MICROSTRUCTURAL FEATURES OF SHORT CRACK GROWTH IN A BETA TITANIUM ALLOY

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ABSTRACT

Initiation and early growth of short cracks during fatigue loading are strongly affected by the local microstructural arrangement, e.g. grain geometry and crystallographic orientation. In the case of the beta-titanium alloy Ti-6.8Mo-4.5Fe-1.5Al (TIMETAL®LCB) used in this study, even under macroscopically elastic loading slip bands and small cracks are generated at highly misoriented grains due to the elastic anisotropy. While high-misorientation conditions promote crack initiation, the propagation of short cracks is hindered by crack path deflection at high-angle grain boundaries or possibly crack closure effects. In the present paper this could be shown by fatigue tests which were carried out in combination with the measurement of local displacement by means of a laser interferometric strain/displacement gauge (ISDG) and the determination of individual crystallographic orientations by means of electron back-scattered diffraction (EBSD) in the scanning electron microscope (SEM). The experimental results were proved by FE modelling of exemplary details of the deformed microstructure.

KEYWORDS

short cracks, crack closure, elastic anisotropy, EBSD, ISDG, beta-titanium alloys

INTRODUCTION

While the growth of long fatigue cracks can be described in a uniform manner by the well-known sigmoidal function of the crack growth rate (log da/dN) versus the stress-intensity factor (log ΔK), the behaviour of short cracks is characterised by the lack of uniformity, which involves the risk of non-conservative design [1]. By observing individual microstructural short cracks it can be found that (1) cracks grow at ΔK values much lower than the threshold value for long cracks ΔK_{th}, (2) crack propagation rates decrease or even stop, or (3) cracks propagate at higher rates than predicted for the long crack curve [1]. This behaviour is due to the fact that the growth rate of microstructurally short cracks is determined by both, the shear stress on the operating slip planes as a consequence of local Schmid factor variations and dislocation pile-up in front of barriers like grain boundaries (GBs) or precipitates [2-4]. Crack blocking at GBs is often followed by a rapid increase of the crack propagation rate leading to oscillating crack growth kinetics, which Navarro and de los Rios attributed to the activation of dislocation sources in a grain i+1 by the dislocation pile up in the neighbouring grain i at the barrier [5]. The blocking effect of GBs as well as the geometrical arrangement of the slip systems are closely related to the crystallographic orientation and the geometry of individual grains.

High-misorientation conditions and low multiplicity of slip promotes plastic incompatibility of neighboured grains and intergranular crack propagation [6,7], while low angle GBs can easily be crossed by slip and therefore by propagating transgranular cracks [8]. Once a short crack begins to grow it suffers further microstructural interactions by crack coalescence and crack closure which speed it up or slow it down [1,9]. Although plasticity-induced crack closure, which plays a considerable role for the growth mechanisms of long cracks, should be negligible for short cracks due to the small plastic wake behind the crack tip [10], roughness-induced crack closure effects and plastic displacement in front of the crack tip might be important factors influencing short crack propagation [11].
Stimulated by the knowledge that life of dynamically loaded components is often determined by the propagation of short cracks, the present study deals with the quantification of microstructural features, which are responsible for stress localisation, crack initiation and early crack growth kinetics, and is focussed on the objective to develop a model for mechanism-oriented life-time prediction.

EXPERIMENTAL DETAILS

The studies on the short crack propagation mechanisms were carried out on the metastable beta-titanium alloy TIMETAL®LCB (low cost beta, Ti-6.8Mo-4.5Fe-1.5Al), which was developed as a relatively cheap and light-weight alternative to high-strength steel in the automotive industry (e.g. for suspension springs). The results presented in this paper are restricted to fatigue tests in the solution heat-treated condition, exclusively consisting of b.c.c. beta grains with an average grain size of 102µm. Shallow-notched (in the gauge length) cylindrical specimens, which were machined from hot-rolled rods and vacuum heat treated (830°C, 5h, water-quenched) were electropolished and fatigued in a SCHENCK S31 servohydraulic testing system using symmetrical pull-push loading conditions at room temperature under stress control. During fatigue testing, initiation and early growth of fatigue cracks was observed by an optical microscope which is attached to the testing system and connected to a digital image analysis system to generate individual crack propagation vs. crack length plots (*da/dN* vs. *a*, see ref. [12]).
In order to estimate the contribution of crack closure effects to the propagation behaviour of short cracks the testing system is equipped with an self-made laser interferometric strain displacement gauge (ISDG), according to Sharpe's concept [13], which allows high-resolved determination (theoretical resolution: 1.85nm) of local strain. Data is produced by moveable photodiodes which follow the displacement of an interference pattern generated by a laser beam, which is reflected at two Vickers microhardness indentations in the specimen surface (Fig. 1). The shift of the microhardness indentations, placed approx. 50µm above and below a growing short crack, yields the crack opening displacement COD, which contains information e.g. about local crack closure effects.

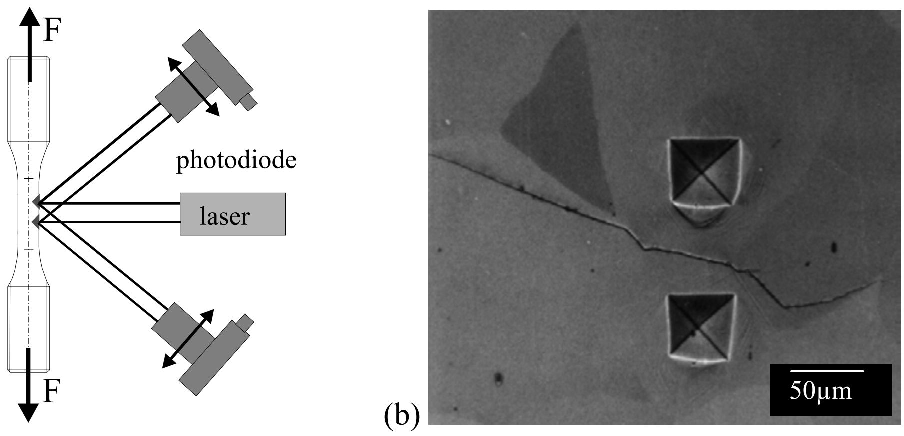
(a) (b)


Figure 1: Measurement of local displacement by an ISDG system: schematic representation (a) and Vickers microhardness indentations above and below a growing short crack ($\Delta\sigma/2$=600MPa ⇳)(b)

To carry out microstructural examinations, the specimens were removed from the fatigue testing machine after certain numbers of cycles. Due to the smooth electropolished surface, the grains are observable in electron channeling contrast (ECC) in the SEM (PHILIPS XL30 LaB_6). The individual crystallographic orientations of the grains participating in cracking processes on the surface and the corresponding misorientation angles were evaluated by an EBSD system (TSL). The physical principle and the evaluation methods are described in detail in [14].

RESULTS AND DISCUSSION

Crack Initiation

Fatigue tests, carried out at a stress amplitude of $\Delta\sigma/2$=600MPa (approx. 50% of the yield stress σ_Y of LCB in the solution heat treated condition) and a stress ratio of R=-1, were interrupted several times during the fatigue life of approx. N_f=5000 cycles to characterise the initiation sites of microstructural short cracks. Fig. 2 represents that predominantly cracks originate at GBs either intergranularly (Fig. 2a) or transgranularly along slip bands (Fig. 2b). EBSD measurements of more than 600 grains reveal that during fatigue life (1) approx. 80% of all the cracks initiate transgranularly and (2) intergranular cracking is restricted to high-angle GBs, which are defined by misorientation angles Θ>15° [8]. Analogously to the results Zhang and Wang obtained for Cu bicrystals in [6], high-angle GBs seem to prevent slip across the boundary, hence they give rise to plastic incompatibility and therefore crack initiation.

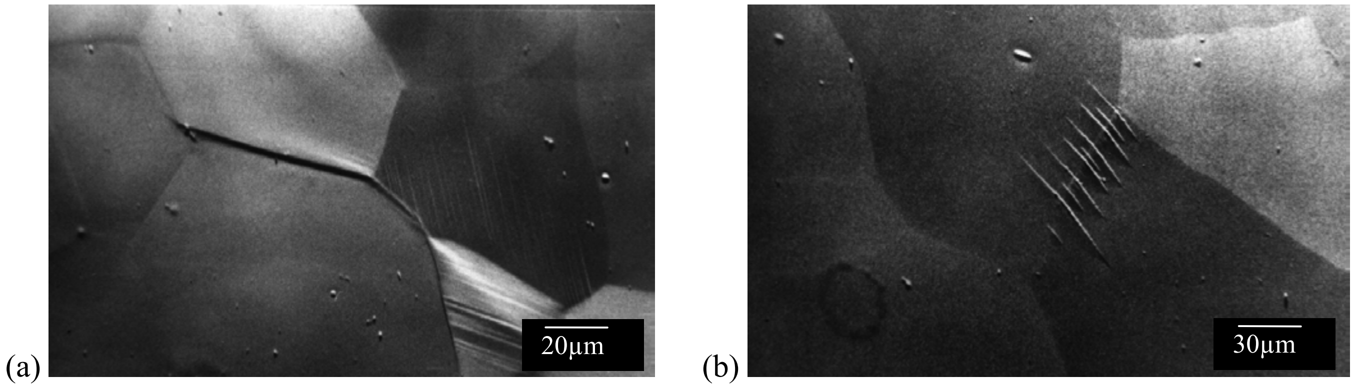


Figure 2: Intergranular (a) and transgranular crack initiation (b) at high-angle GBs ($\Delta\sigma/2$=600MPa ⇔)

Since nearly all the cracks observed initiated at GBs without any interactions with precipitates, pores etc., it can fairly be assumed that elastic anisotropy creates high stress levels in the vicinity of highly-misoriented grains. A mechanical confirmation requires the three-dimensional stiffness matrix of the test material LCB. Due to the lacking of such data in literature, a method has been developed to calculate the anisotropy on the base of local displacement measurements (by means of ISDG) and crystallographic orientation determination (applying EBSD) [15]. By FE modelling of several real cracked microstructures as it is shown exemplary in Fig. 3 using the determined anisotropy value of A≈0.7 it was demonstrated, that local stress reaches maximal values at exactly those GBs which cracked during fatigue loading.

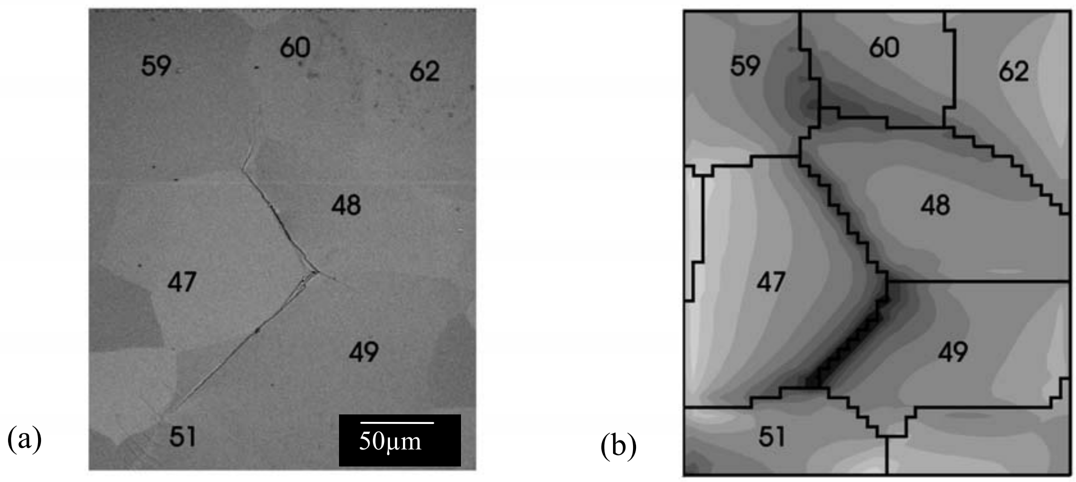


Figure 3: Intergranular crack initiation: SEM micrograph (a) and corresponding calculated stress distribution (b) (maximum mean stresses: high stress values = dark) ($\Delta\sigma/2$=600MPa ⇔)

Crack Propagation

Once a short crack has been initiated its propagation rate and direction depends on (1) the direction of the operating slip planes with regard to the loading axis and the GBs, (2) the crystallographic misorientation of neighbouring grains, (3) crack closure effects, and (4) mutual interactions and coalescence of short cracks. Fig. 4a shows several microstructurally short cracks on the surface of the beta-titanium alloy LCB. The one marked with a solid arrow follows slip planes inclined 45° with regard to the loading axis, i.e. in direction of the maximum shear stresses. It could be observed that high deflection of the crack path at a barrier is in most cases connected with a temporary arrest, while dislocation pile-up activate crack continuation in the neighbouring grain, as proposed in the model of Navarro and de los Rios [5]. On the contrary, at high constrained GBs (e.g. in Fig. 4a, dashed arrow) many slip bands are activated leading to high back stresses towards the GB which fails by slip-step cracking instead of growing slip band cracks. The observation of several growing cracks during fatigue testing by an optical microscope supplemented by SEM/EBSD examinations reveals, that in particular low-angle GBs or low intersection angles between slip planes of neighboured grains or intergranular cracks as well as slip band cracks parallel to the maximum shear stress are prone to short crack propagation.

Furthermore it might be mentioned that fatigue damage cannot be attributed to a single short crack. Fig. 4b demonstrates the intensified plasticity in the vicinity of two approaching crack tips. By applying the replica technique the following coalescence events were identified as the substantial mechanisms leading to the transition from short to long crack behavior. It is worth mentioning that modelling of crack coalescence requires a statistical approach, which has been done by a promising microstructure simulation in combination with the short crack growth model of Navarro and de los Rios in [16].

(a) (b)

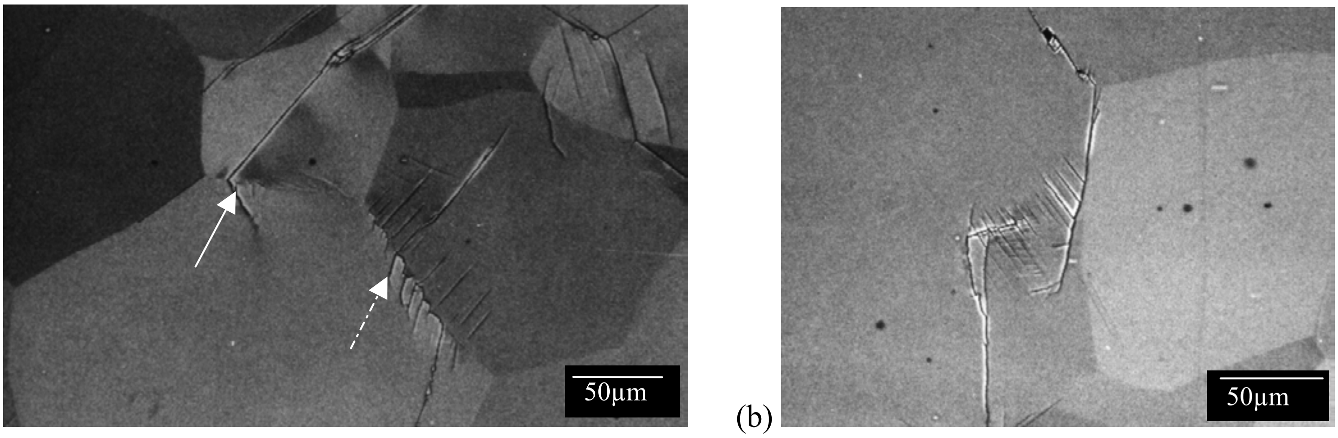


Figure 4: Crack deflection and plastic incompatibility at a high-angle GB (a) and plastic distortion in the vicinity of two coalescent cracks (b) ($\Delta\sigma/2$=600MPa ⇔)

Since direct observation of short cracks is restricted to the two-dimensional surface, the crack shape is often assumed to be semicircular or semi-elliptical. For long cracks, which are almost independent of microstructural features, this simplification might be true, the depth of short cracks however tends to be dictated by the crystallographic orientation and the GB arrangement in a similar way than it was shown for surface crack propagation. The local aspect ratio a/c (crack depth / half surface crack length) was determined by preparing and evaluating metallographically various cross sections along selected surface cracks. It was found, that early short crack growth into the material is slower than at the surface. This can be derived from an average aspect ratio of approx. a/c=0.5 (for more details see ref. [12]).

The studies of short cracks on the surface of the beta-titanium alloy LCB gave rise to the assumption, that shear stresses in the slip planes are the predominant driving force for crack propagation and GBs act as crack growth retarding barriers (or in some cases as intergranular crack initiation sites). For long cracks it is generally accepted that crack propagation depends strongly on crack closure effects, originally reported by Elber in [17]. He established, that the plastic zone formed during tensile loading at the tip of a growing crack leads to a premature contact of the crack surface during unloading. Therefore the part of the cycle in which the crack is open is reduced by plasticity-induced crack closure, i.e. the crack driving stress intensity factor range ΔK is reduced to an effective stress intensity factor ΔK_{eff} by the closure stress intensity factor ΔK_{cl}:

$\Delta K_{eff} = \Delta K - \Delta K_{cl}$. Furthermore, a jagged crack surface in combination with a displacement of the opposite crack areas result in a premature crack closure, which is known as roughness-induced crack closure.
Although there are numerous studies on crack closure effects during the propagation of long fatigue cracks, there is a lack of data concerning the impact of crack closure on the propagation of microstructurally short cracks. Since the length of the plastic wake behind the tip of a short crack, which is responsible for plasticity-induced crack closure, as well as the extent of roughness are assumed to be quite small, a transient behavior from short to long crack closure might be expected, according to [18]. Fig. 5 shows exemplary results and the evaluation of the COD measurements for two stress amplitudes at a stress ratio of R=-1. The closure stress σ_{cl} (defined by the intersection point of the tangents on the parts of the hysteresis loop where the crack is fully closed and fully open, resp.) for the lower stress amplitude is appreciable for surface crack lengths of more than approx. 50µm and lies in the tensile part of the cyclic loading (Fig. 5b). Since no plastic deformation could be observed at the crack tip, the positive closure stress is attributed to roughness-induced crack closure. For higher stresses, the crack becomes fully closed in the compressive part of the loading cycle. Following the remarks given in Liaw's review article [19] about crack closure at near-threshold fatigue crack growth levels, higher plastic strain at the crack tip of short cracks as compared to long cracks and the simultaneous absence of a plastic wake can explain the negative crack closure stress at higher stress amplitude. This assumption is supported by the plastic opening of the hysteresis loop for the higher stress amplitude in Fig. 5a (both curves represent surface crack lengths of approx. 300µm). Crack closure effects, which extend the part of the loading cycle when the crack is open, might additionally rise to higher propagation rates of short cracks compared to long cracks (based on the usual assumption, that K_{min} is set to zero if σ_{min} is negative). Concerning such comparisons it should be mentioned, that the application of the ΔK concept to microstructurally short cracks implies a violation of the assumption of continuum mechanics and linear elastic fracture mechanics, respectively.

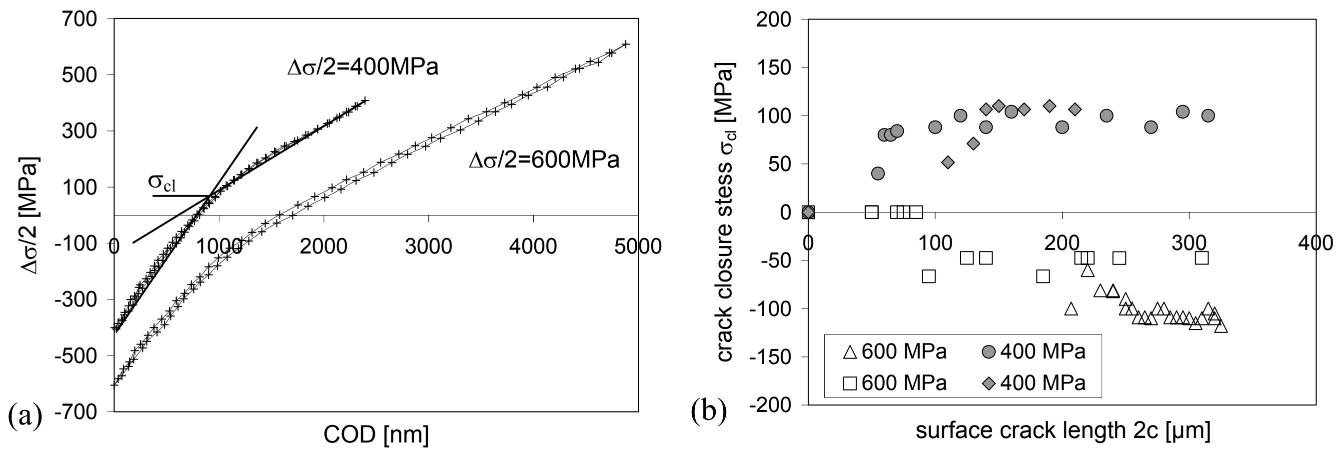


Figure 5: Evaluation of crack closure effects: stress vs. crack opening displacement (a) and crack closure stress vs. crack length (b)

CONCLUSIONS

The behaviour of short cracks in the beta-titanium alloy LCB in the solution heat treated condition is strongly dependent on the local crystallographic orientation. In-situ observation of growing cracks by an optical microscope during fatigue testing in combination with SEM examinations using the EBSD technique to determine local crystallographic orientations revealed that cracks initiate exclusively at high-angle grain boundaries either trans- or intergranularly. FE calculations of the microstructure on the base of the experimentally estimated value of the elastic anisotropy of A=0.7 yielded the highest stresses at these grain boundaries, which were actually observed to be cracked. Generally, favourable conditions/sites for crack propagation are:

- low angle grain boundaries,
- low angles between operative slip systems of neighbouring grains or intergranular cracks,
- slip planes parallel to the maximum shear stress (45° inclined to the loading axis).

The transition from short to long crack behaviour is substantially influenced by interactions and coalescence of short cracks. Measurements of the crack opening displacement by an ISDG system showed, that crack closure effects do not only influence the driving force for long cracks but also affect the propagation of microstructurally short cracks. Particularly if the applied stress amplitude exceeds approx. 50% of the yield stress of about 1200 MPa (at a stress ratio of R=-1) cracks tend to stay open in the beginning of the compression phase of the loading cycle. This can be attributed to the high crack tip deformation of short cracks as compared to long cracks and might be an additional reason, why short cracks frequently propagate much faster and at (nominally) lower stress intensity factors than the threshold value of the stress intensity factor of long cracks.

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