated low-cycle fatigue strength shown by the FL ingots with coarse colony size can be attributed to their larger resistance to large crack propagation: in FL microstructures small cracks can easily nucleate between lamellae at applied stresses below the large-crack growth threshold (ΔK_{th}), especially in favourably oriented lamellar colonies, by decohesion of interlamellar slip bands; the interlamellar nucleated cracks can easily propagate up to the colony boundary where they remain arrested until stresses are increased to a level that allows further fatigue crack growth into the neighbouring colonies. Resistance to fatigue large-crack propagation should therefore increase with grain size along with fatigue strength, as the S-N curve shown in Fig. 2 confirms: the flattest curves correspond to the two coarser grained microstructures (n°2 and n°3).

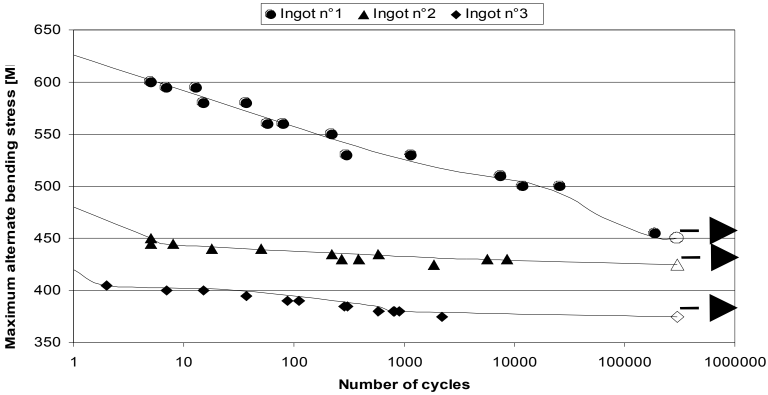

Fig. 2. S-N curves for microstructures of the ingots n.1, n.2 and n.3; specimens were tested at a rate of 10Hz with R=0.1.

Fractographic observations on the specimens obtained from ingots n°1, n°2 and n°3 which failed during fatigue tests indicate that fracture starts at weak interfaces by decohesion of lamellae parallel to the fracture surface and then propagates tortuously between the grains and translamellarly. Many secondary cracks depart perpendicularly from fracture surfaces and are arrested in their propagation by the unfavourable lamellae orientations. Debonded lamellae on secondary crack surfaces proves the existence of small zones characterized by some plastic deformation, typical of fatigue cycling. Secondary cracks at grain boundaries are also observed.

In order to investigate the effects of low-cycle fatigue at high stress levels on mechanical behaviour of NFL γ-TiAl, the residual flexural strength, fracture toughness and hardness were measured on specimens which survived the fatigue testing. The mechanical properties after 2×10^5 alternate bending cycles at σ_{max} equal to the $(\sigma_{max})_L$ are reported in Table 2. An increase in flexural strength and elongation to failure can be observed for all samples. Conversely, no increase in hardness was observed. This behaviour testifies the singular influence of fatigue "training" at $(\sigma_{max})_L$ on mechanical properties of the cast ingots. The load vs. displacement curve, which is representative of bending behaviour of all the cast ingots after 2×10^5 alternate bending cycles at the correspondent $(\sigma_{max})_L$/MOR ratio, points out that specimens show plastic strain to fracture, non existent before fatigue "training".

The ductility after fatigue "training" is confirmed by the increased fracture toughness.

The analysis of fracture surfaces confirms an increased ductility of lamellar colonies and a larger amount of translamellar fracture. The increased fracture toughness values and the macroscopic ductility of the fatigued specimens can be attributed to the increase of intrinsic ductility of lamellar colonies. The larger ductility ahead of the crack tip may arise from lamellae twin toughening due to alternate bending cycles and/or to a possible fatigue-activated toughening mechanism at lamellae interfaces. The mechanical behaviour of the material can be interpreted as a multilayered-microcomposite. As a consequence not only toughening mechanisms, like microcracking, but also twin-sheet toughening are activated by low-cycle fatigue.

Said behaviour may supply an explanation for the experimental data regarding the increased values of Kc and MOR after fatigue.

4 CONCLUSION

The results of the mechanical tests performed testify that even in the presence of size-selected microdefects it is possible to define a fatigue limit which allows the structural employment of un-HIP'ed γ-TiAl components.

The present study confirms the strong influence of microstructure on fatigue behaviour of NFL γ -TiAl base intermetallics, in this case un-HIP'ed Ti-47Al-2Cr-2Nb-1B.

Some general features can be summarized as follows:

1. The $(\sigma_{max})_L$/MOR ratio increases with the grain size.
2. MOR and fracture toughness increase after fatigue "training".
3. Appearence of a macroscopic plastic behaviour in σ-ε plot after fatigue.
4. The larger the $(\sigma_{max})_L$/MOR ratio applied to each microstructure during fatigue "training", the larger the improvement of the mentioned mechanical properties.
5. Unchanged values of hardness before and after fatigue

On the base of the remarks reported thus far, in order to explain the enhancement of after-fatigue mechanical features, it is feasible to put forward an interpretation in terms of activation of toughening mechanisms resulting from low-cycle fatigue.

5 REFERENCES

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