EFFECT OF STRAIN LOCALIZATION ON FRACTURE BEHAVIOUR OF FATIGUED GAMMA TiAl-2Nb ALLOY

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Abstract
Cyclic strain localization, in nearly lamellar cast gamma-TiAl based alloy Ti–48Al–2Nb–2Cr–0.82B (all in at. %) was experimentally studied during low cycle fatigue using observations of surface relief and fracture surfaces. Parameter of hysteresis loop $V_H$, the cyclic stress-strain curve (CSSC) and fatigue life curves at room temperature were assessed. The surface relief and the fracture surfaces of fractured specimens were observed using two high resolution scanning electron microscopes (SEM) TESCAN MIRA 3 and LYRA 3 equipped with FEG sources. The dislocation structure of persistent slip bands and the surface profile of persistent slip markings from which fatigue cracks started was studied by transmission electron microscope (TEM) on FIB prepared surface TEM lamella. The variations of the $V_H$ parameter indicate the cyclic slip localization. The localization has a pronounced effect on the surface relief formation. Persistent slip markings arise preferably on the surface along interlamellar gamma/gamma interfaces. These locations were nuclei for fatigue cracks which propagated in the interior of grains and led to the formation of smooth flat areas on the fracture surface corresponding to the persistent slip bands. The rest of the fracture surfaces reveal rather brittle behaviour originated during static fracture due the low fracture toughness of gamma-TiAl based alloys.

Keywords: low cycle fatigue, lamellar TiAl alloy, fracture surfaces, persistent slip markings, loop shape parameter

INTRODUCTION
TiAl alloys represent a serious candidate material for applications demanding low density, good corrosion resistance and high strength at elevated temperatures. Recently, applications using TiAl alloys appeared e.g. as parts of vehicle's engines or in aeronautics [1]. The material in service undergoes cyclic loading at high temperatures. However, the occasional large mechanical or thermal transients could also give rise to a significant amount of plastic deformation. Under these conditions, commonly referred to as low cycle fatigue (LCF), the total design lifetime may involve only few hundreds or few thousands of these large strain cycles. Turbine engine blades or discs are prime examples of components subjected to this type of strain-controlled fatigue. Local areas of these components can undergo high stresses and strains due to external load transfer, abrupt changes in geometry, temperature gradients, and material imperfections [2]. The relation between dislocation microstructures and cyclic stress-strain response at room and 750°C temperatures in
polycrystalline of TiAl alloys have been studied by Gloanec et al. [3]. Satoh et al. [4] evaluated the changes of the stress-strain hysteresis loops using energy parameter $\beta_E$ or using variations of loop shape parameters $V_H$. Surface relief observations and fatigue crack initiation after stress-controlled cyclic loading were studied by Jha et al. [5]. A detailed complex study of the cyclic strain localization in gamma based TiAl alloys with effect on fracture behaviour has not been performed. Some LCF data on TiAl alloy with 2 at. % of Nb have been already reported [1,6,7]. Fatigue parameters at room temperature and at 750 °C were evaluated [1,6], fracture surfaces were described [6] and the cyclic plasticity was analysed using the generalized statistical theory of the hysteresis loop [7].

In this paper, complementary new results on nearly lamellar cast $\gamma$-TiAl-2Nb based alloys on the relation between internal lamellar structure, dislocation structure of persistent slip bands and resulting fracture surface. Fatigue crack initiation starting from cyclic strain localization areas during room temperature low-cycle fatigue experiments in air are reported and analysed.

**EXPERIMENTAL DETAILS**

**Experimental material**

The experimental material Ti–48Al–2Nb–2Cr–0.82B at.% (TiAl 2Nb alloy in the following) was prepared by casting in the GfE Metalle und Materialien GmbH company in Nürnberg in the form of a cylindrical ingot of 220 mm in length and 90 mm in diameter. The ingot was subjected to hot isostatic pressing (HIP) at 1280 °C and 140 MPa for 4 hours. The alloy shows nearly lamellar microstructure with $\gamma/\alpha_2$ laths (with average thickness of 1.95 μm), variable grains size (80 μm – 1 mm) and some smaller areas without the lamellar substructure of single $\gamma$ phase on the grain boundaries (see Fig. 1).

![Fig.1. Structure of the cast TiAl 2Nb alloy: (a) etched to reveal the heterogeneous grain size in a section perpendicular to the ingot axis (optical micrographs); (b) electron micrographs of lamellar microstructures.](image)

The room temperature tensile properties are shown in Table 1. Some other material parameters are can be found elsewhere [1,11].
Tab.1. Tensile properties at room temperatures of studied cast TiAl 2Nb alloy [11].

<table>
<thead>
<tr>
<th>E [GPa]</th>
<th>0.1 % Yield stress [MPa]</th>
<th>Fracture stress [MPa]</th>
<th>Fracture $\varepsilon_\text{p}$ [%]</th>
<th>Fracture $\varepsilon$ [%]</th>
</tr>
</thead>
<tbody>
<tr>
<td>180 ± 2</td>
<td>398</td>
<td>415</td>
<td>0.13</td>
<td>0.364</td>
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Low cycle fatigue testing

Cylindrical specimens used for fatigue tests were carefully mechanically and then electrolytically polished in the gauge length. Fatigue tests were performed under strain control using MTS servohydraulic machine in symmetric tension-compression cycle ($R_e = -1$). The total strain amplitude ($\varepsilon_a$) and strain rate of $2 \times 10^{-3}$ s$^{-1}$ were kept constant in all tests. The tests were performed at room temperature (RT). During cyclic loading, hysteresis loops of selected cycles were recorded for further analysis. The total strain amplitude $\varepsilon_a$, stress amplitude $\sigma_a$, and mean stress $\sigma_m$ were recorded. After the test termination, a special program was used to evaluate plastic strain amplitude ($\varepsilon_{ap}$) from the half-width of the hysteresis loop [11].

The loop shape parameter $V_H$ [8] was evaluated from the recorded hysteresis loops using relation

$$V_H = \frac{W}{4 \varepsilon_{ap} \sigma_a},$$

where $W$ is the area of the hysteresis loop.

Microscopy observation

The surface relief and fracture surface observations were studied on the electropolished gauge length of specimens fatigued to failure using high resolution scanning electron microscope with field emission MIRA 3 FEG-SEM from TESCAN company. The dislocation structure of persistent slip bands (PSBs) of specimen cycled with $\varepsilon_a = 0.285$ % to fracture ($N_f = 1,794$) was observed using transmission electron microscope Philips CM-12 on lamella prepared by perpendicular cut to the specimen surface and to the direction of the persistent slip marking (PSM) by focused ion beam (FIB) system LYRA3 made by company TESCAN. The final FIB polishing step was performed at 30kV with probe current 56 pA.

RESULTS AND DISCUSSION

Cyclic plastic response and fatigue life curves

The shape of the hysteresis loop significantly changes during cycling for all strain amplitudes [7,12]. Cyclic hardening/softening curves show the increase of stress amplitude with the number of cycles (Fig. 2) which results in the decrease of the width of the hysteresis loop [12]. The character of all curves is similar and shows significant cyclic hardening up to fracture which is in agreement with earlier observations [1,3]. The intensive short initial cyclic hardening is followed by slow secondary cyclic hardening with a tendency to reach stabilized cyclic stress-strain response.

The loop shape parameter has different evolution with number of cycles (Fig. 2). The initial drop and local minimum are followed by an increase reaching a local maximum and then the decrease of $V_H$ until failure. Characteristic $V_H$ dependence vs. number of cycles $N$ corresponds to that found in copper polycrystals and single crystals [8-10] and recently also in TiAl [4]. The minimum of $V_H$ has been identified with the start of the appearance of PSBs and PSMs. Our observations show that a similar behaviour is also found in lamellar cast $\gamma$-TiAl based alloy. Surface observations (see below) give support to the interpretation
of $V_H$ changes. The minimum of the $V_H$ parameter defines the start of the strain localization which is characterized by the nucleation of the PSBs. During intensive formation of new PSBs the $V_H$ parameter increases and reaches a maximum at which the microstructure of PSBs is formed and the stress amplitude and plastic strain amplitude have tendency to saturation. Last stage during which $V_H$ parameter decreases is connected with the growth of short fatigue cracks [8-10].

![Graph](image)

Fig. 2. Characteristic quantities (stress amplitude $\sigma_a$ and the loop shape parameter $V_H$) vs. number of cycles $N$ for selected strain amplitudes $\varepsilon_a$ of TiAl 2Nb alloy at room temperature.

![Graph](image)

Fig. 3. The Coffin-Manson fatigue life curves of TiAl 2Nb alloy at two temperatures. And different values of cumulative plastic strains $\varepsilon_c$. 

Coffin-Manson plot in Fig.3 shows plastic strain amplitude $\varepsilon_{ap}$ vs. number of cycles to fracture $N_F$ in bilogarithmic representation. The RT fatigue is characterized by significant scatter of experimental data, which reflects low fatigue ductility.

**Surface relief and fracture surface observations**

The surface on gauge length was observed in SEM after low cycle fatigue tests (see Fig.4). The surface relief is formed by numerous persistent slip markings (PSMs) consisting of very thin remains extrusions (20 nm in thickness) along lamellae of $\gamma$ and/or $\alpha_2$ phases (see Fig.4b). Consequently, many fine and short fatigue cracks initiate in these regions (see Fig.4b,c).

Fig.4. The surface relief of cycled of TiAl 2Nb alloy (RT, $\varepsilon_a = 0.285 \%$, $N_f = 1794$): (a) overview of a broken specimen, (b) detail of PSMs on a grain and (c) fatigue cracks initiated along PSMs.

This is in agreement with earlier observations [6,12] where surface cracks initiated at lamellae interfaces after cyclic loading at RT and at 600°C [5]. Dislocation structure was observed in surface TEM-lamella (see Fig.5) prepared perpendicularly to PSMs by FIB of a specimen cyclically strained to fracture. The internal dislocation structure consisted from twins, ordinary dislocations and superdislocations mainly in the gamma phase. The intrusion is closely located at the surface and follows the twins. Its depth is about 200 nm.
Fig.5. The TEM image of surface relief and dislocation structure in a lamella prepared by FIB perpendicularly to the PSMs shown in Fig.3. The white arrow indicates the position of an intrusion.

The fracture surfaces reveal rather brittle behaviour of TiAl 2Nb (see Fig.6). All the fatigue cracks initiated at surface (Fig.6b) creating smooth flat areas with the size around 250–900 μm corresponding to the size of PSBs and the grain size. The rest of the fracture surface originated during final static fracture as a result of the low fracture toughness of γ-TiAl based alloys [5].

Fig.6. The fatigue fracture surface of 2Nb alloy (RT, ε_a = 0.31 %, N_f = 10 376), (a) the overview and (b) detail of fatigue crack initiations region – smooth flat area.

CONCLUSIONS
The research in cyclic strain localization and fracture surface observations after low cycle fatigue tests at constant strain amplitude at RT in gamma TiAl 2Nb alloy lead to the following conclusions:
• Material exhibits continuous cyclic hardening which starts with rapid initial stage and continues with slow secondary cyclic hardening and is followed by stabilized cyclic stress-strain response. This is accompanied by intensive microtwins formation and their interaction with dislocations.
• Plot of the V_{II} parameter reveals pronounced cyclic strain localization. It begins already early in cyclic loading.
• Localization of the cyclic strain to the PSBs results in the formation of PSMs along $\gamma/\alpha_2$ and $\gamma/\gamma$ interfaces and in fatigue crack nucleation and growth.
• The slope of Manson-Coffin fatigue life curves indicates the low fatigue ductility of the alloy.
• The fracture surface reveals rather brittle behaviour corresponding to final fracture. Presence of smooth flat areas on the fatigued produced fracture surface indicates the stage I cracks initiated along PSBs parallel into interlamellar interfaces.

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