Crack path of titanium alloy Ti-6Al-3Mo-0.4Si in high- and very-high-cycle-fatigue regime after tempering and hardening procedures of smooth and notched specimens

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ABSTRACT. The influence of different technological procedures on titanium alloy VT3-1 behaviour has investigated in the range of 10^4 – 5.10^7 cycles under the specimens' tension with frequency 35Hr. Smooth and notched round bar specimens have been tempered and hardened or subjected to tests without one or both technological procedures. Tests were performed in the range of maximum stress level 140-920MPa at the R-ratio from -1.0 till + 0.67. It was revealed the stress range when in material behaviour takes place transition from cracks origination at the surface to the subsurface. The subsurface area of the crack origination for smooth and notched specimens is the facetted pattern of the material quasi-cleavage. The Mode-III mechanism of twisting under three-axial compression is discussed to explain the way of subsurface origination of the first smooth fracture facet (fatigue crack origin) followed by creating a plastic zone around the crack origin area, weakened by the fist facet. Material cracking out-off subsurface origin determines by combined Modes I+III mechanism. The critical values of τ_w and σ_w for the crack subsurface origination have been estimated and it was shown that ratio τ_w / σ_w for the first weakened smooth facet of titanium alloy VT3-1 ranged between 2.37 and 2.6. The fatigue curves were reconstructed and bimodal distribution of material durability N_f was demonstrated. The crack growth period, N_p , has unified description for all combinations of specimens surface states and material conditions in term of N_p / N_f versus N_f .

KEYWORDS. Titanium alloy; Tempering; Hardening; External tension; Bimodal distribution; Durability; Fatigue cracking; Subsurface; Modes-III; Crack growth; Combined modes I+III; Fatigue striations.

INTRODUCTION

Titanium alloys which used for manufacturing compressor disks of aircraft engines have dependence of their behaviour under cyclic loads on a stress level and cyclic loads duration [1 - 3]. The hardening procedure of titanium alloys that influenced on their behaviour in low-cycle-fatigue (LCF) has been executed with reference to titanium alloy Ti-6Al-2Sn-4Zr-2Mo-0.1Si with lamellar and globular structures [3]. It has shown that increase of durability after through hardening specimens takes place because of increase in a period of a fatigue crack origination. Development of a crack occurs subsurface, and after an output of a crack on a specimen surface, its lifetime to fracture makes some percent from durability, N_f.

The problem of subsurface fatigue cracking for titanium alloy VT3-1 has been discussed based on the acoustic emission (AE) monitoring material behaviour at the moment of crack origination and short crack propagation [4]. The subsurface crack origination takes places in titanium alloys because of hydrostatic (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). Intensive analyses of FWF orientation have shown that it is primary-grain slip by the basal plain [5].

To understand in-service fatigue cracking material with crack origination at the surface or subsurface it was introduced bifurcation diagram [6 - 8]. We should point out that the cracks defined as subsurface nucleate in a nonstrengthened material and at the stresses never above the fatigue stress σ_{w2}: at this stress level cracking never began at the specimen surface earlier than after 10^6—10^7 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. This new area for materials fatigue behaviour was named Very-High-Cycle-Fatigue (VHCF) [9] or Gigacycle-Fatigue [10].

Discussed below is the fatigue-fracture (crack initiation and propagation) behaviour of the round bar surface-hardened and tempered or soft specimens of titanium alloy VT3-1. The smooth or notched specimens were tested and S-N curves were constructed. Researches were carrying out to analyse material behaviour in wide range of cyclic loading conditions with different asymmetry down to durability of 5.10^7 cycles and to reveal bimodal distribution of the material durability.

EXPERIMENTAL PROCEDURES

Tests were performed on titanium alloy VT3-1 which is widely used in manufacturing compressors disks of aircraft engines. Specimens for tests have been cut out from the disks of the second stage one of the engine compressor. One part of manufactured specimens has the same material state that the manufactured disk. Another part of manufactured specimens has been subjected to heat treatment (HT) with tempering at T = 530°C within 6 hours, and then cooling on air. The material ultimate tensile stress for all specimens was in the range of 1040-1100 MPa at elongation and section area reduction (10.8-16) % and (3.1-4.7) %, respectively.

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.

Therefore the performed investigation has been realized on 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 at temperature 20°C on the hydraulic test machine. The maximum stress level was in the range of (140-920) MPa with stress ratio *R* in the range of (0.3-0.67) for tension and at -1.0 for tension-compression.

All specimens after the tests were investigated in the scanning electron microscope of the firm “Karl Zeiss”. Fracture surfaces were investigated and fatigue striation spacing was measured. Then the analysis of structure of titanium alloy VT3-1 on several specimens has been executed in the regime of reflects electrons.

RESULTS OF SPECIMEN INVESTIGATIONS

Test results have shown that the fatigue-limit of smooth specimens without heat treatment being 340 MPa reduces up to 300 MPa after HT on the bases of $2{\cdot}10^7$ cycles. The fatigue-limit reduces up to 130 MPa at various R-ratios with mean stress level σ_m = 600 MPa, i.e. in 2.3 times in comparison with a symmetric cycle.

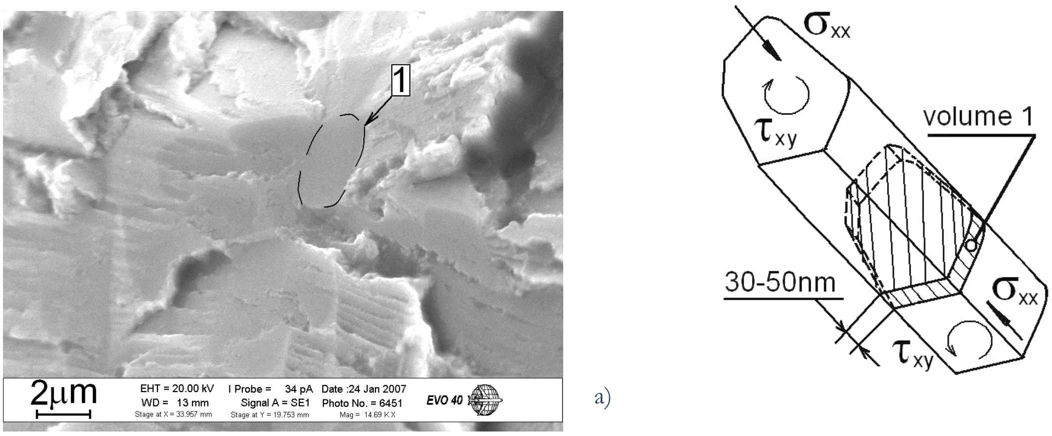
SP of smooth specimens increased durability at small numbers of cycles. Comparison in tests results for peened and unpeened specimens at 10^6 cycles reflects difference in stress level more than 1.3 times.

Test results of notched specimens have shown that they have approximately the same value of fatigue-limit for peened and unpeened surface. Lower fatigue-limit discovered for specimens after HT.

The fractographic analysis has shown, that is direct subsurface has been generated facet of quasi-cleavage, having the significant area of a surface, Fig. 1. It was performed because of destruction of globular or lamellar structure of a material in area of FWF. The weakness occurs under compression and rotation of a local volume of the cracked material [4].

The received result of fractographic analysis shows that in the range of durability 10^6 - 10^7 cycles the probability of fatigue crack origination subsurface grows for increasing durability of smooth specimens with different R-ratios. Therefore, it was

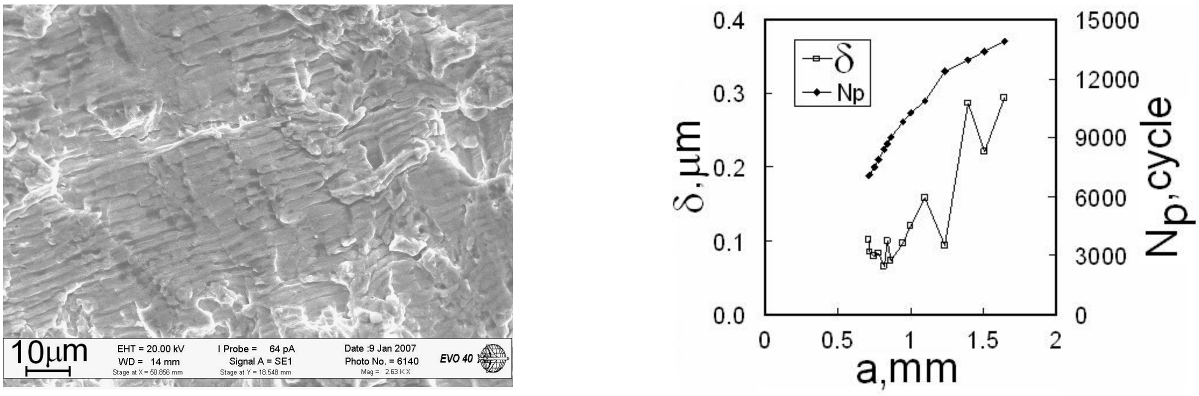
important to analyse influence hardening procedure on the material behaviour which can affect fatigue crack origin subsurface.
Let be considered group of peened specimens are tested up to fracture in the range of durability 2.9x10^5 - 1.2x10^7 cycles.
In all specimens, a fatigue crack origination occurred subsurface. At the lower durability there was performed several areas of crack origins subsurface. It is typical, that the centres of origins are formed by different structural elements of a material. The area of origins represents the extended site quasi-cleavage with oxides on a surface and around of it, which reflects lamellar or globular structure of a material. The Mode III cracking is evident process for the FWF formation (see Fig.1).



a) b)

Figure 1: Fracture surface (a) with FWF, shown by circle, which is origin of subsurface cracking for the HT+SP smooth specimen with durability N_f=5x10^5 cycles, and (b) schema of the FWF creation [4].

In the direction of the fatigue crack propagation, there were formed fatigue striations, Fig.2. Crack growth period estimations have shown the tendency of decrease in the crack growth period at the stage of fatigue striations formation with maximum stress level increasing. However, it is necessary to emphasize, that the crack growth period, obviously, grows with increase of durability.



a) b)

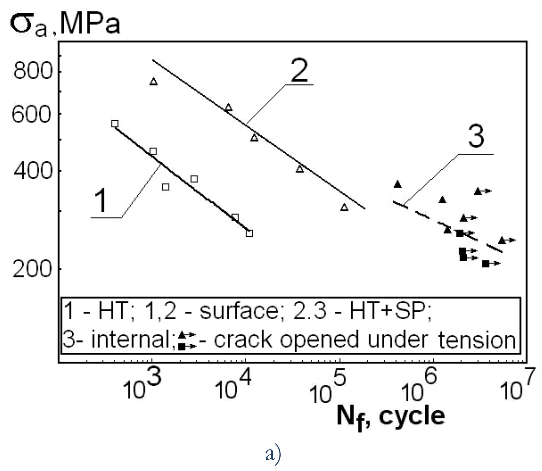
Figure 2: Fracture surface (a) with fatigue striations and (b) striation spacing δ with number of crack growth cycles N_p versus crack growth length, a, for (a) the HT+SP notched specimen tested at σ_{max}=300 MPa, R=-1, with durability 1.2x10^6 cycles, and (b) the SMD notched specimen tested at σ_{max}=770 MPa, R=0.56, with durability 9.5x10^5 cycles.

The received result of the expressed origin fatigue cracks subsurface in the smooth peened specimens, within the bifurcation area has been compared with test results of notched specimens.
The first group of unpeened specimens without tempering that was considered, have been tested at R>0. Their durability was in the range of 9.5x10^5 - 2.0x10^6 cycles.

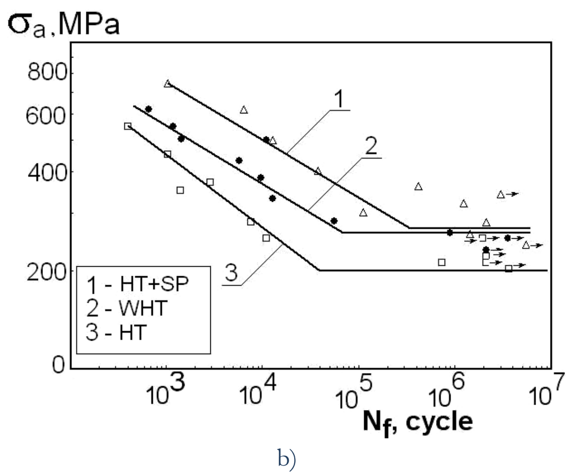
Only two specimens have formed subsurface origins on the depth near to 25μm had having durability of $1.4x10^6$ and $2.4x10^6$ cycles. There was facetted pattern relief in the area of origin that reflected mix globular-lamellar structure of investigated titanium alloy. Then in the crack growth direction fatigue striations were seen on the fracture surface.
Hardened after tempering notched specimens were tested at R>0. The fractographic analyses have shown that all fracture surfaces damaged because of the interaction effect between crack edges. In fact, there was possible to investigate fracture surfaces in area of origins for many specimens. It was discovered that fatigue cracks originated subsurface in the range of durability more than $1.2x10^6$ cycles. In the crack growth direction, there were discovered fatigue striations that allowed estimating crack growth period.

DISCUSSION

The SP specimens have shown that cracks in them arise subsurface at durability more $5x10^5$ cycles independently on R-ratio. The quantity of the centres for crack origination is not to the full connected to a cyclic stress level. Nevertheless, it is necessary to specify specimens tested under stress levels 920 and 900 MPa, which is near to material yield stress, but at asymmetry of 0.33 and 0.30 respectively. Cracks origination was typical of them subsurface with the several centres.
At a high level of the maximum stress 720MPa with R=0.67 the peened specimen has broken after $1.2x10^7$ cycles. The unpeened smooth specimen at approximately the same stress level 730 MPa and R=0.64 has stood $2.3x10^7$ cycles, has not broken. Further, it has been broken cyclically completely at a higher stress level or opened under monotonically tension. Investigations have shown that the surface of the fatigued specimen has facetted pattern relief. The centre of the crack origination has been generated subsurface. The specified fact testifies that within the bifurcation area behaviour of the material defines his durability irrespective of, in what condition there is his surface. Such conclusion is in a good agreement with investigation results of SP steels and aluminium alloy [3, 6, 8, 9].
At the SP specimens crack origin occurs subsurface, and at unpeened specimens the crack arises at a surface. After reduction of R-ratio the probability of origin formation subsurface for unpeened specimens increases, and fatigue curves for the SP and WSP specimens appears identical. It serves as the proof of that fact that transition to in part closed system for a material is connected by that the condition of a surface of a specimen ceases to play a main role in behaviour of a material. The arising free surface remains inside metal, as the stress concentrator as the gas environment existing in metal, flow in area of the centre of the crack origination subsurface and creates the neutral environment as thin oxides layer.
During HT the surface of the titanium alloy sensitive to gas saturation, undergoes saturation by gas, including that went from internal volumes out-side. More damaged surface appears more sensitive to cyclic loading, than internal volumes and metal collapses faster after heat treatment, originating cracks at the surface. Material subsurface cracking out-off the origin performs under Modes I+III crack opening [4].
The received results testify that the data of fatigue tests should be submitted in quality of bimodal distributions of fatigue durability, Fig. 3.



a)



b)

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

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 behaviour 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.
The lead estimations of fatigue cracks growth period in specimens have shown that there is a general tendency of change this period in accordance with durability in process of change of stress level, Fig. 4. It is obvious, that the tendency of decrease in ratio N_p/N_f in process of increase of durability is characterized by the curve of an identical kind for all tested specimens though parameters of these curve are various for different considered situations of cyclic loading. The SP smooth specimens show that the share of the crack growth period in durability decreases in comparison with un-peened specimens. Introduction of the stress concentrator, on the contrary, increases a share of this period at identical durability. Such tendency is kept on the following hierarchy: the greatest duration of the crack origination is for notched unpeened specimens, further peened, then smooth, but unpeened and, at last, SP smoothes specimens.

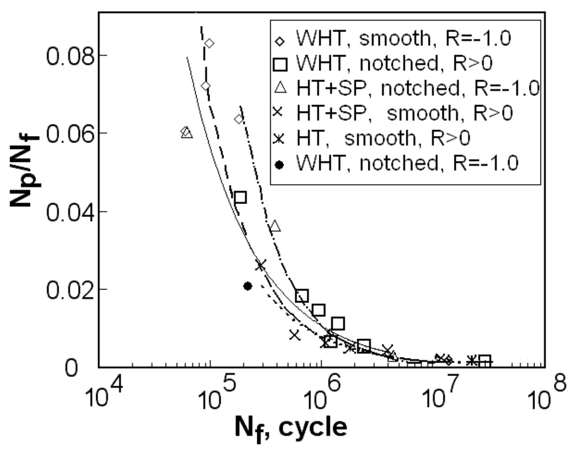


Figure 4: Ratio N_p/N_f versus durability N_f for all loading conditions of investigated specimens of titanium alloy VT3-1.

This implies that change of the mechanism of the crack origination and transition in an arrangement of its centre from a surface subsurface do not change the general law in the ratio between durability and the crack growth period. This result shows, that in various cases of the crack origin disposition - at the surface or subsurface, durability of construction elements can be estimated on the bases of the discovered dependences (see Fig.4) by the crack growth period that can be determined on the stage of the fatigue striations formation.

STRESS LEVEL FOR SUBSURFACE MATERIAL CRACKING

Let be considered for material circular cracking the model of the strip which has influence of some fictitious force, F [11]. The strip has timely the same position that a crack front has in cracked metals under cyclic loads, Fig.5. In the case of regular cyclic loads the force F produces work dW on the distance of a strip extension da which can be described by the well known relation:

$$dW_f = Fda \tag{1}$$

This work has not dependence on the crack shape but on the manner of crack extension only. The material behaviour under regular cyclic loads will be described by the relation (1) independently on the crack initiation site.

In fact, that the crack extension takes place because of metals cracking. It will be as a result of some volume, dV, failure which is equal, $dadS$, where dS can be considered as an area along the crack front being placed perpendicularly to the direction of the crack extension. Let be multiplied right site of the Eq. (1) on the dS and, at the same time, divided on this value.

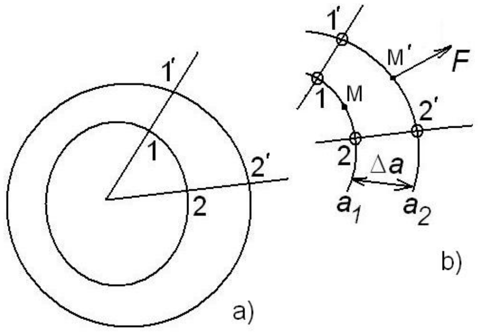


Figure 5: Scheme (a) of the subsurface crack front positions and (b) sketch of a string extension from point M to M' due to constant force F acting in the crack growth plane which reproduces the crack front fragment in two positions at the crack length a_1 and a_2 because of crack propagation.

In this case the value of FdS will be considered as "fictitious" fracture stress, σ_f, which has proportion to the tension stress equivalent σ_e for anyone external "multiparametric" cycle of cyclic loads. Therefore it can be written simply relation for density of fracture energy:

$$(dW_f / dV)_e = C_f \sigma_e \tag{2}$$

The proportion factor C_f can be considered as non dimensional parameter which shows influence of strain energy on the crack extension being $(1-\nu^2)\sigma_e / 2E$. That is why in the final form density of fracture energy in the case of uniaxial cyclic loading can be written as:

$$(dW_f / dV)_e = (dW_f / dV)_o = (1-\nu^2)/(2E\sigma_0^2) \tag{3}$$

In Eq. (1) σ_0 is the equivalent stress, corresponding to the uniaxial pulse-mode cycling; E – Modulus of elasticity; ν - Poisson factor. The stress level σ_0 may be increased or diminish, influenced by the loading conditions. It is assumed here that as concerns the macroscopic scale level, Mode-I opening of the crack is always dominant.

The density $(dW_f/dV)_e$ of fracture energy controls the crack-growth behaviour for the multiparameters applied conditions however different from the reference ones [11]. During fatigue crack propagation stress intensity factor has proportion to the density of the fracture energy and can be considered in complicated case of cyclic loading in the form of $K_e=K_I f(X_1,X_2\dots X_i)$, where $f(X_1, X_2 ... X_i)$ is the correction function for the stress intensity factor K_I [11]. That is why multiparameters of cyclic loads can be considered for the fatigue crack growth description using value the fracture energy density which can be determined based on the tension stress equivalent σ_e and known correction functions on the external loading conditions (for instance, biaxial or multiaxial cyclic loads) or on the different material states because of residual stress influence on the crack extension.

In the case of subsurface material cracking there is dominates combined Mode I+III opening of the crack in area out-off the origin. Consequently equivalent of stress intensity factor K_e for subsurface crack propagation has consideration based on the equivalent stress value σ_e with correction function $f(X_1, X_2 ... X_i)$ on the biaxial stress state. The biaxial stress-state directed to consideration in Eq. (1) equivalent stress value σ_e using correction function for stress σ_0 value in a form:

$$\sigma_e = \sigma_o [1 + f(X_1, X_2 ... X_i)]^{1/2} \tag{4}$$

In Eq. (4) $f(X_1, X_2 ... X_i)$ is correction function for the parameters of applied conditions. In the considering case of local complicated material stress-state the correction function determines of the three components, τ_{xy}, τ_{yx}, and σ_{xx} of the stress tensor, which describe the condition of material torsion under compression in-time of crack nucleation (see Fig. 1) and, then, of material torsion and tension during crack propagation [12].
For example, Chan *at al.* [13] proposed a complicated correction function for the tension and torsion cases of single-crystalline pipe specimens of a nickel alloy. The correction function takes a form

$$\sigma_e = \sigma_0[1 + C_1(\tau_{xz} / \sigma_{xx})^2 + C_2(\tau_{zx} / \sigma_{xx})^2]^{1/2} \quad (5)$$

In the Eq.(5) C_1 and C_2 are non-dimensional proportional factor which only has dependence on the Poisson's ratio.
In the discussed case of subsurface fatigue cracking of titanium alloy VT3-1 equivalent stress value σ_e has dependence on the value of FWF area. Material has not only one FWF area under different external conditions – R-ratio, biaxial stress ratios τ_{xz} / σ_{xx} or τ_{zx} / σ_{xx} and etc. That is why it ought to be summarized number of isolated origin areas.
However, in the considered case of external cyclic loading, durability of specimens with subsurface cracking has dependence on the Mode III intensiveness and stress $-\sigma_{xx}$. The discussed material stress-state can be expressed in term of hydrostatic case. In a local volume the loaded material has compression in the direction σ_{xx} during unloading part of cyclic loads and in a plane, which is perpendicular to the direction σ_{xx} takes place biaxial tension-compression as well as equivalent to torsion. This situation for material cracking from a single origin area has been considered to predict fatigue limit under torsion τ_w for steels [14]. The prediction equation for τ_w was introduced as:

$$\tau_w = [0.93 / F(b/a)](HV + 120) / (\sqrt{area})^{1/6}, \quad (6)$$

$$F(b/a) = 0.0957 + 2.11(b/a) - 2.26(b/a)^2 + 1.109(b/a)^3 - 0.196(b/a)^4, \quad (7)$$

Where $0.1<b/a<2.0$

In the Eq. (6) *HV* is Vickers hardness, kg/mm^2; *b* and *a* are axes of the elliptically-shaped FWF; "*area*" is surface area of FWF; *F(b/a)* is correction function on the crack geometry.
In the case of studied titanium alloy subjected to uniaxial cyclic tension with various R-ratio Eq. (6) have been used for estimating τ_w-value. FSF areas were measured for the crack origins. Their values ranged as 8 to 18 μm^2 at 280-kg/mm^2 Vickers hardness of the metal matrix. Ratio (b/a) has variation in the range of 1.5 – 2.0.
The calculated τ_w values ranged between 190 and 198 MPa for different subsurface crack originating at the one FWF.
In the case of crack start out-off the FWF the crack opening by the Mode I exists and the critical value of σ_w can be estimated by the following equation [14]:

$$\sigma_w = 1.43(HV + 120) / (\sqrt{area})^{1/6} \quad (8)$$

Using the same parameters for material critical state by the criterion crack opening by the Mode I it can be considered values of σ_w in the range of 470-495 MPa. The estimated value of σ_w has not contradiction with the value of τ_w.
In the case of material biaxial stress state because of torsion in local area around crack front, ratio of σ_w / τ_w can be approximately 2.4 [14]. The estimated ratio σ_w / τ_w, based on Eqs. (5) and (7), ranged between 2.37 and 2.6 for the investigated titanium alloy VT3-1.
Consequently, material critical state for subsurface cracking in different cases studied for HT+SP, SP or SMD conditions of smooth and notched specimens ought to be considered because of hydrostatic stress-state. In all conditions material has cracking by the FWF which is structural parameter determining such as equiaxed - α grain or lamellar colonies [5]. The probabilities of occurrence of any given level of heterogeneous deformation will decrease with decreasing stress level. The stress equivalent σ_e value that can be calculated by the Eq. (3) or (5) has to be used for fatigue crack growth period estimation in investigated titanium alloy based on the equivalent of the stress intensity factor $K_e = K_I f(X_1, X_2 ... X_i)$ [11].

CONCLUDING REMARKS

The performed investigation of titanium alloy VT3-1 has shown that in area of durability more than 10^6 cycles the transition to material fracture with fatigue crack origination at the surface to subsurface is realized. Influence on behaviour of a material of heat treatment, short peening, and the stress concentration resulting in stress variations for this transition, demand essentially estimation of fatigue durability for titanium alloy VT3-1 by the different S-N curves with bifurcation area. It is necessarily to introduce bimodal S-N curves for titanium alloy VT3-1 with the left branch for the fracture mechanism in connection with crack origination at the surface, and the right branch for cracks origination subsurface.

The subsurface crack initiation in titanium alloy VT3-1 takes pales in hydrostatic condition in wide range of cyclic loads parameters with different material conditions because of treatments, peening and their combinations. The weakness occurs under compression and rotation of a local volume of the cracked material by the α grain or lamellar colonies and Mode III is dominant mechanism of the fist weakened smooth facet creation for external condition of cyclically tensed material.

The critical values of τ_w and σ_w for first weakened smooth facet have been estimated based on Murakami's $\sqrt{area}$ model. The result of estimation has shown that ratio σ_w / τ_w for the first weakened smooth facet of titanium alloy VT3-1 ranged between 2.37 and 2.6. It is in accordance with the discussed earlier ratio in the case of material torsion.

Variations of material state because of heat treatments, peening and stress concentration, do not change the general tendency of the crack growth period increasing when durability increases for investigated alloy VT3-1. The ratio N_p/N_f versus durability remains unified in the both considered cases of fatigue crack origination from a surface and subsurface, if crack growth period has been estimated for stage of material cracking with fatigue striation formation.

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