An understanding of crack growth in VHCF from an internal inclusion in high strength steel

Chong Wang [1*] **, Alexander Nikitin** [1] **,Andrey Shanyavskiy** [2] **and Claude Bathias** [1]

[1] LEME/U-Paris Ouest LEME, 50 rue de Sevres, Ville-d'Avray, 92410, France
[2] SCCAFS, Air.Sheremetevo-1, PO Box 54, Moscow reg,Chimkinskiy State,141426, Russia

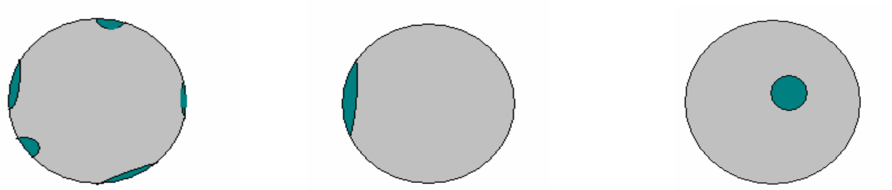
[*] wangchongscu@163.com

ABSTRACT. *Through our own results and from the literature it is observed that the micro plasticity in the gigacycle fatigue induced more than dislocations sliding. Sometime, phase transformation, refining of the grain, twinning, and instability of the yield point, occur even when the loading is small during a very high number of cycles. In high carbon content and in high strength steel, there is a transformation of the microstructure starting from a defect, in relation with the stress concentration and the stress field.*
Keywords*: Steels, very-high-cycle-fatigue, synergetic, subsurface crack initiation, ultra-high-plasticity, twisting mechanism*

INTRODUCTION

At the microscopic level, Mughrabi [1] shows that the initiation of fatigue crack in the gigacycle fatigue regime can be described in terms of a micro structurally irreversible portion of the cumulative cycle strain. It means that there is no basic difference between fatigue mechanisms in low, mega, and giga-cycle fatigue except for the strain localization. However, specific mechanisms can occur depending on the fatigue life. The fatigue life seems to be a key parameter to determine correctly the fatigue initiation location. In low cycle fatigue, in megacycle fatigue and in gigacycle fatigue different mechanisms can operate at different scales of plasticity. In the low cycle fatigue regime, the cyclic plastic deformation is critical at the surface but exists also in the bulk of the metal. Typically, several cracks nucleate from the surface. When the fatigue life is below 10^5 cycles, general plastic deformation of the specimen bulk governs the initiation. When the fatigue life is between 10^6 and 10^7 cycles the plastic deformation depends on the plane stress surface effect and the presence of flaws which explains the critical location of fatigue initiation. Typically, the initiation starts with one crack only, from the surface. However approaching

10^9 cycles the plastic deformation in plane stress conditions is vanishing; the macroscopic behavior of the metal is elastic except around flaws, metallurgical defects or inclusions. In very high cycle fatigue, the plane stress conditions are not enough for a surface plastic deformation according the Von Mises criteria. (Fig 1) The initiation may be located in an internal zone.



$N_i=10^4$ cycles $N_i=10^6$ cycles $N_i=10^9$ cycles

Figure 1. Location of the fatigue crack initiation

In gigacycle fatigue ,when the crack initiation site is in the interior, this leads to the formation of a “fish-eye” on the fracture surface, and the origin of the fatigue crack is an inclusion, a “super grain” (micro structural homogeneity), or porosity. At the macroscopic scale, under the optical microscope (or naked eye), the fish-eye area looks white, whereas the region outside of the fish-eye looks grey. In almost all cases, this fish-eye appears circular, with a dark area in the centre, inside which the crack initiation site is located. Controversies exist on the origin of this dark area, which some authors have named: “Optically Dark Area (ODA)” [2] “Fine Granular Area (FGA)” [3] “Granular Bright Facet (GBF) [4]. But the problem has not been resolved yet. It is not clear what kind of processes is dominant. Murakami model of crack origination is based on the knowledge about hydrogen influence on the crack initiation. It exists in internal inclusions which trapped hydrogen during manufacturing procedure. But in the case of supergrain which is the first step of subsurface cracking there is no hydrogen in the volume around [5].

Another question is about FGA around inclusion. Following by Sakai this area occurs because of process of polygonisation, but following by Murakami this area occurs because of hydrogen influence. Nevertheless grey color of the surface in the center of the crack origination under optical microscope cannot be explained only based on the knowledge about hydrogen influence [2]. Grey or black color can be result of fracture surface formation with different roughness but, at the same time, it can be difference in manner of chemical composition diffusion during material cracking. For example, the mechanism of carbon diffusion in area of metals cracking has been introduced in the case of steel pressing during rolling contact [6]. But it is not clear could be this effect considered for another cases of steels fatigue cracking in VHCF regime or not.

For example, new effect has been considered for steel SAE 52100 in UHCF regime [5]. It was discovered that in FGA carbon is dominant being more in percent than in material composition. This effect has been explained because of carbides formation during FGA subsurface occurring. But this explanation has contradiction with mechanism of carbon diffusion as a result of carbides destruction [6].

At last, Murakami mechanism of ODA formation based on the hydrogen effect but mechanism of carbides destruction considered hydrogen as such chemical element which accelerates material cracking but it not plays dominant role in the subsurface fracture mechanism [6].

Nevertheless, but the FGA has been seen not only in steels. It has occurring in Al-Cu based alloy [8] and in compacted superalloy EP 741 NP [9]. The same pattern but is not strongly expressed in the center of the subsurface cracking was discovered in Ti-based alloys [9], and it can be seen during first stage of metals cracking at the surface but in vacuum [10, 11].

That is why the question grows up: if the same manner of metals subsurface cracking takes place for different metals may be it is fundamental process that has to be considered using unified physical approach. This paper tried to introducing possible steps in this direction.

TEST PROCEDURE

To achieve the very high number of cycles as much as 10^9, piezoelectric ultrasonic fatigue system was used with advantages of time saving and lowers cost. Different high strength steels were studied, martensitic or bainite in UTS range 1400~2000 MPa. Ultrasonic fatigue testing under cycle loading in tension-compression, R=-1.

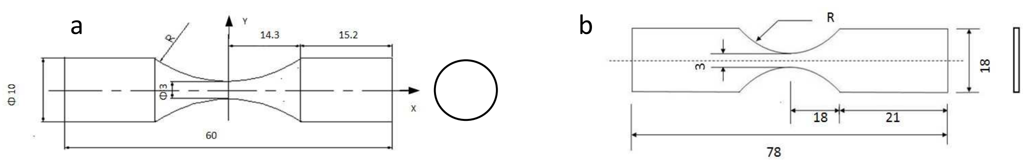


Figure 2. Specimens for ultrasonic fatigue (Unit: mm)

Generally the ultrasonic tests are performed with round specimens. For a mechanical viewpoint, it is better to use cylindrical specimens avoiding edge effect, Fig 2a. However for microscopic observation, a flat specimen is more convenient. It must be pointed out that a plane stress field is emphasized in a flat specimen of one millimeter thickness or less. On the contrary, the plane stress effect is limited to the surface in a round specimen. A new designed flat specimen (1 mm thickness) as shown in Fig 2b is used for ultrasonic frequency fatigue testing, Specimen, special attachment and piezoelectric fatigue machine constituted the resonance system working at 20 KHz. Before fatigue testing, surface of specimen polished to the roughness level R0.2. During test specimen maintain around the ambient temperature by a cooling system.

RESULTS OF INVESTIGATION

Facture surface observation on optical microscope shows same characteristic in both flat sample and cylinder specimen at very high cycle fatigue regime in agreement with the Paris-Bathias model. Fig 3

- A dark area zone (so-called ODA) due to the initiation mechanisms. [12]
- A penny-shaped zone (short crack growth).Whatever the crack initiation site (spherical or elongated inclusion, super-grain, pore), the fracture surface becomes circular around the initiation site.
- A zone with small radial ridges corresponding to the short to long crack transition
- A zone with large radial ridges (long crack growth). In this zone, the fatigue crack propagation produces striations for which the mean distance between striations is a function of ΔK^2, in good agreement with the CTOD.[5]

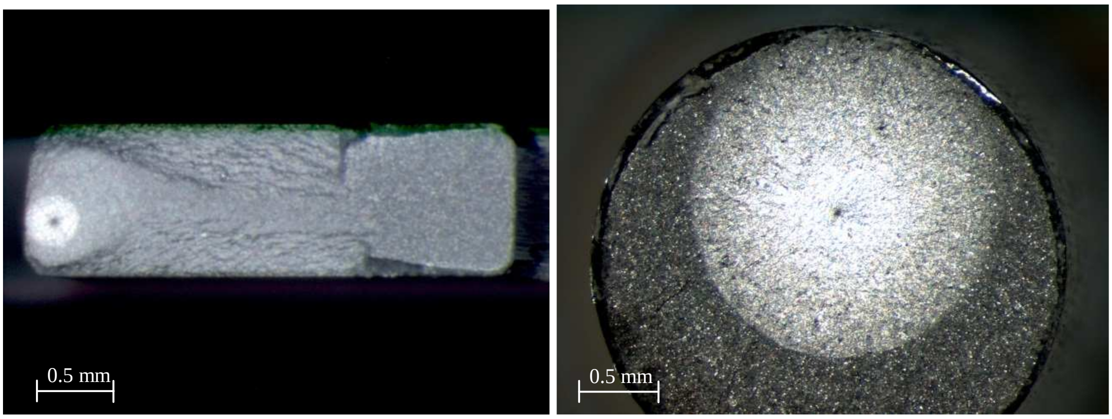


Figure 3. Optical microscope observation on specimens

Fracture surface analyses on scanning electron microscope have shown three types of subsurface fatigue crack initiation.

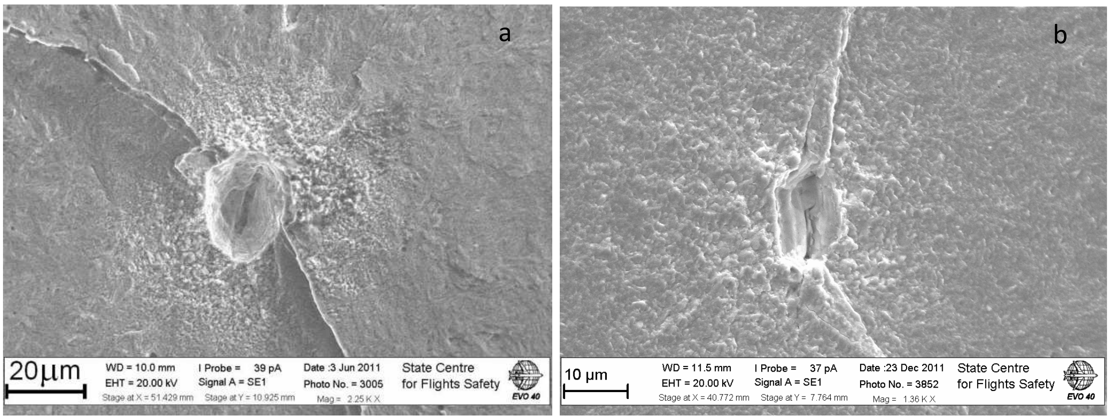


Figure 4. Well-known first type of fatigue crack initiation from inclusion.

In the first case it was representative internal initiation in VHCF which usually occurred from metals inclusion, Fig.4. There were seen the same fracture surface patterns that usually discussed in all papers (see, for instance, Murakami [2]). It is clear that difference in inclusion shape determined difference in fatigue crack initiation site (see Fig.4b).

There is evidence in difference between two fracture surface patterns in area of crack origination from inclusion: (1) two semi-elliptically-shaped areas of FGA; (2) areas of quasi-brittle cracking from the inclusion between FGAs. FGAs and areas between them placed in different plains.

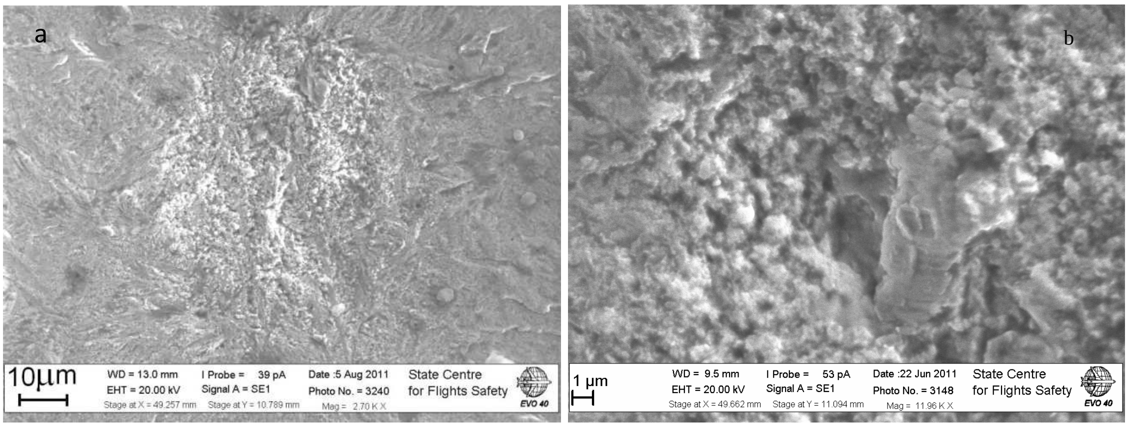


Figure 5. Area of crack origination in the "second" case without inclusion

Second case of subsurface cracking, in the centre of crack origination exist only FGA without inclusion, Fig.5. In some area of FGA a pattern can note which may discussed as brittle fracture of material structure (see Fig.5b). It was compared with brittle fracture of this material in area of fast opening fracture. Comparing has shown that FGA has occurring, may be, around one of the material grains which cracking can be before or after FGA formation. But there is no clear evidence in fracture surface patterns that have significance about influence material structure on the crack origination manner.

Third case of subsurface crack origination is more complicated, Fig.6. There is supergrain that has been broken first, and several FGAs and areas of quasi-brittle cracking have been formed around.

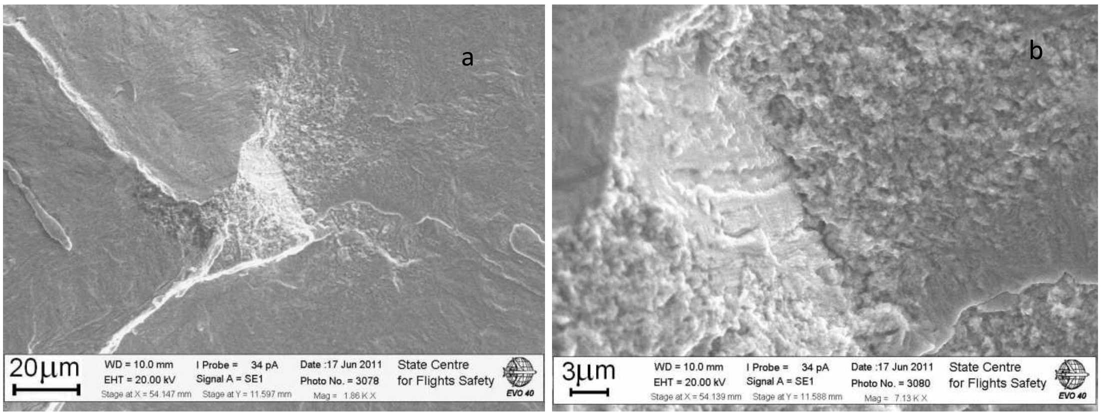


Figure 6. Area of crack origination from the supergrain

Facture surface patterns detailing consideration in area of supergrain cracking has shown that the crack start took place in small area around the corner were cracking for the specimen had origination (see Fig.6b). It is as first-smooth-facet (FSF) out of which took place material cracking in the direction through the supergrain with meso-beach-marks formation because of crack retardation under external cyclic loads. Also, there have regular patterns which looks like as fatigue striations in the direction of crack propagation through super-grain. But they are not fatigue striations. They have occurring because of structural elements cracking such as martensitic colonies.

Discussed three cases of subsurface fractures surface patterns occurring reflect systematic behavior in metals cracking: (1) always FGA formation; (2) existing FGA only or with areas of quasi-brittle fatigue cracking; (3) in the centre of FGA exist inclusion or FSF formation because of structural elements cracking.

DISCUSSION

Take the specimen present in the Fig4b in to consider. Fatigue life of specimen is 2.339 x 10^8 cycles under loading $\Delta\sigma$=1340 MPa. Fatigue intensity factor ΔK in the edge of FGA then is 7.68 MPa·m$^{-1/2}$. According the assumption of effective fatigue intensity factor [13]

$$\Delta K_{eff} = K_{max} - 2/\pi \cdot K_{open} - \sigma_{nom} \cdot \sqrt{\pi d / 2}$$

Where, Kopen=$0.5 \cdot E \cdot h \cdot \sqrt{2\pi / d}$. h is distance between two patterns of fracture surface, it is negligible when crack size is very small. $K_{max}=\Delta K/2$ and $\sigma_{nom}=0$ for stress rate R=-1. Therefore $\Delta K_{eff} = K_{max}$ = 3.84 MPa·m$^{-1/2}$. Approximate, magnitude of Burges vector b in steel is in same scale of crystal lattice or double size of atom radius of steel. That means, b=2R≈2.65 Å (1 Å= 10^{-10} m). Young's modulus of steel is E≈210 GPa. Therefore E√b=3.42 MPa·m$^{-1/2}$. It is suggesting $\Delta K_{eff} \approx E\sqrt{b}$ which well agreement with Paris-Hertzberg law. While ΔK_{eff} at edge of FGA can be regard as the threshold of fatigue intensity factor ΔK_{th}. Beyond this edge of FGA, mechanism of fatigue fracture transform from initiation to the propagate stage. According to the previous co-work of Bathias and Paris [5] that fatigue life after threshold just very small part of total life in the case of VHCF, mostly less than 10%. Therefore inside of FGA or below the threshold in another word, crack taken more than 10^8 cycles to forming the FGA which has radius about 25 microns in this example. That means average crack propagate speed inside FGA much less than magnitude of Burges victor. Therefore it is inappropriate to explain the mechanism here by fatigue crack propagation. A distinguished mechanism must use to demonstrate the formation of crack initiation.

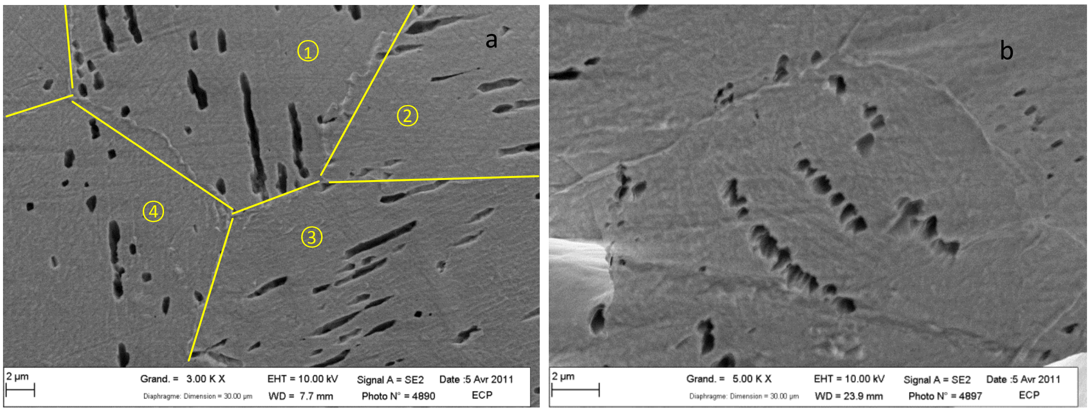


Figure 7. PSB structure in ferrite grains (a) subsurface (b) cross section

From view of simple microstructure, dislocation slip characters in the very high cycles fatigue was well studied in fcc copper crystal and recently also studied in bcc ferrite crystal. Both two of basic lattice of crystal can have persistent slip bands (PSBs) during certain stress in very high cycles loading. Fig. 7 shows the PSB structure in ferrite grains on the subsurface which parallel to loading direction and on the cross section which perpendicular to loading direction. Links of PSBs was regarded as the dominative mechanism of fatigue crack initiation in VHCF regime on single phase copper and ferrite material. Could dislocation slip also be the fracture mechanism on high strength steel after very high cycle loading?

Let’s start from the typical initiation of internal crack which has inclusion in the center of FGA. In this case around inclusion can be realized different sequence of events in dependence on the cyclic loading shape, Fig. 8.

During uploading portion of cyclic loads (between points “0”-“1”) matrix and inclusion experienced in areas “A” and “B” different influence of stressing. In the point “A” material has tension and destruction by the border between inclusion and matrix grows up. In point “B”, horizontal plane, occurs matrix pressing from the inclusion because during uploading portion of cyclic loads matrix deformation directed to deformation in the inclusion direction. Maximum stressing of destruction in area “A” and maximum pressing for matrix at area “B” takes place at the point “1” of external loading. During unloading portion of cyclic loads at the point “2” there exist additional influenced from residual stresses always exist in metals. When the cyclic loads goes to the last half of cycle, between points “2”-“3’ and “3”-“4” it will has the same effects for matrix stressing but in opposite.

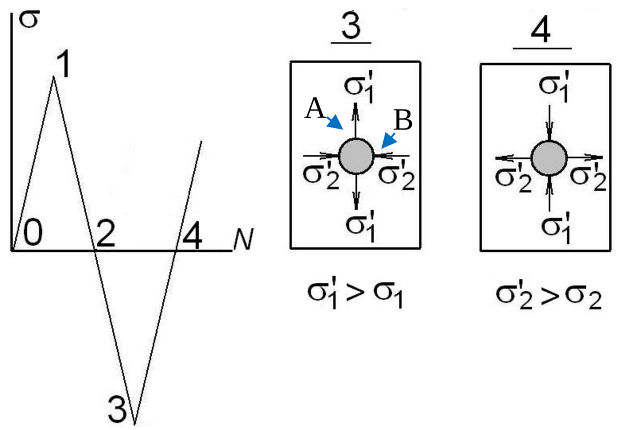


Figure 8. Matrix stress-state around inclusion schematized of external cyclic loading

Furthermore, during discussed cyclic loads above, there will be inclusion rotation because of not symmetrical case of its shape and asymmetry in acting loads by the surface of inclusion (see, Fig. 4). More or less but rotation components will be always exist under external cyclic loading. That is why it should be considered stress-state around inclusion as triaxial because of residual stresses and component of rotation for matrix in area of crack origination by the border between matrix and inclusion. Inclusion rotation has been also discussed in the paper [14].

Earlier the stress-state of metals in local place in the case of external tension has been estimated using parameter $p=\sigma^*/\sigma_{ext}$ [15]. For example, value of "p" has been demonstrated in the range of 80-290 for Zn and Al. The same effect has been discussed based on elastic energy calculation that needs matrix to be broken during interaction process between acting dislocations [16].

Therefore, at the considered points "A" and "B" there is growing up energy accommodation process under cyclic loads influencing material state and directed to appear cracking by the border of one inclusion which has the highest stress-state.

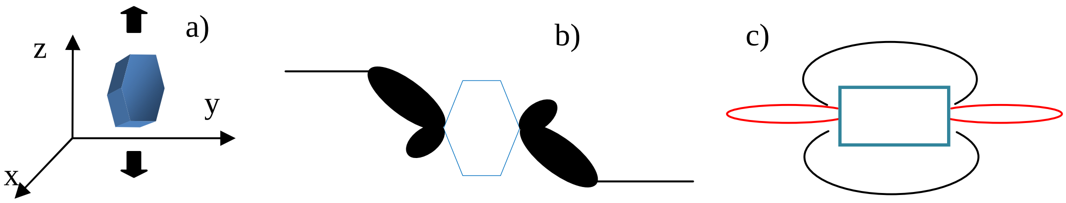


Figure 9. Sketch of inclusion initiation.

Sketch in Fig. 9a was used to illuminate the mechanism of internal crack initiation in the high strength steel. During the fatigue loading, because of local stress-state discussed above, some point of matrix around inclusion has the stress reach to critical shear stress of dislocation slip. Followed by plastic zone occurred in the bulk of steel. Orientation of plastic zone has an angle with loading axis (fig.9b). Inside of this plastic zone, dislocation slip laid to deforming in a very small scale. Even dislocation slipping is very difficult in the high strength steel, but due to the concentration of fatigue stress and very high number of cycle some dislocation has irreversible slipping. Because of the microstructure of high strength steel, those irreversible dislocation forms a slip marks in the low length-width ratio instant of forming PSBs. When the crack open have two pattern surfaces though those dislocation slip marks at plastic zone, fracture surface has formed FGAs.

In reasons that geometry of inclusion and plastic zone, cracks opened in two FGAs always at different planes. Along with the progress of FGAs formation, axial cyclic loading and size of crack start to play a major role compare to the stress concentration at border around inclusion. Unique crack intended to growing perpendicular to the loading direction. Therefore matrix between two FGA planes was under shear stress. During formation of FGA, bulk of matrix between two FGAs has additional stress in cyclic compression condition besides mean fatigue loading (narrow volume in red at Fig. 9c). Finally, two FGAs at plastic zones coalesce induce a stage formation. Mostly because of higher distortional strain energy at boundary of metastructure, bigger inclusion at boundary has priority to initiate the fatigue crack accompanied with stage formation along the boundary (see Fig 10). Considering loading history of this wing-like stage formation, it is reasonable to related wing structure to the butterfly structure or white etching area (WEA) in the fracture analysis of rolling contact fatigue testing on bearing steel.

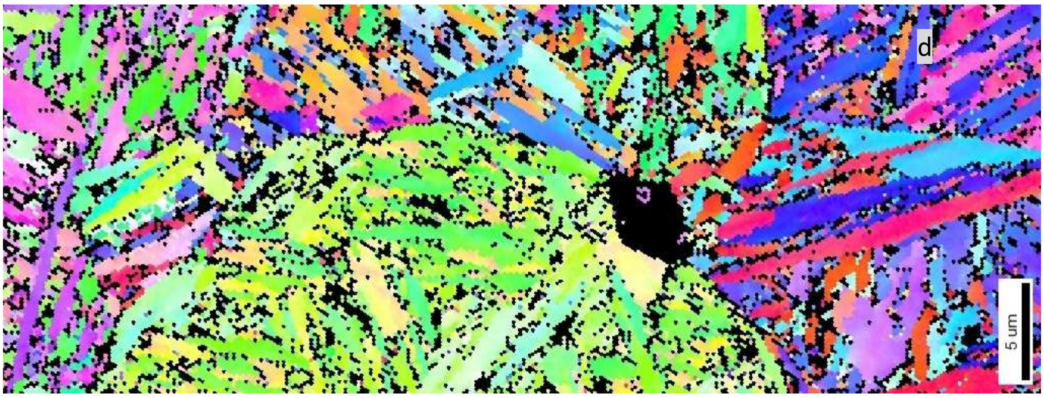


Figure 10. EBSD image of inclusion initiation

In brief, FGAs formed in the mechanism of irreversible dislocation slip marks though the plastic zone caused by fatigue stress concentration around inclusion after many cycles loading. It is should be point out that when origination of crack was inside of metastructure or FGAs were in the same plate in matrix, the wing-like structure in the crack initiation stage well not appears. Therefore metallurgical microstructure misfit or anisotropy or weakness of metals also possibly may contribute the localized stress field to driving dislocation slip when it reach to the critical shear stress of dislocation slip and eventually initiate the fatigue crack.

Nevertheless, there exists another manner of metals cracking with FSF occurring (cases “Second” and “Third”). One of the structural elements experienced weakness which directed to the next step of metals cracking. If this FSF very small, matrix loss plastic stability and form such FGA around FSF as circular area.

In case that FSF occurring is the first step of super-grain cracking, there is the same situation around that have been discussed above applicably to the case one with crack initiation from the inclusion (see Fig.5). Several areas around experienced plastic instability but other areas have crack initiation by the sliding mechanism.

Mechanism of the FSF formation was discussed applicably to Ti-, Al-, and Ni-based alloys [9, 17 and 18]. Subsurface metal volumes subjected to hydrostatic stress-state and by one of the crystallographic planes appeared plastic deformation process as whirls moving, Fig.11. Diffusion of rest gases, for example atoms of oxygen, hydrogen and nitrogen, always exist in metals or other chemical elements influenced material weakness and directed to FSF free surface occurring. When critical weakness has exceeded around FSF then there is appeared process of plastic zone formation. This zone introduced tension of matrix around FSF and crack origination occurred.

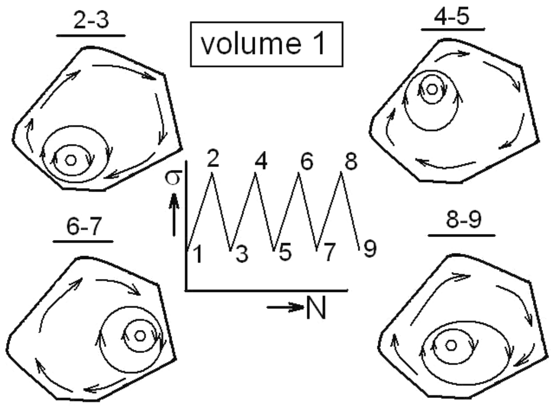


Figure 11. Schema of sequence of events for twisting material volume which occurs in one of the structural element cyclic

CONCLUSIONS

Based on our own results and from literature it were shown that

1). Subsurface crack initiation process in gigacycle fatigue could be realized by several mechanisms. Observation of a material structure shows, that dislocation processes influence on damage accumulation in gigacycle fatigue in case of different mechanisms of crack origination.

2). One of mechanisms is crack origination from the first smooth facet, which could be formed due to micro-rotation and pressing in bulk of material.

3). In case of internal inclusion, redistribution of stress field around it, led to forming nanostructure (spherical particles) in matrix of material. Fatigue crack opening through borders of these particles with forming FGA.

4). Sometimes, forming of nanostructures could be realized in case of material without inclusion. Therefore, the process of nanostructure forming could be a general feature of material to relax a cyclic loading.

5). In case of material with high disorientation of neighbor grains, a fracture could be realized in one of the grain with an unfavorable grain orientation against to axis of loading. Further damage is determined by the crystallographic orientation of neighbor grains.

REFERENCES

1. Mughrabi H. Specific features and mechanisms of fatigue in the ultrahigh-cycle regime, Int. Jl of fatigue 28 (2006) 1501-1508

2. Murakami Yu. Metals Fatigue: Effects of Small Defects and Nonmetallic inclusions. Elsevier Ltd, London, UK, 2002.
3. Sakai T., Li W., Lian B., Oguma N. Review and new analysis of fatigue crack initiation mechanisms of interior inclusion induced fracture of high strength steels in very high cycle regime. In: *Proc. of the Int. Conf. VHCF5*, Berlin, 19-26.
4. Shiozawa K., and Lu L. Very high-cycle fatigue behavior of shot-peened high-carbon chro-mium bearing steels. *Fatigue Fract. Engng. Mater. Struct*., 25, (2002), pp.813-22
5. Huang Z., Wagner D., Bathias C., Paris P.C. Subsurface crack initiation and propagation mechanisms in gigacycle fatigue. Acta materialia, 58, 2010. pp.6046-54.
6. Gegner J., Nierlich. Sequence of microstructural changes during rolling contact fatigue and the influence of hydrogen. In: Proc. of the Int. Conf. VHCF5, 2011, Berlin, pp.557-62
7. Karsch T., Bomas H., Zoch H.-W. Influence of hydrogen and microstructure on the fatigue behavior of steel SAE 52100 in the VHCF regime. In: Proc. of the Int. Conf. VHCF5, 2011, Berlin, 201-6
8. Carboni M., Annoni M., Ferraris M. Analyses of premature failure of some aluminum alloy sonotrodes for ultrasonic welding. In: Proc. of the Int. Conf. VHCF5, 2011, Berlin, pp.589-94
9. Shanyavskiy A.A. Fatigue cracking of smooth and notched specimens of compacted superalloy EP741 NP in high- and very-high-cycle-fatigue regime. In: Proc. of the Int. Conf. VHCF5, 2011, Berlin, pp.107-12
10. Stanzl-Tschegg S., Schönbauer B. Near-threshold fatigue crack propagation and internal cracks in steel. 10th Intern Fatigue Congress, June 6-11, 2010, Czech Prague, Procedia Engineering 2 (2010), 1547–55
11. Nakamura T., Oguma H. Influence of environment on the formation unique morphology on fracture surface in subsurface fractures. In: Proc. of the Int. Conf. VHCF5, 2011, Berlin, pp.257-63
12. Murakami Y. The Mechanisms of Fatigue Failure in the Ultra long Life Regime. *Metal Fatigue*. Elsevier. Oxford. 2002. UK
13. Paris P.C. Tada H. Donald J.K., Service load fatigue damage – a historical perspective,International journal of Fatigue,21,1999 S35-S46
14. Alfredsson B., Watz D., Olsson E. Fatigue initiation at inclusions and subsequent crack growth in a bainitie high strength roller bearing steel at uni-axial experiments. In: Proc. of Intern. Symposium on Fatigue Design & Material Defects, Trondheim 2011, p.13
15. Ivanova V.S., Terentiev V.F. Nature of metals fatigue. Metallurgy, Moscow, Russia, 1975
16. Oding I.A. Izvestiya “AS USSR. Metallurgy and fuel”, 3, 1960, 3-10.
17. Shanyavskiy A.A., Banov M.G. The twisting mechanism of subsurface fatigue cracking in Ti–6Al–2Sn–4Zr–2Mo–0.1Si alloy. Engng Fracture mechanics 77 (2010), 1896-906.
18. Shanyavskiy A.A. Mechanisms of the 2024-T351 Al-Alloy Fatigue Cracking in Bifurcation Area after Laser Shocks Hardening Procedure. Key Engineering Materials Vol. 465 (2011), 511-4.