FATIGUE OF HIGH STRENGTH MATERIALS: SOME CRITICAL PROBLEMS OF FRACTURE MECHANICS

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On alloyed steels of spring, instrumental and bearing duty the features of the fatigue fracture process of high strength materials were investigated. It was shown that for such materials instead of the traditional consequence of the fatiue fracture process (crack initiation-subcritical crack growth — final quick rupture) practically full resource is concentrated in the material damaging in the site of the critical crack generation. Cyclic damaging in the critical area of the specimen accumulates gradually, exhausting practically full lifetime, forming at last nucleation site of the critical (griffith type) crack.

The paper regards the dependence of the critical damaging on the structure and treatment of materials, their influence on the statistical distribution of the fatigue curves, their relation on the purity of materials and metallurgical features of the metals preparation. Such approaches of the critical damage concept are also extended on the composites and ceramics. The fatigue damaging and rapture of such materials should be regarded from the point new area of fracture approach — the physics and mechanics of the critical damaging of intensively strengthened materials which are congested with thick network of structural and substructural barriers. This area of knowledge has for the present only descriptive appearance and needs in working out new critical theory of deterioration and damaging in site of critical defect for which can not be used contemporary instrumentation of fracture mechanics that operates only with the stages of macrocrack development.

INTRODUCTION

According to the steady and generally approved insight the fatigue fracture as time distributed phenomenon can be clearly divided in two stages: the stage of crack initiation and the stage of crack propagation. The essence of the problem which is well developed for low strength metals and alloys [1, 2] is represented on Fig. 1.

As it concerns pure unstrengthened metals fatigue fracture intiates in persistand slip bands where the original tangentially oriented microcracks are formed. The concentration of such microcracks in statistically unhappily places leads to formation of short cracks that gradually transform in the normally

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oriented mackrocrack. So begins the second stage of fracture, whis is connected with the subcritical fatigue crack growth.

Unlike to the knowledge about the crack initiation phenomena that has qualitative descriptive character the methods of fracture mechanics based on stress intensity factors (SIF) form the ground for quantitive estimations of the mechanical durability of machine parts and constructions on the basis of fatigue crack growth diagrams in coordinates: *V* (crack growth velocity) — *D K* (amplitude value of SIF). And today fatigue fracture mechanics represents the great number of explorations based on studies of *V* — *D K* curves and making physical sense of their parameters for great variety of structural materials [3, 4]. There are also frequently encountered various scientific speculations for use *V* — SIF data as universal information about fatigue resistance of materials in any conditions including fatigue behavior of smooth specimens and machine parts and determination of their fatigue limit [5].

In spite of that the problem of fatigue fracture in particular the question of crack nucleation essentially complicates for high strength materials, hardened by martensitic transformation or various methods of dispersion strengthening, similiary as for new composite materials and ceramics. For such materials do not exist precise and moreover quantitive notions concerning crack initiation on second phase particles, crack formation in sites of grain boudary intersection or others. Taking into account pronounced brittleness of high strength materials for them it seems very applicable linear fracture mechanics method on the stage of subcritical crack growth. However the fatigue fracture of high strength materials has a series of peculiarities that withdraw the fracture problem from the area of above mentioned conventional schemes. This paper is devoted to the evoluation of specific fatigue fracture behaviour of one of the mentioned classes of high strength materials — quenched alloyed high carbon steels.

EXPERIMENTAL PROCEDURE AND MATERIALS

The investigations were performed on alloyed high carbon steels after quenching and final tempering (100–550 °C). The mentioned steels (Table 1) belong to the high strength materials of instrumental appointment or bearing and spring duty. The materials after termal treatment were tested on plain bending of precracked specimens for evaluation K_{IC} and in separate cases for drawing up *V–D K* curves.

The fatigue tests were performed on cylindrical specimens (diameter — 6,5 mm). In special cases the fatigue tests on the statistically arranged program were performed (the testing base 10^8 cycles), for individual Wöhler curve 150 specimens were used. The special procedure for estimation fatigue crack initiation period on cylindrical specimens with round oriented structural concentrator was applied.

For evolution of the fatigue demaging the acoustic emission method was used and microfractographical analysis by SEM of fracture surfaces was generally utilized.

Table 1 Chemical content (wt %) of materials

Material	C	Si	Mn	P	S	Cr	Ni	Mo	V	T
45ChN2MF	0,46	0,28	0,50	0,021	0,032	1,00	2,20	0,45	0,35	—
60ChS	0,63	1,02	0,35	0,015	0,024	1,02	0,10	—	—	—
7Ch2	0,69	0,25	0,30	0,030	0,040	1,90	—	—	—	—
7Ch3	0,72	0,35	0,31	0,025	0,035	2,80	—	—	—	—
30ChGST	0,32	1,20	1,02	0,025	0,026	1,03	—	—	—	0,45
90ChGST	0,91	1,12	0,98	0,032	0,029	1,10	—	—	—	0,62
ShCh15	1,02	0,24	0,21	0,011	0,013	1,20	—	—	—	—

MECHANICAL FATIGUE PECULIARITIES OF HIGH STRENGTH MATERIALS

The high strength steels of mentioned kind are distinguished with some special features of fatigue behaviour that distinct them from usually used and tested structural alloys.

1. The wide spectrum of fatigue tests performed on alloyed steels with high carbon content after quenching and low tempering testify that on such materials can be achieved the highest levels of high cycle fatigue strength on smooth specimens during tests at rotating bending (Fig. 2). Supplementary thermomechanical treatment of such steels provides the record levels of fatigue limits, which was shown by Zackay [6] on H-11 steel (the analogue of 45ChN2MF steel) and in our investigations for torsional suspenders operating under pulsating torsion (Fig. 3)[7].

2. The high strength instrumental and bearing steels after quenching or thermomechanical strengthening essentialy differ from widespreaded structural alloys in features of subcritical crack rate (V) — SIF ($D\ K$) curves. Such curves lose typical three stage form, their sharp dependence testify the diminishing of the $D\ K$ range in which is possible stable subcritical crack growth and the raprochement K_{th} and K_{fc} parametrs (Fig. 4). The supplementary distinction consist in possible lowering of K_{fc} in comparison with K_{IC} determined in conditions of static tests [4].

3. By special method of detection of crack appearance, using control of the general deflection of the cylindrical specimens with circle notch under rotating bending it was showed that the period of propagation of the fixed by indicator method macrocrack N_p is essentially smaller than the period of initiation of such crack N_I (Fig. 5). The relation $\alpha = N_I/N$ usually exeeds 0,8. The α value growths when the loading drops, nearing to the conditions of high cycle fatigue and to the fatigue limit of the material. Oftly determination of the ready crack is very complicated and persuade that the crack growth period is very small and α nearest to 1,0 [8, 9].

4. The use of acustical emission for evaluation fracture features and crack propagation processes on smooth specimens of high strength steels fixed crack demaging only on the final stage of the lifetime 0,90–0,95 N [10].

5. The features of the fatigue fracture zone on the fracture surface of mentioned high strength steels principally distinguish from the ductile alloys of the low and middle strength. For the latters is typical crack formation in slip bands by means of extrusions and intrusions [3] with following development of

main crack, which forms and propagate an essential part of living resource. In investigated high strength steels the nucleation site has the features of “fish eye” around one as a rule dominate inclusion or second phase precipitation (especially in high purity alloys with respect to wrong impurities). So the fatigue spot is formed behind which the rest of fracture corresponds to final spontaneous rapture (Fig.6) [11].

BRITTLE AND DUCTILE MODE OF FATIGUE FRACTURE OF STRUCTURAL STEELS

It is well known that mechanical behaviour of alloys from the point of view of the degree of brittleness of fracture depends on the structure of material and also mode and temperature of testing [4]. Very evident proof of brittle — ductile transition (BDT) depending on materials structure and testing temperature can be received in torsional quasistatic test of high strength steels on smooth specimens. As it is shown on Fig. 7 and 8 for instrumental 90ChGST steel the change of structure (varying tempering temperature) or testing temperature (on low tempered steel) lead to the changes of strength τ_B and plasticity Θ_B that can be fixed as the BDT from the brittle to shear fracture. Such transition comes in tempering or testing temperature point of the maximum torsional strengh τ_B and displays as an threshold growth of torsional plasticity θ_B and as the clear change of fracture mode from cleavage to shear fracture (Fig. 7) [12].

Similar move as a curve with the maximum has the dependance of the fatigue limit under cyclic torsion (tests on the base $5\text{-}10^6$ cycles), althougt the maximum of the fatigue strength is sihifted to the higher temperatures in comparison with static short time tests (Fig. 9). The testing of such steels under rotating bending also leads to formation to similar dependence of fatigue limit with maximum. And it is symptomatic that the point of maximum fatigue strength distinguishes with the strictly change of high cycle fatigue fracture surfaces from the brittle mode with “fish eye” to characteristic for ductile steels with typical fatigue striations on the fracture surface. Taking into account the analogy with analysis of features of fracture under short time loading such distinctions of the fracture fatigue behaviour allow to characterize the structural mechanical behaviour at left from the maximum on the curve σ_{-1} and τ_{-1} (Fig. 9) as a *brittle cyclic state* (BCS) and on the right side as *ductile cyclic state* (DCS). Similar changes of the fracture behaviour may be reached in a low alloyed quenched steels after low temperature tempering (t=150 °C) in connection with a growth of content of carbon (Fig. 1) or in connection with variation of the testing temperature.

Special probabilistic tests of one of the investigated steels (7Ch3) detected principal distinctions between probabilistic regularities of high cycle fatigue rapture for steel's structures which correspond to BCS a DCS (accordingly for tempering temperatures 150 °C and 550 °C) — Fig. 10. For plotting Wöhler curves with probabilistic distribution in rotating bending tests for every batch 150 specimens were used. Principal distinction for BCS consists in the preservation of the essential probability of fatigue fracture on very large lifetimes. And vice versa for steels in DCS clearly reveals physical fatigue limit (PhFL). The special fatigue test were performed for boundary conditions of the physical fatigue limit in tests at rotating bending and cyclic torsion in dependence on tempering temperature and carbon content (Fig. 11). There are reasons to consider the area of PhFL as the area of DCS and mentioned border lines as the lines of the brittle - ductile cyclic transition.

THE SPECIFIC MECHANISM OF FATIGUE DEMAGING AND RAPTURE OF HIGH STRENGTH STEELS

For steels which were outlined after the conditions of tests as to be found in BCS it is very difficult and practically impossible to fix the traditional scheme of fatigue fracture: crack initiation — subcritical crack growth - final quick rapture. It is connected with high strengthening of the matrix which excludes possibility of long term subcritical crack growth. Instead of that for such materials realizes practically two stage process in the course of which demaging in the area of fatigue nucleation site transits in the stage of the final spontaneous rapture. Cyclic demaging in the critical area of the specimen accumulates gradually, exhausting practically full lifetime, forming at last nucleation site of the critical (griffith type) crack. Such scheme confirmed the evolution of SIF level for the critical area: the level of K_{cr} is quite near to K_{fc} (Table 2).

Table 2 The critical levels of SIF on the nucleation sites of the ShChl5 steel

Specimen	K_{cr}, MPa $\sqrt{m}$	K_{fc}, MPa $\sqrt{m}$
1	10,8	12,2
2	11,1	12,2
3	12,0	12,2

Nucleation site of fatigue fracture forms as a rule in the stress concentration zone near the “critical” inclusion with a following development of the debonding on hard second phase particles (as usually carbides).

The stage of deterioration completes with the critical coalescence of the local microgaps in the griffith crack [13].

In accordance with mechanisms proposed by Crussard [14] and taking into account the conditions of the loading the coalescence can be fulfilled as a normal or shear rapture (Fig. 12). The completion of the griffith crack formation on the damages practically exhausts the fatigue lifetime of the specimen because the following crack propagation has instant short time character. Thus the dominant part of the lifetime should be regarded as the *process zone formation* near local structural defects (inclusions, segregations, pores or other inhomogenities) which completes with formation of the critical crack. Hence it follows that for the calculation of the lifetime of such high strength materials it is very problematic application of fracture mechanics methods. Here entrance into operation uninvestigated phenomena of the fatigue debonding, the following deterioration and coalescence acts.

CONCLUSIONS

Structural alloys of the type of the high carbon alloyed steels strengthened by quenching and thermomechanical treatment are subjected to the high cycle fatigue fracture due to the mechanism in which the main and principle importance have the processes of demaging in the site of critical particles with the next coalescence of microdetoriorations into the spontaneously moving critical crack.

CBS which corresponds to such mechanism of fatigue fracture is distinguished with sharp degenerative *V–D K* curves, also with specific probabilistic distribution of high cycle fracture, and absence of physical fatigue limit for smooth specimens. CBS and its structural areas depend on the carbon content, condition of thermal treatment, on the temperature and mode of the fatigue tests.

Distinguished role of the fatigue demaging phenomena in providing high cycle service life probably is universal enough and can be extended on the ceramics, intermetallides and strengthened composite materials.

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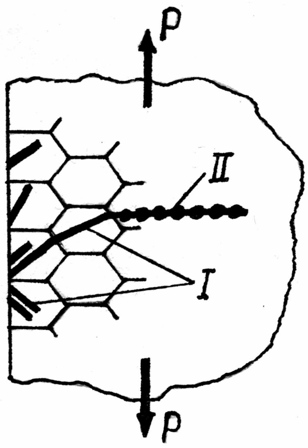


Fig 1 Two stages of fatigue fracture of metals

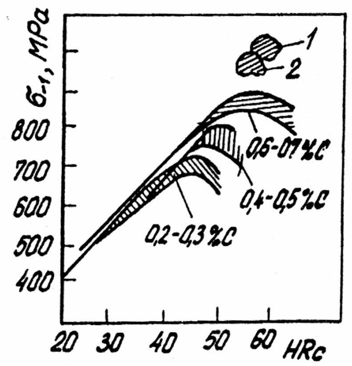


Fig. 2 Influence of carbon content and hardness HRc after quenching on the fatigue limit σ_{-1} determined at rotating bending (1, 2 – data received after thermomechanical treatment)

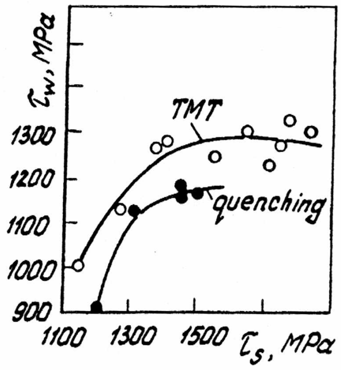


Fig. 3 The record levels of fatigue limit τ_w of 45ChN2MV steel at pulsating torsion: O – after thermomechanical treatment; ● – after control quenching

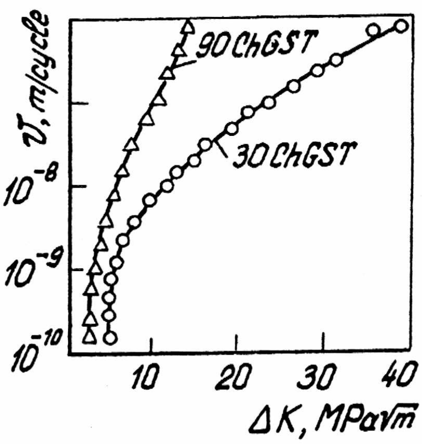


Fig. 4 Typical $V - \Delta K$ curves of low tempered high carbon (Δ) and low carbon (O) steels

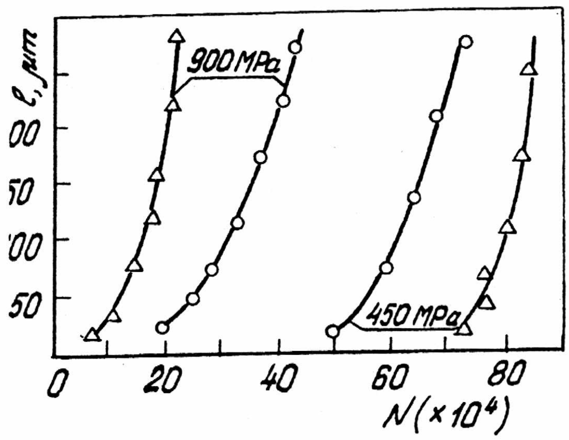


Fig. 5 Crack length kinetics fixed under rotating bending of high carbon 7Ch3 steel (Δ) and medium carbon 40Ch steel (O) after quenching and tempering (200 °C). Tests were performed on two levels of nominal stress: 450 and 900 MPa

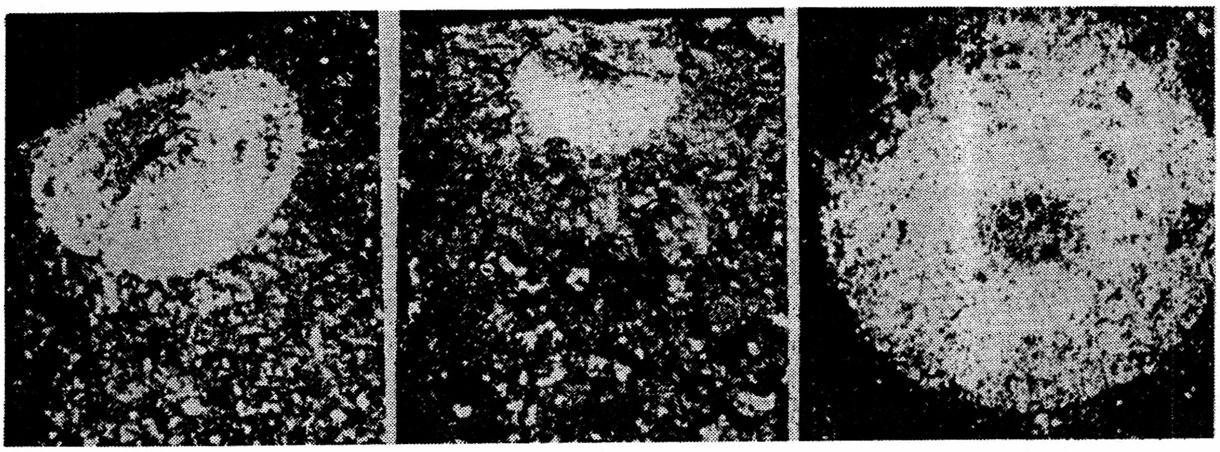
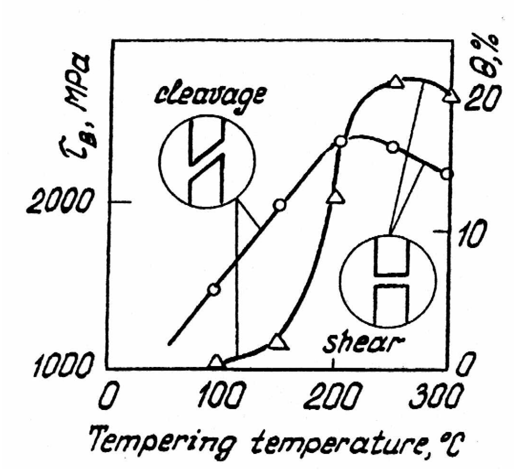


Fig. 6 Typical nucleation sites of fatigue fracture of high strength ShCh15 steel (x200)



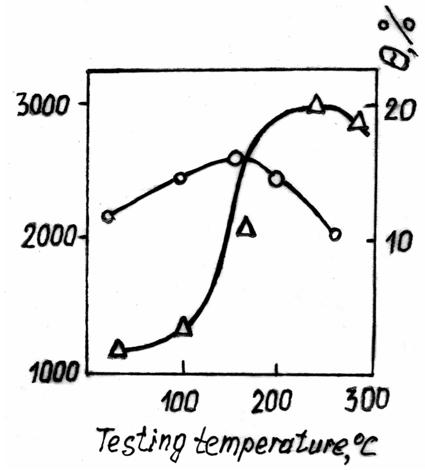


Fig. 7 Brittle – ductile transition in torsion tests of spring 9 ChGST steel, caused by tempering temperature structural changes

Fig. 8 Brittle ductile transition in torsion tests of 9ChGST steel, caused by changes of testing temperature

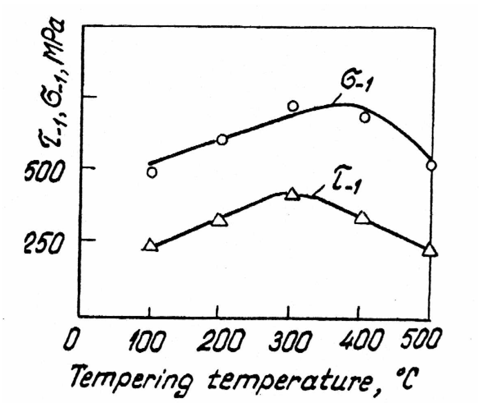


Fig. 9 Influence of tempering temperature on the fatigue limit at rotating bending σ_{-1} and cyclic torsion τ_{-1} of the spring 60ChS steel

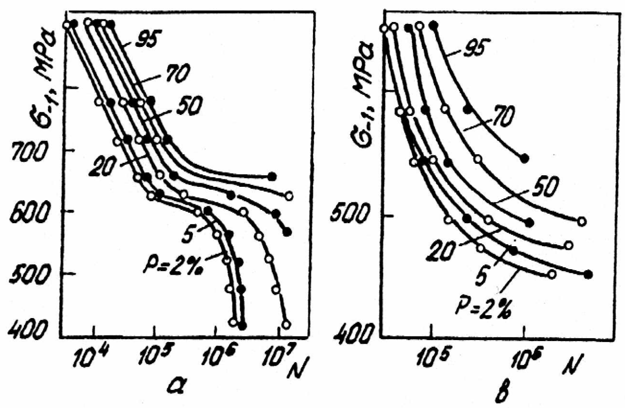


Fig. 10 Probabilistic Wöhler curves at rotating bending of the 7Ch2 steel after quenching and tempering at 200 °C (a) and 500 °C (b). Figures on curves determinate the probability of fatigue fracture in %

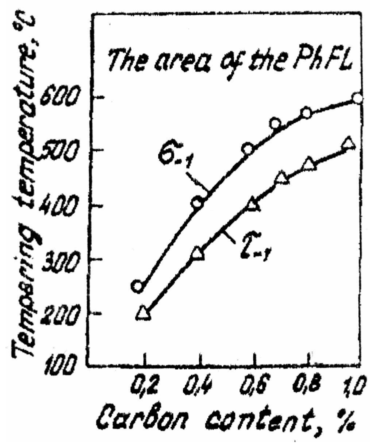


Fig. 11 The boundarie curves of the physical fatigue limit in structural chromium alloyed steels (1–2 % Cr) at rotating bending σ_{-1} and cyclic torsion τ_{-1} in connection with changes of carbon content and tempering temperature

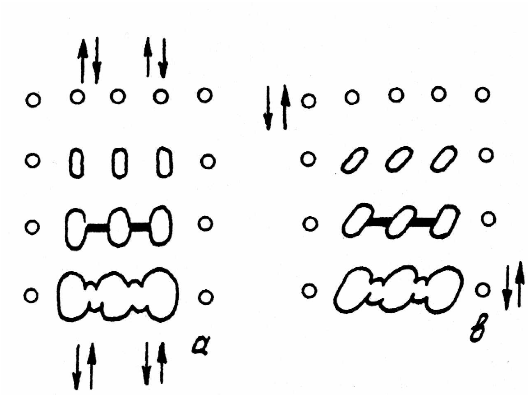


Fig. 12 Critical coalescence of demages on second phase particles as the mechanism of fatigue: a) normal rapture mode; b) shear rapture mode