Mechanisms of fatigue crack initiation subsurface and growth of the surface hardened titanium alloy Ti-6Al-3Mo-0.4Si

A.A.Shanyavskiy[1]*, T.N. Zaharova[2], Yu.A.Potapenko[1] and M.A.Artamonov[1]

[1]State Centre for Civil Aviation Flight Safety, 141426, Moscow region, Chimkinskiy State, Airport Sheremetievo-1, PO Box 54, Russia

[2]FSUP CIAM, 113235, Moscow, Aviamotornaya street, 5, Russia

*shana@flysafety.msk.ru

Keywords: Hardened Ti-alloy, High-cycle-fatigue, Bifurcation area, Very-high-cycle-fatigue, Fractography, Crack initiation, Subsurface, Facetted pattern, Crack growth, Fatigue striations.

Abstract. This paper demonstrates how to select results of fatigue tests of notched and smooth specimens tested in transition area from High- to Very-High-Cycle-Fatigue. Fatigue tests were realized on the round bar specimens of Ti-6Al-3Mo-0.4Si alloy with 8mm in diameter and some of them had circular notch with 2mm in the depth with radius 2mm. Tests were performed in the range of maximum stress level 120-750MPa at the R-ratio in the range of (-1.0) ÷ (+ 0.67). The discussed transition area for notched specimens was demonstrated in the range of (300-400) MPa based on the fractographic analysis. Both manners of the crack initiation subsurface and at the surface were seen for different specimens at the same stress level in this area. The crack initiation subsurface was discovered for hardened smooth specimens in all range of stress level (120-400) MPA tested at stress ratio R=-1.0. The crack growth period, N_p, estimated based on the fatigue striation spacing measurement, has unified dependence of N_p / N_f on the lifetime to failure N_f.

Introduction

The hardening procedure of titanium alloys that influenced on their behavior in *Low-Cycle-Fatigue* (LCF) regime has been executed with reference to titanium alloy Ti-6Al-2Sn-4Zr-2Mo-0.1Si with lamellar and globular structures [1]. It has shown that increase of durability after through hardening specimens takes place be-cause 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.

We should point out that the cracks defined as subsurface nucleate in a no strengthened material and at the stresses never above the fatigue strength σ_{w2}: at this stress level cracking never began from the specimen surface earlier than after 10^6—10^7 loading cycles [2-7]. A new (second) S—N curve, shifted towards the greater fatigue lives and showing fatigue life to further increase with de-creasing applied stress, appears characteristic of the undersurface fatigue cracking. This new area for materials fatigue life was named *Very-High-Cycle-Fatigue* (VHCF).

The mechanism of quasi-cleavage with facetted pattern formation was earlier discovered for titanium alloys VT3-1 (Ti-Mo-Si) and VT8 (Ti-Mo-Si-Cr), and it was shown that in this case the durability and fatigue crack growth period have minimum [8,9]. This situation thoroughly studied in area of LCF that has allowed to reveal sensitivity of titanium alloys VT3-1 and VT8 to dwell time of cyclic loads that is typical for in-flight loading of compressor disks [9].

Discussed below is the fatigue-fracture (crack initiation and propagation) behavior of the round surface-hardened or soft specimens of titanium alloy VT3-1 tested in regimes of LCF and *High-Cycle-Fatigue* (HCF). Researches were carrying out in cyclic loading conditions for the notch-free (smooth) or circularly notched specimens with different asymmetry down to durability of 5.10^7 cycles.

Experimental procedures

Material, specimens and the hardening procedure. 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, marked “01”, had the same material state that manufactured the disk. Another part of finally manufactured specimens, marked “02”, has been subjected to heat treatment (HT) with tempering at T = 530°C within 6 hours, and then cooling on air. Specimens without tempering procedure marked *WHT.*

The material *Ultimate* tensile stress was in the range of 1040-1100 MPa at elongation and section area reduction respectively (10.8-16) % and (3.1-4.7) %.

Two kinds of specimens of 8mm in diameter have been made: smooth and 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.

Fatigue tests and investigations procedures. Three types of cyclic loads were realized in tests, Fig.1. 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, σ_{max} was in the range of (140-920) MPa with stress ratio in the range of (0.3-0.67) for tension because of various stress amplitude σ_a and at -1.0 for tension-compression. Tests with various R-ratios were performed at the constant mean stress level of σ_m = 600MPa.

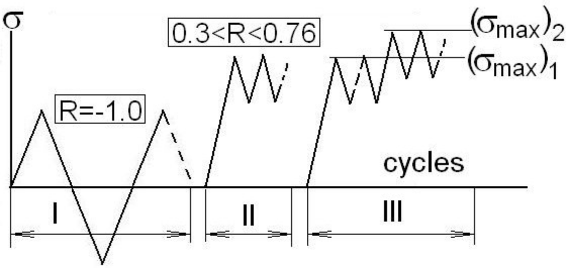


Fig.1. Schema of cyclic loads types (I-III) used in fatigue tests of specimens of the VT3-1 titanium alloy.

In irregular tests specimens were tested under the stress level in area of bifurcation up to (1.4-5.4)10^7 cycles and then stress level was increased and the specimen was failed under the cyclic loads during not more than 5.10^5 cycles.

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. With this purpose have been prepared slices from specimens in a plane, it is perpendicular their axes.

Results of investigations.

Tests data. 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 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, Fig.2. Lower fatigue-limit discovered for HT-specimens.

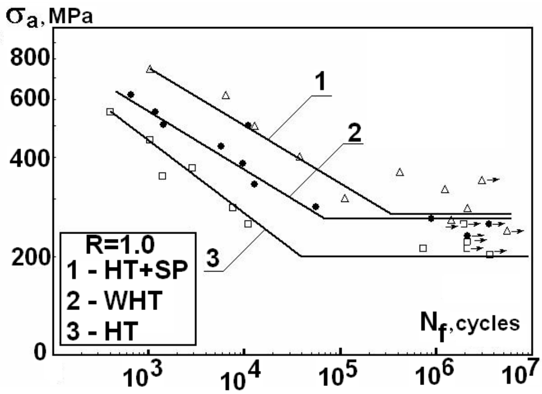


Fig.2 S-N curves for notched specimens of titanium alloy VT3-1.

Fractographic analyses. First, the smooth unpeened specimens was investigated that was tested after HT up to $2.3x10^7$ cycles at R=0.64 without fracture. Then the maximum stress level was increased from 730 MPa up to 790 MPa with R=0.52 and specimen was cyclically failed after $4.1x10^5$ cycles.

The fractographic analysis has shown, that is direct subsurface has been generated facet of quasi-cleavage, having the significant area of a surface, Fig. 3. It was performed because of destruction of lamellar structure of a material and covers three blocks of this structure. It is possible that all three blocks are located within the limit of one grain which size in a plane of material fracture makes about 100 μm. This specified facet surrounded the similar relief of quasi-cleavage of a material on significant distance from it.

The revealed features of fracture surface relief in area of the fatigue crack origin have allowed to conclude, that long time of the specimen cyclic loading under stress level 720 MPa without material fracture has led to it damage in internal volumes on removal from a surface. That is why the crack origin could be realized subsurface at continuation of tests at the stress level 720 MPa at R=0.64.

In the direction of the crack growth from the origin, fatigue striations were formed. Their spacing allowed estimating crack growth period, which has made not less than 9000 cycles.

The received result of fractographic analysis shows, that in the range of durability 10^6 - 10^7 cycles smooth specimens with different R-ratios grow probability of the fatigue crack origination subsurface. Therefore, it was important to analyze influence-hardening procedure on the material behavior, which can affect fatigue crack origin subsurface.

The considered group of peened specimens is tested up to fracture in the range of durability $2.9x10^5$ - $1.2x10^7$ cycles.

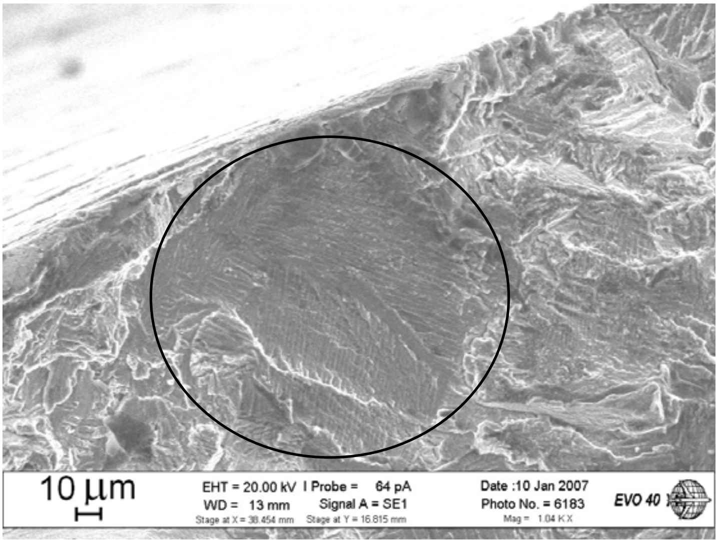


Fig. 3. Fracture surface in the area (shown circle) of origin (a), (b) at different places for two HT-specimens tested up to $2.3x10^7$ cycles at R=0.64 and 730 MPa without fracture and then fatigued up to fracture at 790 MPa with R=0.52 during $4.1x10^5$ cycles.

In all specimens, a fatigue crack origination took place subsurface, Fig.4. At the lower durability there was performed several areas of crack origins subsurface. For example, at the durability $5.7x10^5$ cycles four zones with specific facetted pattern. One of the origins has been removed from a surface on depth about 0.5mm, and the second - 0.3mm.

It is typical, that the centers 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 (see Fig.4). Focus of the crack origin has taken place on border of the greatest heterogeneity of properties of a material.

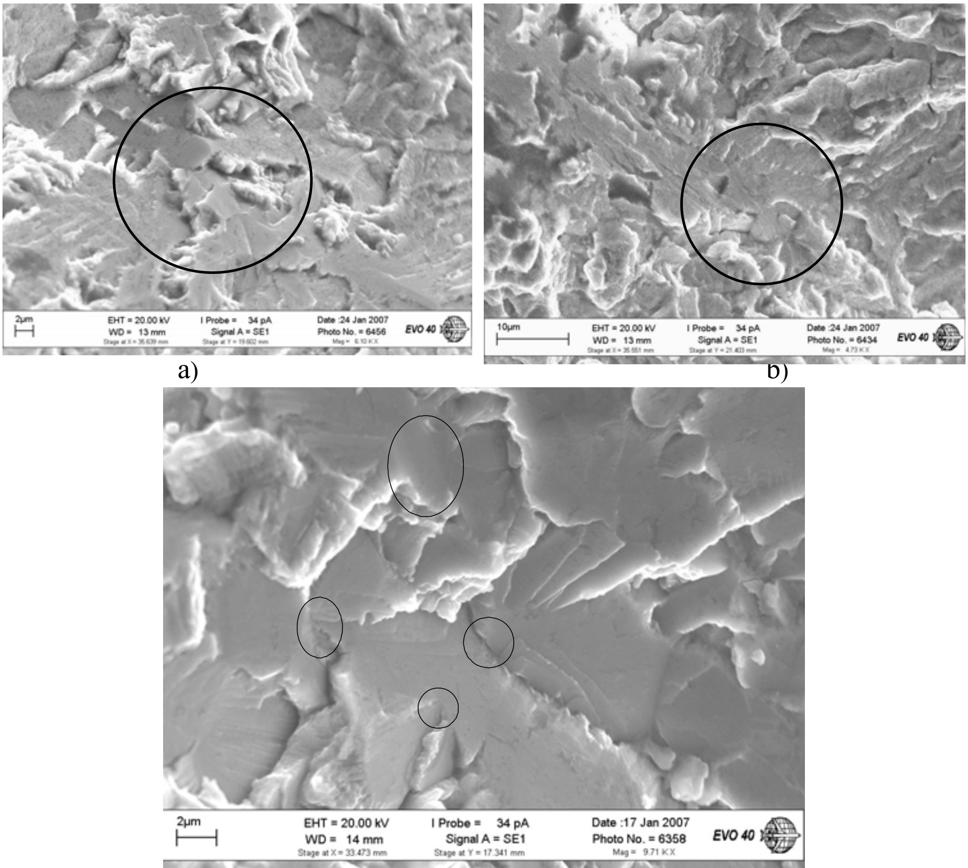
In the direction of the fatigue crack propagation, there were formed fatigue striations. Crack growth period estimations have shown the tendency of decrease in the crack growth period at the stage of fatigue striations formation with maximum stress level increasing that is quite natural is obvious. However,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

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 whose durabilities were $1.4x10^6$ and $2.4x10^6$ cycles. For both specimens, there was facetted pattern relief in the area of origins that reflected 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 shave shown that all fracture surfaces damaged because of the interaction 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.



a) b) c)

Fig.4. Fracture surfaces in area of origins for hardened smooth specimens with globular material structure fatigued at (a) $5.7x10^5$ cycle, (b) $4.0x10^6$ cycles and (c) 3.10^5 cycles. Places of subsurface crack origination at different areas shown by circles.

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 centers 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 centers.

At a high level of the maximum stress 720МПa 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. Investigations have shown that the surface of the fatigued specimen has facetted pattern relief. The center of the crack origination has been generated near to a surface (subsurface). The specified fact testifies that within the bifurcation area behavior 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 steel and aluminum alloys [3], [4], [6].

At the SP specimens crack origin occurs subsurface, and at unpeened specimens the crack arises at a surface. At reduction of R-ratio the probability of origin formation subsurface for unpeened specimens occurs. After the specified transition, fatigue curves for the peened and unpeened specimens appear 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 behavior of a material. Determining there is his structure, which realizes the mechanism quasi-cleavage of globular or lamellar structures. The arising free surface remains inside metal, as the stress concentrator as the gas environment existing in metal, flow in area of the center of the crack origination subsurface and creates the neutral environment as thin oxides layer.

Thus, influence SP at smooth specimens on their durability is caused by that the material passes in the behavior to in part closed condition when his ability to resist to cyclic loading does not depend on an environment and a condition of his surface. During heat treatment the surface of the titanic 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.

Notched specimens have shown less expressed tendency to crack origination subsurface after their hardening. At an operating time $(1.2\text{-}4.2).10^6$ cycles the center of the crack origination has been located subsurface, but it was close to it. This fact reflects influence not only a notch on features of fatigue cracks origination, but also a role of high asymmetry R=-1.0. To distinguish these two factors from the lead analysis it was not possible.

Nevertheless, despite of strong damage of specimens surface at negative asymmetry it is possible to assert, that their destruction is caused behavior of in part closed system at which at durability more than 10^6 cycles origin of fatigue cracks occurred subsurface.

The received results testify that the data of fatigue tests should be submitted in quality of bi-modal distributions of fatigue durability. 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 investigation results have shown that SP smooth specimens in the case of a high stress level and high R ratios the fatigue curve for in part closed system settles down essentially more to the right, than for notched specimens, Fig.5. 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. In addition, the R-ratio influence also takes place on the material behavior.

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 multiple parameter 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. There can be not only bimodal situation for S-N curves because of bifurcation area. It depended on the surface state because of environment influence. For titanium alloys with environment of high temperature influence there can be introduced oxygen at the surface layer [6]. In this case can be seen multi modal distribution of durability and one of the cause of this dispersion in tests results is surface state degradation because of surface layer brittleness.

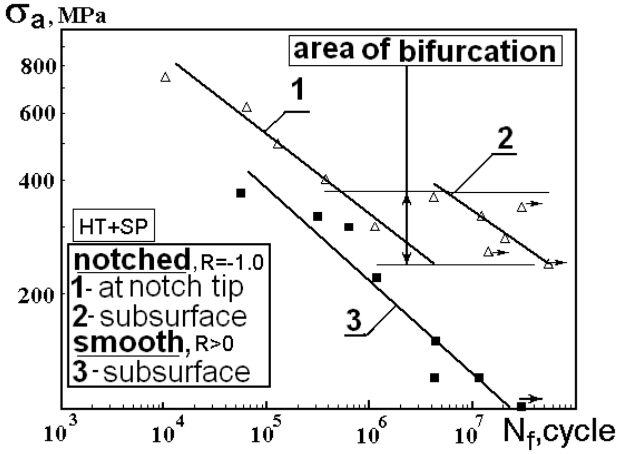


Fig.5. S-N curves for tempered and hardened (HT+SP) notched (1), (2), and (3) smooth specimens. S-N curves described bimodal distribution of the lifetime to failure (or durability) for notched specimens with crack origination (1) at the surface (at notch tip) and (2) subsurface.

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. 6. It is obvious, that the tendency of decrease in ratio in process of increase of durability is characterized by sedate function of an identical kind for all tested specimens though parameters of this function are various for different considered situations of cyclic loading.

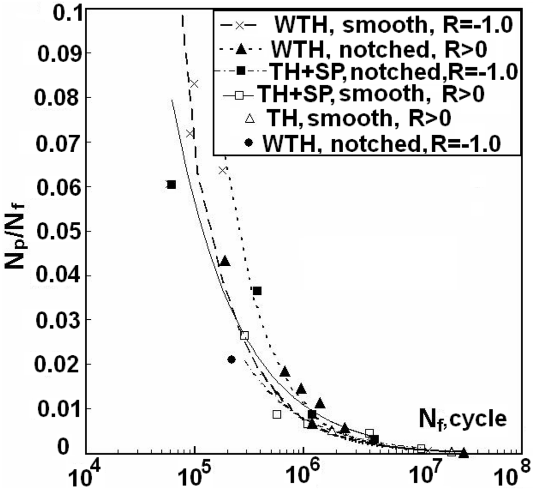


Fig.6. Ratio N_p/N_f versus durability N_f for all investigated specimens of titanium alloy VT3-1 after tests by I and II types of cyclic loads (see Fig.1.)

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. It testifies that with other things being equal the role of the stress concentrator is connected to downturn of duration of the period of the crack origination. 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. This implies that change of the mechanism of the crack origination and transition in an arrangement of its center 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.5) by the crack growth period that can be determined on the stage of the fatigue striations formation.

Summary

The performed investigation of titanium alloy VT3-1 have shown that in area of durability more than 106 cycles are realized transition to material fracture with fatigue crack origination at the surface to subsurface. Multiple parameter influence on behavior of a material of heat treatment, peening, and the stress concentration resulting in stress variations for this transition, demand essentially estimation of fatigue durability of 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. The left branch answers the fracture mechanism in connection with crack origination at the surface. The right branch answers cracks origination subsurface.

Variations of multiple parameter influence on alloy VT3-1 - heat treatments, peening and stress concentration, do not change the general tendency of change of the crack growth period at change of durability. The ratio N_p/N_f versus durability remains unified in the both considered cases of fatigue crack origination at the specimen surface and subsurface.

References

[1] A.A. Shanyavskiy: Fatigue Engng. Mater. Struct. Vol. 28 (2005), p. 195

[2] A.S. Stanzl-Tschegg and H. Mayer (Eds): Fatigue in Very High Cycle Regime (Proc. 2nd Intern. Conf. VHCF, BOKU University of Agricultural Science pub., Vienna, Austria, 2001)

[3] C. Bathias and P.C. Paris: Gigacycle fatigue in mechanical practice (Marcel Dekker, NY, USA, 2005)

[4] T. Sakai and Y. Ochi (Eds): Very High Cycle Fatigue (Proc. Third Intern Conf VHCF-3, September 16-19, 2004, Ritsumeikan University, Kusatsu, Japan, 2004)

[5] A.A. Shanyavskiy: Int. Journ. Fatigue, Vol. 28 (2006), p. 1647

[6] USA 2007

[7] A.A.Shanyavskiy: Modelling of fatigue cracking of metals. Synergetics in aviation (Ufa, Monography, Russia, 2007)

[8] A.A. Shaniavski and A.I. Losev: Fatigue Fract. Engng. Mater. Struct. Vol. 22 (1999), p. 193

[9] A.A. Shanyavskiy: Tolerance fatigue cracking of aircraft components. Synergetics in engineering applications (Ufa, Monography, Russia, 2003)