Scale Levels for Durability Bimodal Distribution of Titanium Alloys: Fatigue Mechanisms and In-Service Experience

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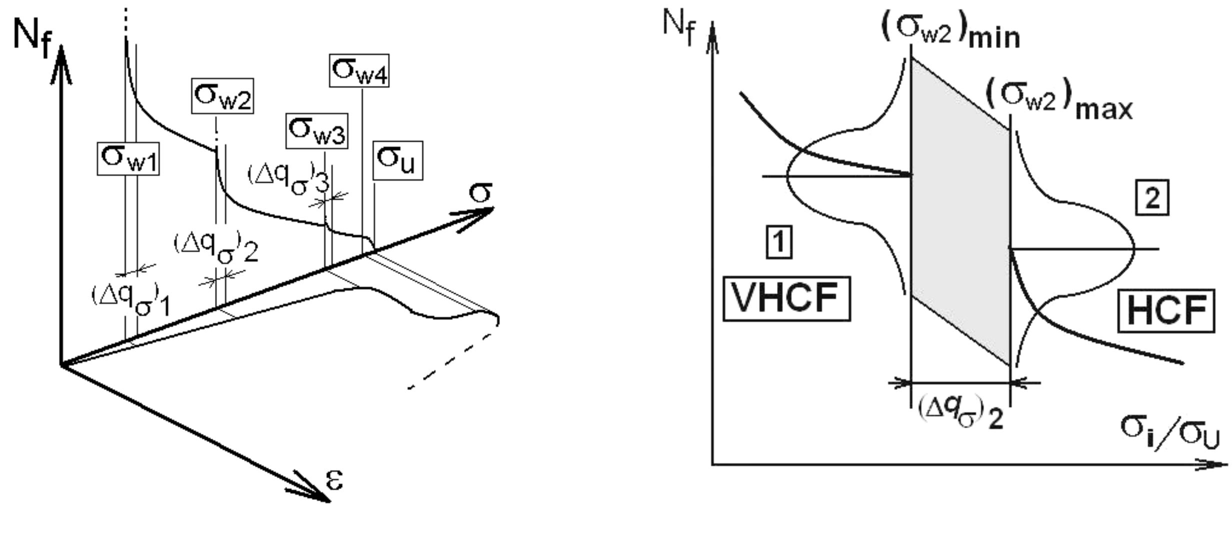
Abstract. Ti-alloys of different combinations such material characteristics as the α- and β-plate sizes in the lamellar or ($\alpha+\beta$) □□globular sizes microstructures and their various orientations to the external loading axis, the effective grain size and the metallurgical state with combinations of external cyclic loading conditions are consider in order to micro- or nano- (Very-High-Cycle-Fatigue), meso- (High-Cycle-Fatigue), and macro- (Low-Cycle-Fatigue) scale levels of material fatigue cracking. Based on the introduced earlier in consideration bifurcation diagram for fatigued metals, it was discussed causes of bimodal distributions of Ti-alloys durability with considering variations in features of fatigue fracture surfaces for different regimes of materials cracking. In many cases studied, fatigue cracking mechanisms have been suggested to explain many combinations of external conditions for cyclic loading and material reactions at the subsurface and at the surface crack origination and growth, when material sensitivity to interphase defects takes place under hold-time in cyclic loads. Models for fatigue cracks growing under hold-time conditions are discussed. All combinations for Ti-alloys cracking in different fatigue regimes have been illustrated in the case of in-service failures compressor blades and disks.

Introduction

Fatigue fracture process in metals usually taken into consideration applicably to different scale levels: micro- or nano- (Very-High-Cycle-Fatigue or VHCF-regime), meso- (High-Cycle-Fatigue or HCF-regime), and macro-scale level (Low-Cycle-Fatigue or LCF-regime) [1] - [4]. Transitions from one to another scale levels strongly expressed in accordance with introduced bifurcation diagram for metals [1], [3], Fig.1.

The bifurcation diagram allows describing fatigued metals behavior based on uniform methodological principle applicably to systems which evolution occurs far from the equilibrium position. Stated the synergetics concept allows to connect among themselves all experimentally demonstrated data on research of metals fatigue at different scale levels and to explain increase and decrease of dispersion of fatigue durability in process of increase of cyclic stress level at achievement of critical stress levels. Such manner of metal fatigue behavior consideration is lawful, even if in process of evolution metal undergoes only one unstable condition and consequently has only one bifurcation area between two boundary conditions when it is not loaded (one border) and when it is completely failed (the second border).

In the bifurcation areas fatigue durability distribution is bimodal. It means that at the same stress level of cyclic loads can be seen two possibilities in material manner of crack originates. The probability of each manner of crack origination has change in the direction of stress level increasing. At the left border of the bifurcation one mechanism has start and its probability of appearance is not far from zero.



a) b)

Fig.1. The bifurcation (a) diagram of metals fatigue has constructed according to the tension diagram [1]–[3] and (b) schema of probability of metals cracking variation for the bifurcation area $(\Delta q_\sigma)_2$. Bifurcation areas are specified at transitions from nano-$(\sigma_{w1}-\sigma_{w2})$ to meso-$(\sigma_{w2}-\sigma_{w3})$, and macro-$(\sigma_{w3}-\sigma_{w4})$ scale levels of fracture.

Then, for stress level increasing the new mechanism has probability increasing of its appearance but previous mechanism died and its probability of appearance decreases. That is why after the transition from bifurcation area, there is existed only one, new mechanism of material cracking. From the discussed consideration followed conclusion that bifurcation area reflects material instability in its behavior. In the range of stress level $(\Delta q_\sigma)_2$ probability of subsurface cracking, that dominated in VHCF-regime for material behavior as «In Part Closed System» (PCS) [3], decreased but crack origination at the surface, that dominated in the HCF-regime for «Open system» (OS), became more evident with probability 100% at the border of bifurcation area. In VHCF regime when duration to failure exceeds 10^9 cycles, behavior of metal appears close to thermodynamic system.

Instability increasing in material behavior at approaching to the border of bifurcation area has been demonstrated based on statistic analyses of fatigue durability data applicably to transition area from VHCF to HCF regime [4] - [7]. The specified situation can be considered based on experimental facts, which show that dispersion of durability increases for cyclic stress level increasing in the case of discussed transition. Increase of durability dispersion testifies to increase of instability in behavior of metals. Increase of instability of open system behavior during its evolution along any coordinate, in our case - a variable stress level, with other invariable parameters, testifies to approach bifurcation area. In the range of cyclic stresses where takes place transition from one distribution of fatigue durability described by the one Wöller-curve (or S-N curve) to another, there is exists bimodal distribution of fatigue durability described by the two different curves with differenced coefficients.

Below bimodal fatigue durability (BFD) distributions will be considered for Ti-based alloys on the different scale levels discussing influence of external cyclic loading shape in LCF regime. First step in the consideration will be done for bifurcation area $(\Delta q_\sigma)_2$.

Micro scale level for durability distribution in area of $(\Delta q_\sigma)_2$.

Specimens. Material behavior of Ti-alloy VT3-1 with two-phase mixed globular and lamellar (α + β) or dominantly lamellar microstructure typical of the disk material have been considered for tested bar specimens under 35 Hz [5], [8], [9] and 20 kHz [10]. Fatigue cracks origination was registered using acoustic emission (AE) monitoring [8]. Round bar specimens with 8mm in diameter have been made such as smooth (SS) and notched (SN) with circular notch of 2 mm in depth with stress raiser near to factor 1.46. Prepared specimens were then subjected to surface-hardening treatment with either hydraulic shot-peening (SP) by microballs with diameters in the range of (0.05-0.3 mm). The hardening degree was in the range of 1.1-1.27.

The performed investigations have been realized on notched and smooth specimens with the next combinations of their states: without heat treatment (WHT), with HT only, without SP (WSP) and HT, or with HT+SP.

Specimens were subjected to symmetric and asymmetric tension-compression with frequency of 35 Hz and 20 kHz (for dominantly globular (α + β) microstructure) at temperature 20°C on respectively hydraulic and ultrasonic test machines. The maximum stress level was in the range of (140-920) MPa with R-ratio in the range of (0.3-0.67) for tension and at -1.0 for tension-compression.

Results. The problem of subsurface fatigue cracking for titanium alloy VT3-1 has been discussed based on the acoustic emission (AE) monitoring material behavior at the moment of crack origination and short crack propagation [9]. The subsurface crack origination takes places in titanium alloys because of hydrostatic-like (three-axial) stress-state, material local volume compression and twisting under cyclic loads by Mode-III and gas diffusion in area of a first weakened facet (FWF), Fig.2. Intensive analyses of FWF orientation have shown that it is primary-grain slip by the basal plain [11].

Based on the knowledge about bimodal distribution of the Ti-alloy durability it was introduced method of tests data selection in the range of stresses $(\Delta q_\sigma)_2$ being transition area from VHCF to HCF regime [5], [9]. In this area there is can be seen tests data for material behavior that usually attested as not failed specimens in the range of $(10^6 - 10^7)$ cycles. It is clear that in some of tested specimens there can be crack initiation in this area but it is not clear where located place of origin: at specimen surface or subsurface.

That is why not failed in fatigue tests specimens were subjected to uniaxial tension up to fast fracture and, then, fracture surface were investigated in scanning electron microscope. Failed specimens in area of 10^6 where also subjected to fractographic analyses in electron microscope.

The received results of titanium alloy investigations for different conditions testify that the data of fatigue tests should be submitted in quality of bimodal distributions of fatigue durability as shown in Fig. 3. The left branch of S-N curve referred to cracks origination at the specimen surface, and the right branch referred to cracks origination subsurface. However, even in such representation it is visible, that processing of fatigue curves should be other and take into account bimodal character of durability distribution in respect to transition in crack origination at or subsurface of specimens. The S-N curve for smooth specimens placed lower, than for notched specimens. Such behavior should be carried to a different structural condition of a material, including to a different level of residual stresses in internal volumes of specimens that were manufactured from compressor disks.

However, it is obvious, what even in the case of fatigue cracks origination subsurface it is necessary to consider not one, and, at least, two more fatigue curves. Each curve reflects multiparameters character of interaction of the geometrical factor and the factor describing a material state (structural condition), including in connection with that or other level of residual stresses. It is cause of material cracking under Modes I+III crack opening in different directions out-off FWF.

In some cases of surface damages there can be seen multi-modes of durability distribution. This material behavior was discovered for the titanium alloy VT9 [5]. The material that was investigated had the ($\alpha + \beta$) globular microstructure. Tensile properties of the material were: ultimate tensile strength 1350MPa, yield strength 980 MPa and elongation 29%.

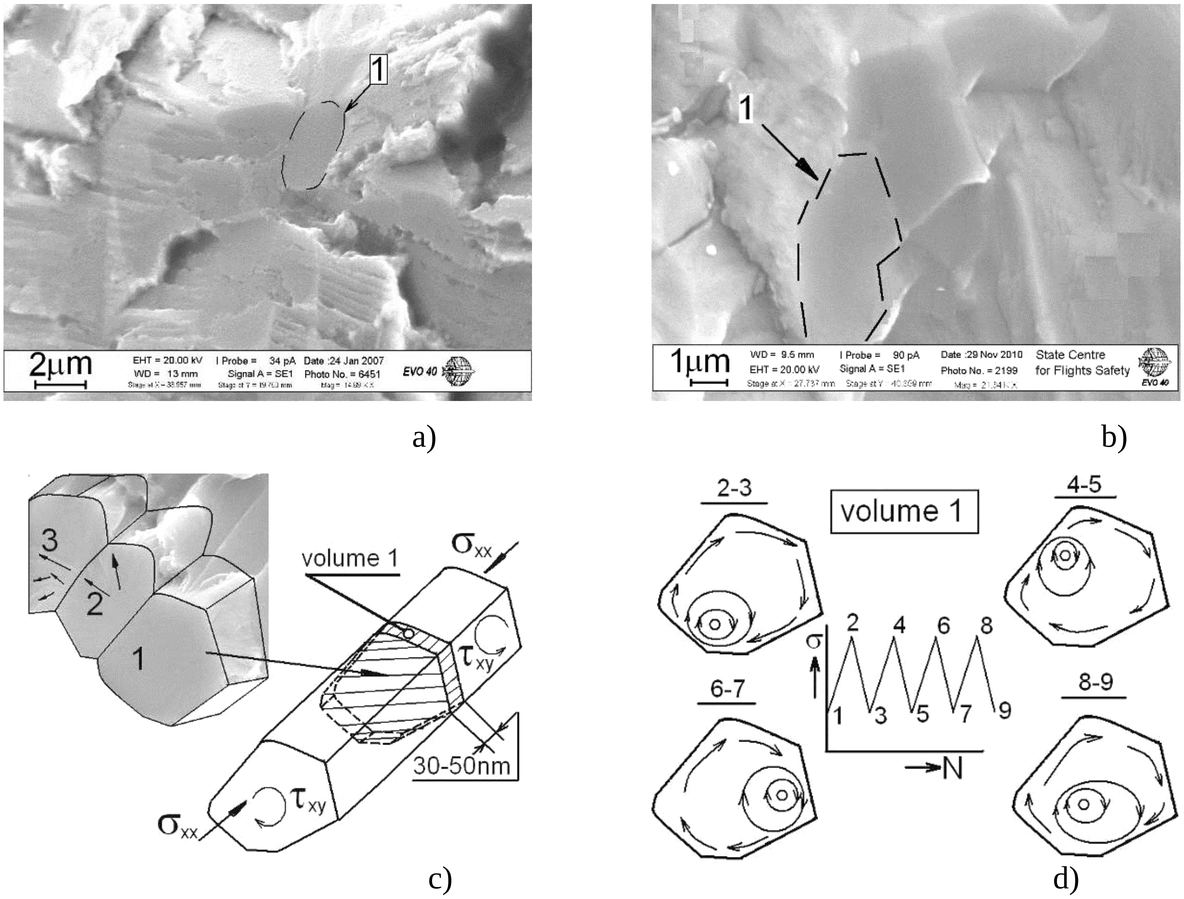
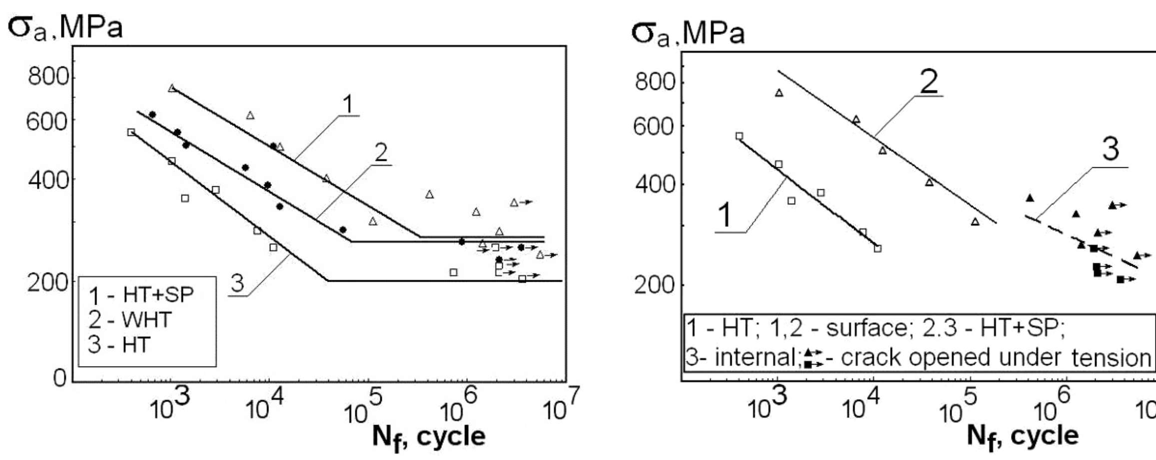


Figure 2. Fracture surface (a) with FWF, shown by circle, of VT3-1 alloy which is origin of subsurface cracking for the HT+SP smooth specimen at 35 Hz with durability $N_f=5x10^5$ cycles, (b) the similar FWF pattern for fatigued specimen at 20 kHz with durability 10^9 cycles [10], and (c), (d) sequential twisting events (schematic) in a material volume “1” takes place under compression which occurs in a structural element on the unloading portions of load cycles [8].



a) b)

Figure 3: S-N curves for notched specimens at R=-1.0 (a) and (b) their BFD after selection specimens by the criterion of the crack initiation at the surface or subsurface (internal).

Specimens of the titanium alloy VT9 were fatigued in the lab environment at the temperatures of 20^0C, 200^0C, and 500^0C. Fatigue tests were all rotating bending done with a 12000 rev/min (frequency 200Hz) loading frequency. Specimens were tested to final fracture at each of the various levels of cyclically applied stress. The stress level was arbitrarily defined as a fatigue limit when no fracture of specimens occurred before application of 5.10^8 loading cycles.
Test results have shown strongly expressed bimodal distribution of fatigue durability, Fig.4. The fractographic analysis has shown that crack initiation occurs at the tip of the scratches (stress concentrators). They are the first cause of the fatigue crack initiation from the specimen surface (see Fig.4c). The oxidized layer of the specimen surface is the second cause of the crack initiation. They influence fatigue crack initiation at the lower levels of the stress amplitude because of material brittleness, that is, a physical stress concentrator.

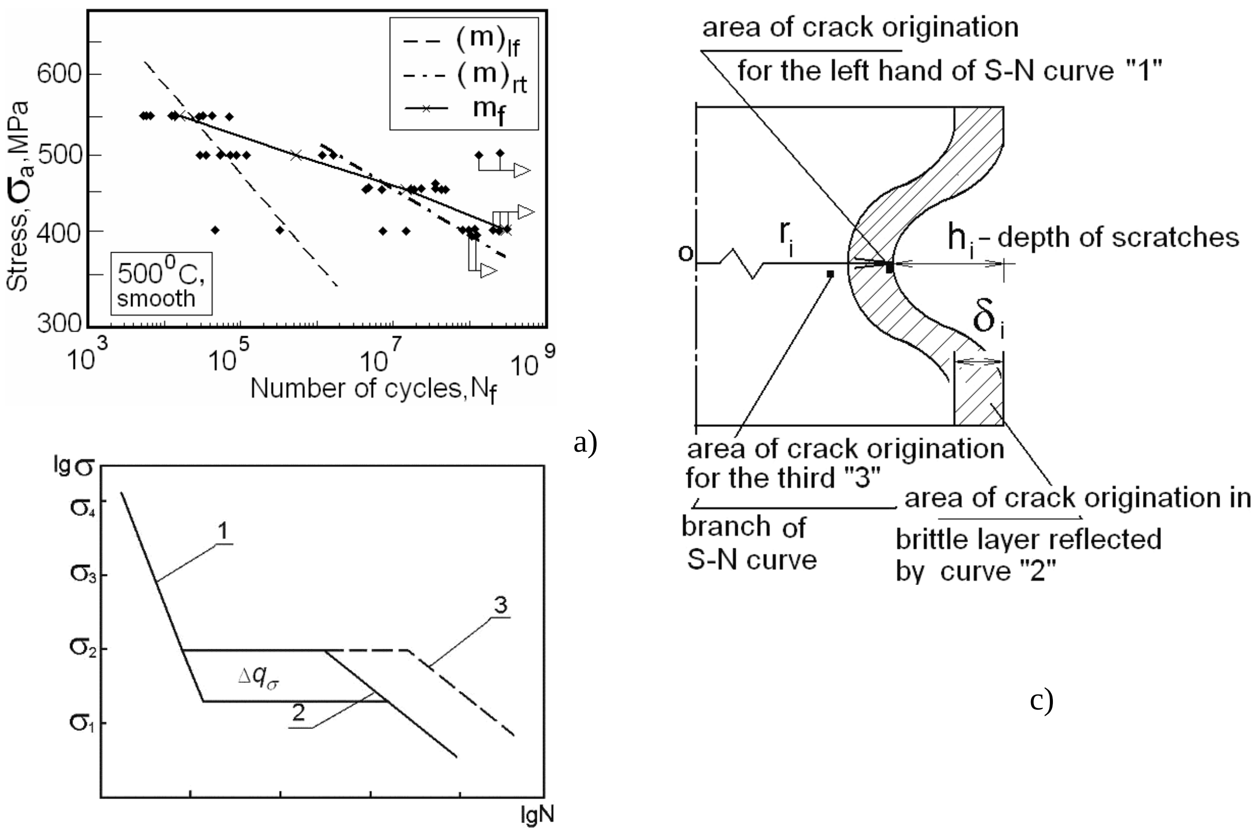


Figure 4. S-N curves (a) show the fatigue durability bimodal distribution discovered in smooth specimens of VT9 titanium alloy at the temperature 500^0C, (b) scheme of multi-modes durability distribution, and (c) comments (see in the text) for each S-N curve branches. Parameter $(m)_{lf}$, $(m)_{rf}$, and $(m)_f$ characterized respectively left-, right-, and mean S-N curves.

However, it was earlier crack initiation at the surface because of scratches and brittleness in the surface layer. In the case of the less stress concentration for lower depth of scratches the durability for crack initiation at the surface will be longer and earlier crack initiation takes place subsurface.
Therefore, in the bifurcation area for tested specimens a transition occurs from cracks initiated at the tip of a mechanical concentrator such as scratches (left hand of the S-N curve $(m)_{lf}$) to crack initiation in the thin brittle surface layer, which was physical concentrator for the titanium alloy VT9. The brittle layer dominantly influenced the crack initiation in spite of the presence of scratches and this influence is reflected in the right hand of the S-N curve with slope $(m)_{rt}$.

The first and second modes (left and right hand of the discovered S-N curve in the tests) characterize the durability distribution by the macro- and meso-scale levels for the crack initiation, which takes place from the material surface. The third mode of BFD shifted towards the greater fatigue lives, showing fatigue life to further increase with decreasing applied stress, appears characteristic of the subsurface fatigue cracking. This is micro scale level for fatigue crack origination when materials have self-organized change to their possibility to prevent fatigue cracking as a partly closed system.

Blades of aircraft engines. In-service blades fatigue failures are usually a result of their damage by a solid foreign object entering the engine [2], [12]. After a fatigue initiation the fracture surface appearance in operated rotor blades is modulated by the sequence of flight-by-flight cyclic loads. Cyclic loads in flight produce or not produce a crack increment, but a well-regulated appearance of some pattern on the fatigue surface in the crack growth direction is an evidence of a flight-by-flight regular repetition of a cyclic loads sequence.

For instance, there were several cases of compressor blades failures of aircraft engine NK-8-2U in flights [2]. In one of these cases fractured blade of the 7 stage of the high pressure compressor had good condition of the fatigue surface. This had allowed developing fractographic analysis to estimate a crack growth period. The blade had typical for Ti-alloy VT8 (Ti-6Al-3.6Mo-0.3Si) two-phase (α+β) globular structure. Material tests had shown a good condition of titanium alloy VT8 having recommended properties.

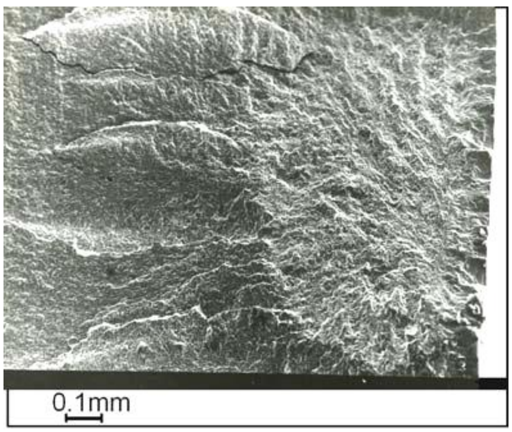
The crack development performed across the section placed on the distance near to 32 mm from the blade base. A stress raiser in the form of mechanical damage by a solid foreign object has been produced on the leading edge of blade. It had a dimple shape with sizes of 1.5 mm and 0.5 mm on the surface and in the depth direction respectively. The fatigue crack has propagated from above-mentioned stress raiser on the leading edge to trailing edge up to 34 mm. Then the fast fracture was produced throughout the rest section having the size of 10 mm in the crack growth direction.

In the case of blades in-service cracking without surface damages there is subsurface crack origination in them [13]. It is well-known phenomena of the subsurface fatigue crack initiation in the VHCF regime occurs for wide range of materials [1], [4], [5]. We should point out that the cracks defined as subsurface nucleate in an unpeened material and at stresses never above the fatigue limit: at this stress level cracking never began from the specimen surface earlier than after 10^8 loading cycles. A new (second) S—N curve, shifted towards the greater fatigue lives and showing fatigue life to further increase with decreasing applied stress, appears characteristic of the subsurface fatigue cracking. The second (longest life) branch (m_{rf}) appeared representative of the cracking initiated on the microscopic scale level, i.e., by the stress concentrators of microscopic dimensions. At this scale level, a metal behaves as PCS, to differ from an open synergistic system, whose energy exchange with the environment is unrestrained [1]. In our partly closed system, external actions would be ineffective with respect to the surface layers as long as the critical state, creative of a new crack, is achieved not at but below the material surface. Here we assume that the crack-initiation period remains the same whether for the surface-hardened or soft condition of the material. If in the bifurcation range, where the system selects, by self-organization, from the two ways of energy dissipation (crack initiation at the surface or in the subsurface region), these ways are equally probable.

For the titanium compressor blades, there was seen in-service blades failure with subsurface crack initiation. In previously performed investigations the more probable cause of these failures was related to the poor technological procedure used for blade repair. At the same time, the subsurface origin had significance in that titanium blades could have a critical state because of in-service critical lifetime to failure for the material in VHCF regime.

Cracks in blades had subsurface initiation sites, but the zone of initial crack propagation up to its intersection with the surface had no structural or chemical anomalies in accordance with requirements of the manufacturing procedure for the titanium alloy VT3-1. The microstructure of material

blades was globular with an $(\alpha + \beta)$ two-phase microstructure, and no other defects in structure were observed. Hence, in-flight fatigue failure of blades is not connected to material quality.
Cases of blades failures were rare, and for this reason, it was obvious that the specified group of blades has some features of their in-service loading that led to cracks initiating subsurface (see Figure 1). For example, the fatigued blades №14 and №25 had flown 8906 and 9919 hours, including 1262 and 2034 hours after the last repair, respectively.
The center of the origin area in both blades has taken place under their surface on the distance about 120μm. The area of origin had extent along a surface about 0.85mm and 1.0mm and in depth from a surface about 0.7mm and 0.6mm accordingly for the blade №25 and №14. There was formed quasi-brittle pattern of the fracture surface in area of the origin, which reflects the $(\alpha + \beta)$ globular microstructure, Fig. 5. The crack propagation represents the cascade of facets of quasi-cleavage, each of which is limited to the globule size of two-phase material structures.



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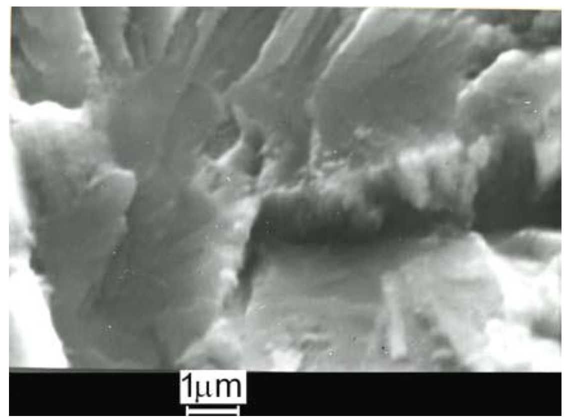


b)

Figure 5. Overview of the fracture surface patterns (a), (b) in area of origin of the blade №25 with center of the crack origination (b) placed under the surface. Area of the crack origination is situated on the photos on the right-hand

The state analysis of bandage-shelves (BS) has shown for both engines that they have damages because of intensive wear by surface contact. The depth of the material damage was so deep that blades could lose contact with other blades, leading to the free blade vibrations at a high frequency under the air stream. The free vibration frequency for blades can exceed 1 kHz. At this frequency, for one of the blades, the number of cycles with the increased stress level and for in service lifetime 1262 hours from the last repair is not less than $1262\text{x}3600\text{x}1000 = 4.5\text{x}10^9$ cycles. It is evident, that the blade fatigue fracture occurred in the VHCF regime.
The similar sizes of zones of initial development of subsurface cracks for different engines indicate that the material stress-state for blades in the crack initiation region was identical.
Based on the crack circular form of the crack front, it is possible to estimate the equivalent stress for average depth of a crack about 0.7mm by the formula:

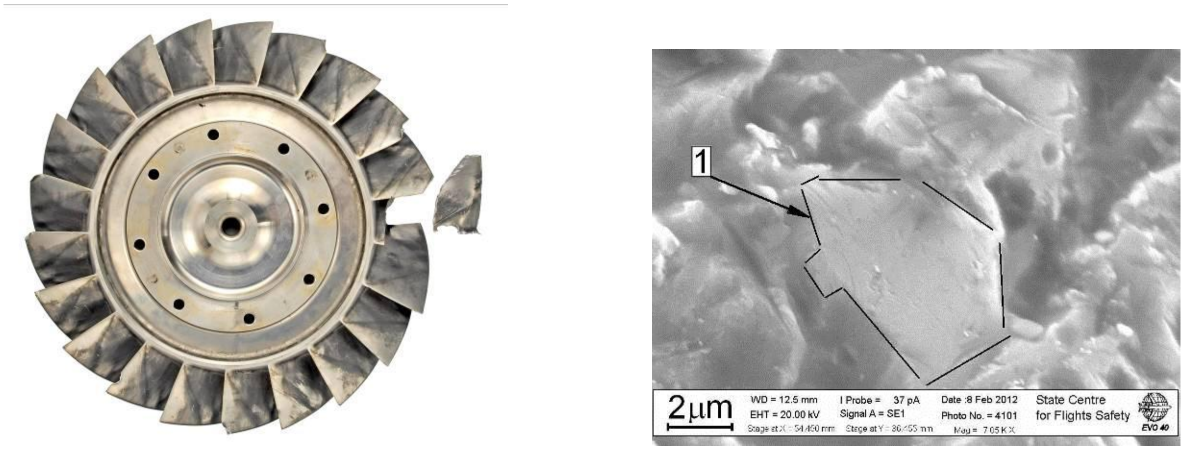
$$\sigma_e = K_{th} / 1.12(\pi\, 7.10^{-5})^{1/2} = 3.7/0.055 = 67\text{MPa} \qquad (1)$$

The stress intensity factor K_{th} =3.7 MPa $m^{1/2}$ for titanium alloy VT3-1 was taken from the paper [1]. The estimated stress equivalent value of 65MPa is in four times lower than fatigue limit of the titanium alloy VT3-1. It shows that blade fatigue fracture occurred really in area of VHCF may be under the resonance condition.

Thus, the subsurface origin is significant in that it has been loaded with high frequency and high amplitude in the VHCF regime. Such situation was possible only because of loss of constrained contact by BSs.

Subsurface crack initiation was registered for blades of VT8 Ti-alloy manufactured as one structural component with disk of the engine first compressor stage. Only one blade had experience fatigue cracking for one cracked disk.

Crack origination taken place not far from the blade surface and there FWF formation in different fracture surface areas were seen, Fig.6.



a) b)

Figure 6. View (a) of the in-service failed disk of VT3-1 titanium alloy of the first stage of an aircraft engine compressor with a separated blade and (b) FWF of subsurface crack origination area in the blade fracture surface.

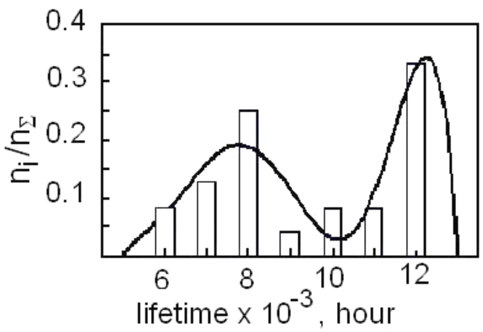
One can see that multiple sites of subsurface cracking were discovered because of material weakness under external loading in simultaneously performed different local volumes. This case can be considered as the same situation that has been registered earlier in tested specimens at 35 Hz for VT3-1 Ti-alloy [9] inspite of discussed blades have been manufactured from VT8 Ti-alloy and had subjected to cyclic loads at the frequency in the range of 1550-1700 Hz.

So, one can see that bimodal distribution of Ti-alloys durability on the micro- or nano-scale level takes place because of transition from subsurface to at the surface crack origination. Multimodal distribution of durability in the bifurcation area $(\Delta q_{\sigma})_2$ occurred because of difference in material conditions by the surface which influenced crack origination for Ti-based alloys at the surface on the different scale levels.

Macro scale level for BFD distribution in the rang of stresses $(\sigma_{w3}-\sigma_{w4})$.

Data for Ti-based alloys in-service fatigue cracking in the range of stresses $(\sigma_{w3}-\sigma_{w4})$ or in LCF regime related to compressor disks of different stages [1]. Usually service lives of titanium disks designed on the basis of the “safe life” principle are established based on the LCF criterion [14]. Cyclic loads sequence for gas-turbine-engines (GTE) working per flight is transformed in one or two cycles per flight to be equivalent for material damage [14], [15]. But in real disks damage accumulation can be produced not only by LCF but also by HCF or mixed LCF/HCF regimes. Kind of criteria which must be used to determine disks durability are not evident for designed titanium disks of various stages for different GTE.

Usually in-service fatigue cracks distribution in time for various aircraft components has two maximums [1], [2], [16]. The first maximum correlates to the material faults and the second reproduces the mean value of durability dispersion for structural components. In contrary for titanium disks the cracks distribution has three or four maximums. For instance, there were two maximums of fatigue cracks distribution of in-service compressor disks of Ti-based alloys manufactured for GTE of D-30 and D-30KU, as shown in Fig.7.



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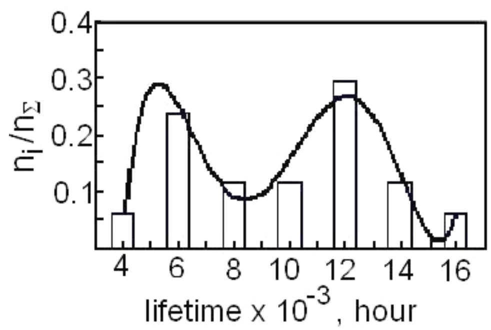


b)

Figure 7. Cracks bimodal in-service time distribution in Ti-compressor disks of GTE (a) D-30 and (b) D-30KU with two maximums: first maximum related to damaged materials and second maximum point out behavior of not damaged materials with high level of their sensitiveness to hold-time [16].

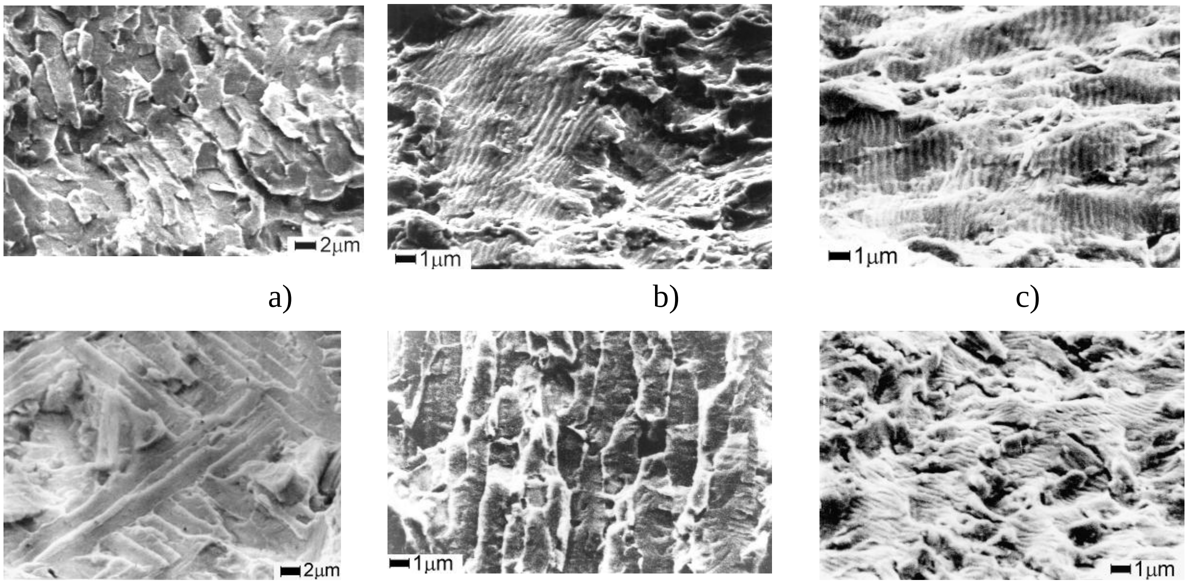
The second maximum did not reproduce the disks expected durability. It was durability for not accounted case of material fatigue cracking when the second branch of S-N curve takes place because of material sensitiveness to the cyclic loads in-flight working GTE. For all branches of S-N curves there will be the third maximum of cracks distribution for in-service disks expected durability that is shown in Fig.7b for the second type of disk failed in the HCF regime as a tendency for a new fatigue cracks distribution. One can see for in-service titanium disks that there is not the same situation with crack inspection in them as for other components because a relation between duration of cracks nucleation and propagation is not the same for one and another branch of S-N curves.

Below the problem of bimodal fatigue lifetime distribution for Ti-alloys because of hold-time influence on their behavior will be sequentially considered.

Fatigue tests results. The bimodal distribution of fatigue lifetime for Ti-alloys in LCF regime was considered in different cases studied applicably to different shapes of specimens and compressor disks [16] – [23]. It should be pointed out that all variations in fatigue mechanisms can be seen in one melt of material that followed the recommended practice for composition and mechanical properties, being manufactured by a unified compressor disks manufacturing technology. Metallurgical analyses have not shown any principal difference in values of structural parameters (effective grain size, real grain size, and sizes of α- and β-lamellar or globular structures) in cases studies of specimens of titanium disks. It was shown:

(1) Ti-alloys have to be divided into three groups by their sensitiveness to external cyclic loads, Fig.8; the discussed material sensitiveness was the cause of the fracture surface pattern appearance that was accompanied by mainly faceted pattern formation on the fatigue fracture surface with very local and rare places of blocks of fatigue striations for the material first and the second group; for the first group material has sensitiveness to cyclic loads at both triangular and trapezoidal shape that is why fatigue fracture surfaces had mainly faceted pattern as a result of crack development in two-phase lamellar or globular structures (the fatigue striation blocks occupied not more than 5% of the fatigue surface); for the second group, material has sensitiveness to cyclic loads of trapezoidal shape only that is why facetted pattern was dominant in the case of trapezium-shaped waveform of cyclic

loads (blocks of the fatigue striations occupied not mot that 20% of all the fatigue surface); for the third group, there is no material sensitiveness to the cyclic loads of different shapes (cyclic loads of both shapes mainly developed a fatigue striation pattern, i.e. the striation pattern occupied more than 80% of the fatigue surface;



d) e) f)

Figure 8. Fracture surface patterns for tested specimens from: (a), (d) first group; (b), (e) second group (c), (f); for triangular (a, b, c) and trapezoidal (d, e, f) cyclic waveform. Specimens manufactured from compressor disks of VT8 titanium alloy.

(2) materials behavior of the first and the second group depended on the duration of hold-time applicably to both crack initiation and propagation periods, Table 2, [17];

Table 1. Durability N_d and crack growth period N_m (measured) and N_c (calculated by striation spacing) at different cycle shape.

N_i, cycles	**Frequency (Hz) or hold-time (s)**				
	1Hz	0.2Hz	3s	20s	60s
N_d	14790	9014	6840	3416	3730
N_m	10790	8000	2840	2500	2890
N_c	12800	7300	2800	2850	2900

difference in crack growth rate for triangular and trapezoidal shape of cyclic loads can be till 10 times for hold-time duration 60s; programmed loading brings about material fracture with formation of chiefly facet pattern and fragmentary striation relief; the calculated number of fatigue striations characterizes the actual material damage which occurs repeatedly in a programmed loading block for different programs with correlations 1:1, 2:1 and 3:1;
(3) the synergetic principle of a self-organized selection of crack growth and deformation processes within the material around the crack tip may prevent the crack increment increasing under various cyclic load conditions. The variation in crack growth increment can be expressed as follows:

$$\delta_h = \delta_\tau + \Delta_i \tag{2}$$

where $\Delta_i = f(PB) + f(r_i)$ (3)

In Eqs (2) and (3), δ_h is the full crack growth increment; δ_τ is a crack increment for a triangular cyclic waveform; Δ_i is a crack increment because of a plastic blunting (PB) process and the plastic zone size increasing at a crack tip; and terms $f(PB)$ and $f(r_i)$ are functions of the developments of the PB process and the plastic zone size, r_i, respectively. The function $f(r_i)$ can cause a small increase under the hold-time condition so that the crack increment, δ_h, can be the same as δ_τ (or Δ_i =0), or a little smaller than is the case for the triangular shape of cyclic loading. Similar results were seen in tests of Ti-alloys 6242S (Ti-6Al-2Sn-4Cr-2Mo-0.1Si) and 5621S (Ti-5Al-6Sn-2Zr-0.8Mo-0,25Si) conducted under a hold-time period of 1 h at the elevated temperature of 450^0C [24]. A principle difference in growth rates was not discovered for the hold-time and the triangular waveform of cyclic loading at a frequency 0.17 Hz;

(4) material anisotropy can be one of the cause of BFD distribution for in-service lifetime [20]; crack propagation in the longitudinal direction for material fibers has the manner that, for example, shown in Fig. 8e, but in the transversal direction crack propagation took place with fatigue striation formation; this behavior was explained based on biaxial loading fibers, when biaxial tension deceased material plastic deformation for the longitudinal direction and it decreases in the transversal direction under the same stress state for metals;

(5) hardening process through all the material section can be very effective procedure for reorganization of material state with high level of residual stresses inside [22]; this reorganization directed to change position of the S-N curve branch, which reflects not hardened material sensitiveness to hold-time, with shifting its position after hardening procedure to the S-N curve, which was discovered under triangular shape of cyclic loads before material hardening; in the discussed case, after the hardening process, subsurface crack origination and growing is dominant material behavior under different shape of cyclic loads in LCF regime;

(6) material sensitiveness to the hold-time cannot be estimated by the Charpy-tests (CCT) being one of the method to study a material behavior for titanium compressor disks of aircraft engines; the technological order to material condition is such for in-service used disks whose material has not a fracture work CCT less than 0.8 kgcm/cm^2 when in specimens preliminary introduced fatigue cracks in them. This critical value was recommended for practice to manufacture titanium compressor disks at that time when Ti-alloys sensitiveness to a hold time was not yet discovered and faceted pattern relief of fatigue fracture surface which reflected this sensitiveness was not seen; results of comparing material behavior for different values of CCT have shown that there is not correlation between CCT value and material sensitiveness to dwell-time for VT3-1 and VT8 Ti-alloys, Table 2.

Table 2. Data of the CCT-test result for Ti-alloys VT3-1 and VT8 compared with fracture surface facetted pattern percent which dominates in the case of material sensitiveness to hold-time.

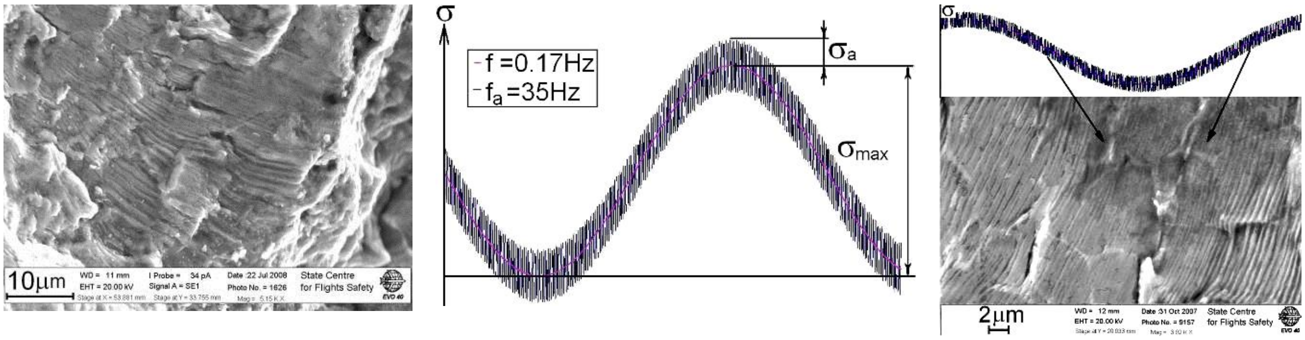
Ti – alloys	VT3-1				VT8		
CCT, kg·cm/cm^2	0.38	0.49	0.82	1.10	1.02	1.50	1.91
Percent of the faceted pattern	95	40	80	5	95	5	80

All possible variations for bimodal in-service lifetime distributions for structures manufactured from different type were systematized and methodology for non-destructive-testing intervals was introduced [16], [19].
Variation of situations can be seen for the fatigue fracture of Ti-alloys of the same technology. Very large variations of the realized material behavior can be seen for the alloys that influence crack increment per one flight. This should be evaluated by the factor $\mathbf{k_{FCGR}}$ that reduces crack increment per flight. One can see that various stages of the same GTE's compressor can have different material states. In this case one of the disks will be first of those whose fatigue limit will be exceeded in service. It will be disk with material sensitive to cyclic loads. That is why all disks should be ranged by the material state because of durability has two peaks of durability distribution or BDF.
Titanium disks tests under cyclic loads sequences imitating loads sequences per flight for several GTE have shown that the material sensitiveness produces one, two, or three striations in a local place of fracture surface (see n.2 above) for mainly faceted pattern in dependence on the type of cyclic loads block [17]. This variation can be evaluated by the factor $\mathbf{k_{FCL}}$.
The process of the fatigue striations formation depends on material state variations. That is why in LCF regime there is very large dispersion of the fatigue striation spacing for titanium alloys [18]. In this case the striation spacing does not equal to the crack increment per one cycle under cyclic loads with constant amplitude. This fact should be evaluated by factor $\mathbf{k_v/\delta}$ to estimate the crack growth life in a number of flights, were δ is striation spacing.
Summarizing factors correction one can see that crack increment in disks per one flight calculated on the basis of fatigue striations measurement must be decreased by the factor

$$n_e = k_{FCGR}.k_{FCL}.k_v/\delta \qquad (4)$$

There is not only one way of disks damage display per one flight in service for all engines or for various compressor stages for the same type of engine. For example, because of in-service disk failure develops under a hold-time with long duration, a factor 1.6 is recommended to decrease number of cycles to failure calculated on the basis of striation spacing measurements for the case of mainly striationed pattern formation on the fracture surface. Consequently, this problem has more complicated evidence because of various fatigue mechanisms that had realization in disks. In each case we should know an equivalent for crack increment per flight estimated from the quantitative fractographic analyses on the basis of the fatigue striations measurements using Eq.(4).
In-service experience. Different cases of titanium disks in-service cracking were studied and one can be conclude that all crack growth period estimations carried out to exclude in-flight disks failure inspite of their durability bimodal distribution [1], [2], [16] – [26]. Summarizing results of in-service titanium disks fatigue cracking it can be seen that:
(**1**) In the hub area of the disks there is can be seen BFD because of material state in the range of stresses $(\sigma_{w3} - \sigma_{w4})$ for LCF regime (macro-scale level).
(**2**) In the rime area there blade vibrations influence on the BFD exists. In the dependence on maximum stress level and range of cyclic loads amplitudes can be realized dominant LCF or HCF regime, and, also, combinations of domination one or another regime.
(**3**) In the rim area there can be seen UHCF with subsurface cracking or HCF regime because of high level of vibrations from blades (micro-scale levels).
In the complicated case of HCF regime influence on the LCF disks fatigue cracking there is existed acceleration in both crack initiation and propagation periods [25], [26]. There was discovered for in-service disks regular changes in facetted pattern surface and areas with blocks of fatigue striations in the crack growth direction. Typical peculiarity for blocks of fatigue striations was of their spacing

variation in each block, Fig.9. This regularity was compared with regularity of material cracking in-tests with variation mean stress level for constant amplitude of cyclic loads.



a) b) c)

Figure 9. Fracture surface (a) with regularities of fatigue striations sequence of the in-service fatigued titanium compressor disk, (b) cyclic loads sequence for tested specimens of VT3-1 Ti-alloy, and (c) fracture surface with fatigue striations sequence in one of the area with comparing of the in-tests cyclic loads sequence [25], [26].

One can see that fracture surface features are the same for in-service failed disk and in-test fatigued specimens under variable amplitudes of cyclic loading. This external bimodal-type of cyclic loads influenced for disk BFD because of existing possibility to realize fatigue cracking in LCF or HCF regime (by one or another branches of S-N curve).

Nevertheless, based on the BFD of the titanium alloys occurred on the different scale levels one can be conclude that their in-service durability estimation have to be changed in the case of damage accumulation summarizing. An equivalent cycle of material loading for safety lifetime in-tests determining has to be included itself long hold-time. Introduced earlier methods for equivalent cycle of in-tests loading titanium alloys forming [1, 2] have to be improved with introduction difference for one and another branches of fatigue curves of BFD in dependence on steps of in-flight disks operation with consideration long time of unchangeable GTE in-flight rotation.

Sequence of cyclic loads in one flight has to be separated by the duration-criterion of material loading in one cycle or by the criterion of "cycle shape" (considering triangular shape of cyclic loads). It means that cycles with triangular shape, which duration is not longer than $0.1s^{-1}$, ought to be consider for right branch of the BFD. This case is the same that has been discussed earlier [14], [15]. Cycles with duration more than $0.1\ s^{-1}$ have to be considered based on the left branch of BFD. Each cycle with hold-time, τ, considers in accordance with relation: $\Sigma\ n_i = \Sigma n_\tau f(\tau)$, where n_i – number triangular cycles are being equivalents to cycles with hold-time n_τ□, when correction factor $f(\tau)$ on the cyclic loads duration was introduced including consideration durability variations because of difference in loads hold-time.

In the case of fatigue-test equivalent cycle preparing there can be recommended simply method of discovered in-tests durability if one of the well-known methods of cyclic loads summarizing was used [14], [15]. Discovered in-tests durability have to be decreased on the factor $f(\tau)$. This factor is ratio between durability for right and left branches of BFD. Usually this factor ranges between 2 and 5.

Summary

Bimodal fatigue durability distribution in Ti-based alloys can be seen in different ranges of cyclic stresses. Main case of this material behavior is related to bifurcation area $(\Delta q_\sigma)_2$ being transition area from VHCF to HCF regime. There takes place transition from the subsurface to at the surface fatigue crack origination for stress level increasing. Subsurface material cracking starts from the first-flat-facet occurred under pressing and rotation or twisting local material volume in the one of the structural element border.

In some cases in the range of $(\Delta q_{\sigma})_2$ can be realized multi-modal fatigue durability distribution because of difference in material sensitiveness to the cyclic loads on the different scale levels from the surface damages.

In-service fatigue cracking of titanium compressor blades in VHCF regime has starting from the first-flat-facet with the same fracture surface patterns formation that revealed for specimens with subsurface cracking.

In LCF regime BFD distribution takes place because of material structural sensitivity to cyclic loads waveforms applicably to two-phase lamellar and globular material structures. This effect can be seen for different values of the fracture energy as determined by Charpy-tests.

Titanium disks with a microstructural inheritance derived from production processes giving a filament-type have a short in-service lifetime because of the material sensitivity to the cyclic loads waveforms when the fatigue crack growth in the parallel direction to the filament planes. The fatigue crack growth on the perpendicular direction to the filaments planes in 2-5 times longer than in the parallel direction to these planes because of the mainly fatigue striation process under various waveforms of cyclic loads.

A synergetics approach has been used to explain structural sensitivity to fatigue crack growth in titanium alloys subjected to cyclic loads with different waveforms which directed to bimodal fatigue durability distribution. Crack growth in disks materials is simultaneously influenced by the effects of crystallographic orientations of lamellar packets, residual stresses, and effective grain sizes of the lamellar or globular structures and thin interphase layers of inclusions. Combinations of many factors simultaneously influence the material sensitivity to interphase fatigue crack growth. Each one of these factors can not be enough when considering the bimodal fatigue durability distribution. Only the hold-time allows one to discover this sensitivity but can not be only factor to influence that sensitivity.

Tests with a hold-time in the cyclic loads are recommended when selecting titanium disks which have roughly similar structural sensitivity to fatigue crack growth. These disks can be used for the compressor main parts, if they have a roughly equal service lifetime.

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