Fatigue behavior of high-chromium alloyed cold work tool steel in the gigacycle regime

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Abstract

The fatigue behavior of wrought cold work tool steel of AISI D2 type (DIN NR. 1.2379) was studied up to the gigacycle regime, employing an ultrasonic fatigue testing system that operates at 20 kHz in fully-reversed tension-compression mode. The role of primary carbides and carbide clusters as potential crack nucleation sites, the effects of material anisotropy, of the degree of deformation and of surface residual stresses on the fatigue behavior have been studied. Obtained SN data revealed material failure beyond 10^6 loading cycles, similar to results observed for high strength bearing and spring steels. Material anisotropy and surface residual stresses were found to have significant effects on the fatigue behavior. The fatigue endurance strength at 10^{10} cycles was lowest for specimens with axis perpendicular to the bar rolling direction (RD), and highest for specimens with axis parallel to RD having high compressive residual stresses at the surface. Primary carbides or primary carbide clusters located near/at the surface acted as fatigue crack initiation sites. Specimens with high compressive stresses at the surface showed internal fish-eye failure originating at primary carbide clusters. Macroscopically, the fracture surfaces were rather flat, and around each crack origin a (semi-) circular fish-eye feature was obtained. Immediately around the carbide particles a granular area (GA) was observed, followed by a zone with low surface roughness. The GA was found to play an important role during the formation of small fatigue cracks. The surface morphology of fracture surface varied distinctly depending on sample orientation (axis parallel/perpendicular to RD), however, formation of GA and subsequent zone with low surface roughness turned out to be similar.

Introduction

Since the observation made by Naito et al. [1] in 1984 that carburized and surface hardened steels do no exhibit a conventional fatigue limit at 10^7 cycles, fatigue behavior of high strength steels in the very high cycle fatigue regime became of high interest. Numerous studies have been published since this time, of which a literature review has been compiled recently by the present authors [2]. The most important conclusion of these fatigue studies on high strength bearing and spring steels is that these steels fail also below the conventional fatigue limit, i.e. in the long life regime above 10^7 loading cycles, usually associated with so-called fish-eye patterns that are formed around the internal crack origins at life times above about 10^7 cycles. In most cases the internal crack initiation sites in high strength steels were found to be nonmetallic inclusions, such as Al_2O_3, TiN, SiO_2, MgO, and CaO or sulfides.

Tool steels differ from the aforementioned steels insofar as tool steels contain numerous primary carbides required for abrasion resistance, often in the same size range as the crack initiating inclusions in high strength bearing and spring steels. Generally, there are two reasons for extending the fatigue research to tool steels for which published fatigue studies are rather scarce. First, fatigue data should be generated up to the gigacycle fatigue regime in order to contribute to the ongoing discussion about the shape of the S-N curve and to answer the question whether a real fatigue limit does exist for these high strength steels. Secondly, fatigue testing at low stress amplitudes is an excellent tool for identifying material defects, especially singularities, and in case ultrasonic frequency is used it is a rather cost and time effective inspection method.
Publications on the fatigue behavior of tool steels [3-7] are scarce, in particular up to the gigacycle fatigue regime. Berns et al. [3,4] showed that primary carbides gave rise to fatigue cracks in conventional ingot metallurgy tool steels, while tool steels produced by powder metallurgy showed crack initiation at pores and, after hot working, at nonmetallic inclusions mostly in the subsurface region. Fukaura et al. [5] investigated the fatigue behavior of a JIS-SKD 11 (AISI D2 type equivalent) tool steel up to 10^7 loading cycles and observed crack initiation at one or more primary carbide particles located at the surface or close to it at stress amplitudes above 1100 MPa. Below 1100 MPa those authors found internal crack nucleation at carbide particles, forming the so-called fish-eyes at the fracture surface, as obtained for high strength bearing steels. Marsoner et al. [6] reported two types of crack starting modes occurring for PM tool steels – internal type at nonmetallic inclusions and surface type at nonmetallic inclusions or at primary carbides (aggregates). Similarly, Meurling et al. [7] observed internal oxide inclusions and carbides and also primary carbides at the surface as crack initiation sites for various types of tool steels. However, all these studies investigated the fatigue behavior of tool steels only up to $N_{max} = 10^7$ cycles. Thus, in the present work the fatigue response up to 10^{10} loading cycles of AISI D2 type tool steel was studied using ultrasonic frequency fatigue testing. The effects of surface residual stresses, of material anisotropy and of the role of primary carbides and carbide clusters as potential crack nucleation sites were investigated.

Experimental

The steel studied in this work was a wrought cold work tool steel 1.2379, acquired from *Böhler Edelstahl GmbH*, Austria, (Böhler grade K110) in annealed condition as cylindrical bars with a diameter of 15.5 mm (heavily hot worked) and 106 mm (moderately hot worked). Chemical composition of the steel in mass% was as follows: 1.55%C, 0.32%Si, 0.32%Mn, 12.7%Cr, 0.85%Mo, 0.89%V, 0.12%W and 0.12%Co. Fatigue test specimens with axis parallel to the rolling direction were prepared from the heavily hot worked steel bar. In order to evaluate anisotropic fatigue properties of conventionally produced tool steels, specimens were prepared from the larger size bars with their axis in rolling direction (RD) and orthogonal to the rolling direction (RD). All specimens were turned to the desired hour glass shaped geometry, i.e. 14.5 mm at shoulders and 4 mm at gage length, respectively, and their surface longitudinally ground prior to the heat treatment in such a way that the specimen slowly rotates while polished by a rapidly rotating disk parallel to the specimen axis. The rotating disk was spring mounted in order to avoid undesired pressure on the sample which might induce stresses in the material. The steel was then austenitized at 1040°C for 25 min and quenched in oil. Tempering was done at 530°C for 2 hrs. All heat treatments were performed under high purity nitrogen atmosphere (5.0 N_2). Relatively high tempering temperature and slow subsequent cooling was chosen in order to reduce the amount of retained austenite, to lower quench-induced residual stresses and minimize thermal stresses in the workpiece, respectively. The heat treated fatigue specimens were then polished to mirror-like finish in the longitudinal direction using 240 mesh alumina oxide abrasive paper, 15 µm and 6 µm diamond suspension, similar to the grinding accomplished prior to the applied heat treatment as

described above. Polishing using 15 µm diamond suspension was applied for material removal (up to 150 µm being removed) in order to eliminate grinding-induced surface compressive residual stresses down to a level of about +40 to -195 MPa and -195 and -340 MPa for samples with axis parallel and orthogonal to the rolling direction, respectively. However, one fatigue test series was performed without polishing with 15 µm diamond suspension in order to evaluate the effect of high compressive residual stresses (about -800 MPa at the sample surface), which resulted from the specimen grinding after the heat treatment.
Devices used for the determination of mechanical properties have been presented earlier [2], which holds also for the applied etching reagents to reveal steel microstructure. Residual stresses have been measured at the surface in the narrowest section of the fatigue samples, employing X-ray diffraction and the $\sin^2\Psi$-method [Cr-K_α, θ=78.06°, lattice plane: {211}, $\frac{1}{2}S_2=6.09*10^{-6}$].
For fatigue testing an ultrasonic testing system (*Telsonic-Ultrasonics*, Switzerland) operating at 20 kHz in fully reversed mode (R = -1) was employed, of which details have been presented recently [2]. The probability of cavitation and corrosion due to the specimen cooling by liquid coolant, which was required due to the heat developed within the samples during fatigue testing, was intensively discussed in recent work [2]; it was concluded that these effects are negligible. Fracture surfaces were examined by means of scanning electron microscopy.

Material characterization

The as-tempered steel microstructure revealed fine tempered martensite. XRD proved that the amount of retained austenite present after the applied heat treatment was rather low – a few percent maximum. The prior austenite grain size ranged at 10-15 µm regardless of the initial bar and direction. In the transverse section the primary carbides, which are chromium carbides of type M_7C_3 capable of dissolving large amounts of iron, are distributed rather homogeneously (Fig.1a). In contrast, in the longitudinal section typical carbide bands can be observed (Fig.1b). Since the primary carbides might act as fatigue crack starting points, rough classification according to the carbides and carbide cluster sizes was performed, of which the obtained data are shown in Table 1. The overall largest carbide clusters of a diameter of about 100 µm were observed in the moderately hot worked steel in longitudinal section. In general the longitudinal sections revealed significantly larger sizes of the carbides than the transversal sections, which is a direct result of the hot working of the steel bars. The carbide volume fraction was 12 ± 2 vol%. The data of the mechanical properties determined are presented in Table 2.

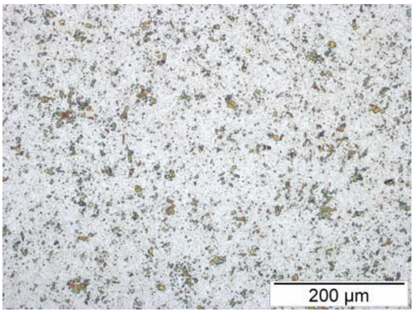

Fig. 1a: Transverse section of heavily hot worked steel revealing uniform carbide distribution.

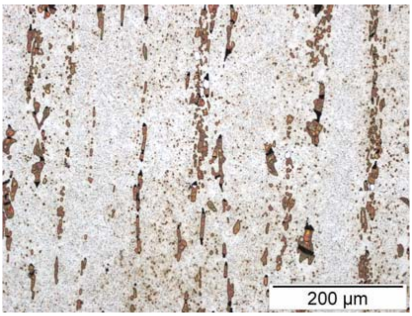

Fig. 1b: Longitudinal section of moderately hot worked steel revealing carbide bands characteristic for wrought tool steels.

Carbide (cluster) diameters (in µm)	relative frequency (in%)			
	strongly hot worked		moderately hot worked	
	transversal	longitudinal	transversal	longitudinal
20	74	68	57	not determined
40	21	19	25	36
50	4	6	10	26
60	1	6	5	18
80	0	1	2	13
100	0	0	1	6

Table 1: Rough classification of primary carbides and carbide clusters according to their sizes.

Steel AISI D2 1.2379	**Rockwell Hardness HRC 150kg**		**T.R.S. (MPa)**	**Dynamic Young's Modulus (GPa)**
	quenched	tempered	tempered	tempered
strongly hot worked (parallel to RD)	64 ± 1	58 ± 2	3600 ± 300	210 ± 7
moderately hot worked (parallel to RD)	64 ± 1	59 ± 1	3900 ± 300	206 ± 6
moderately hot worked bar (perpendicular to RD)	62 ± 2	58 ± 1	1900 ± 100 2000 ± 100	205 ± 6 (?) 211 ± 7 (‖)

Table 2: Mechanical properties of studied tool steel for the applied heat treatment. (⊥: perpendicular to carbide bands, ‖: parallel to carbide bands)

Results

S-N curves

The fatigue behavior of the studied tool steel revealed significant influence of surface residual stresses and material anisotropy, as presented in Fig.2. It turned out that high compressive stresses at the specimen surface considerably improved the endurance fatigue strength within the entire cycle number range, which was studied for specimens with axis parallel to the rolling direction of heavily hot worked material. Furthermore, these compressive stresses at the surface seemed to inhibit the crack nucleation in the surface region, thus, up to about 10^7 loading cycles exclusively internal fish-eye failure due to large carbide clusters were obtained (Fig.2 □). However, residual stress measurements at runout specimens (N=10^{10} cycles) indicated a relaxation of these residual stresses to about one half of the initial value, which probably enabled the at/near-surface crack nucleation at carbides and carbide clusters located at or just below the specimen surface (Fig.2 ■). The fatigue endurance strength at 10^{10} loading cycles was about 200 MPa higher for specimens with high compressive stresses compared to specimens with low ones (Fig.2▲), the latter showing nearly exclusively near-surface failures. Detailed discussion and statistical consideration of the occurrence of internal and at/near-surface failures as a function of surface compressive residual stresses have been presented recently [2].
The material anisotropy, i.e. the anisotropic distribution of the primary carbides, showed a rather significant influence on the fatigue response. The fatigue strength for specimens with axis orthogonal to the rolling direction (Fig.2●) – for which compressive stresses in the range of -195 and -340 MPa were present at the specimen surface – was lower compared to specimens with low residual stresses and axis parallel to the rolling direction. Exclusively at/near-surface failure due to large carbide clusters was observed. Furthermore, it turned out that the degree of deformation did not show any effect, since specimens with axis parallel to the rolling direction from the moderately hot worked bar failed (Fig.2◊) within the same cycle number range as the corresponding samples machined from the heavily hot worked bar (Fig.2▲).

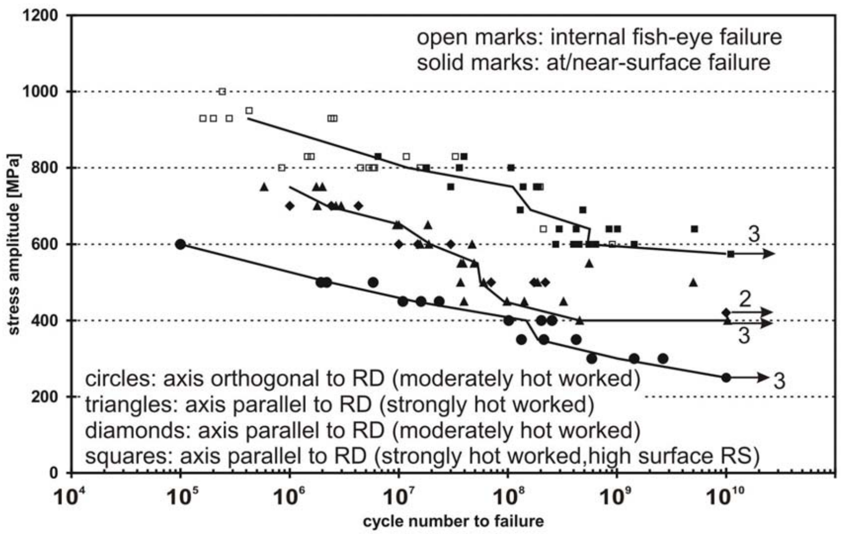

Fig. 2: S-N data for AISI D2 type tool steel
(Figures close to the arrows indicate number of runout - N=10^{10} cycles - specimens).

Fractography

Fig.3a shows the fish-eye pattern formed around the crack initiating closely arranged primary carbides, i.e. carbide clusters (Fig.3b). In the surrounding of the carbides a granular area was observed, such as marked in Fig3b. Specimens with low residual stresses (initially or after partial relaxation) failed due to carbides or carbide clusters located at or just below the surface, as presented in Fig.3d and Fig.4b. Similarly to internal failures also for at/near-surface failures a granular area around the crack origin was found. Outside of this granular area, a zone with rather low surface roughness was detected (Fig.3c,4a). This low surface roughness might be due to the very slow crack growth here, which can be supposed to result in crack growth exactly perpendicular to the stress orientation. Comparing Fig.3c,d to Fig.4b,c it can be seen that with longer fatigue life time the sizes of the mentioned zones definitely increased significantly. Detailed investigation of the obtained fracture surfaces has been described recently [8].

Fig.5 presents a comparison of the macroscopic appearance of the fracture surface of specimens with axis parallel to the rolling direction (Fig.5a) and transverse to the rolling direction (Fig.5b). Fig.5a shows a rather flat fracture surface, with cracks and ridges pointing back to the crack origin, which is marked by an arrow. In contrast, from Fig.5b it can be seen that fractures of transverse specimens exhibited a completely different picture. Here, numerous cracks oriented parallel to the primary carbide bands were observed. However, the microscopic appearance around the crack origins, which were large carbide clusters (Fig.6a,b), seemed to be rather similar to the fractures of samples with axis parallel to the rolling direction. Again, a granular area was observed in the surrounding of the crack nucleating carbides. Detailed discussion of the effect of material anisotropy on fatigue response of AISI D2 type steel has been presented earlier [9]. Note that a real fatigue limit was not obtained for the studied steel in any case and that all obtained S-N curves had a steadily decreasing shape.

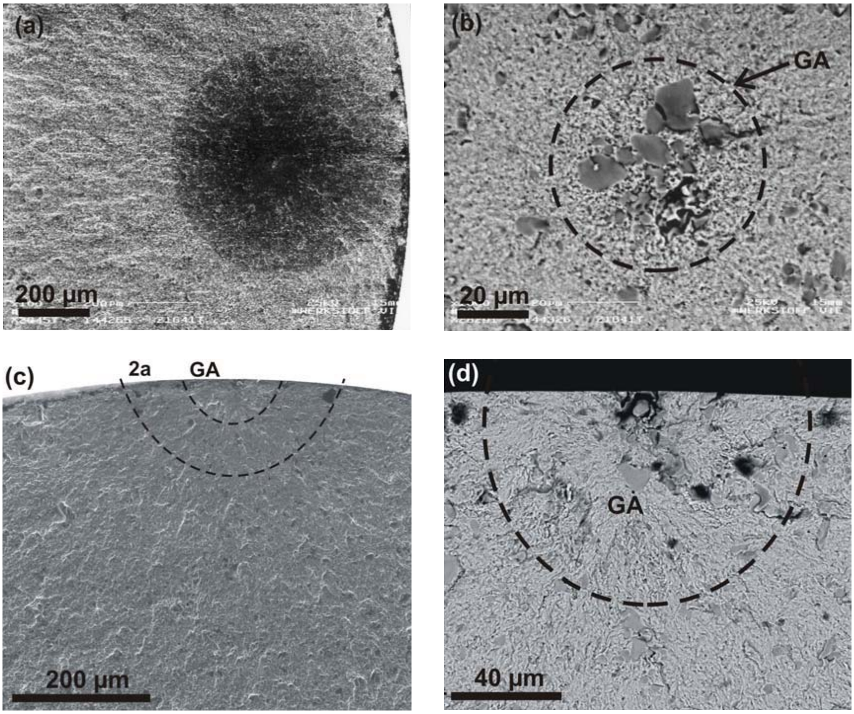

Fig.3: Fractographs of specimens with initially high compressive residual stresses at the surface: (a), (b) internal fish-eye failure of specimen failed at 800 MPa after 3.6×10^7 cycles; (c), (d) failure due to surface carbide at 640 MPa after 2.96×10^8 cycles.

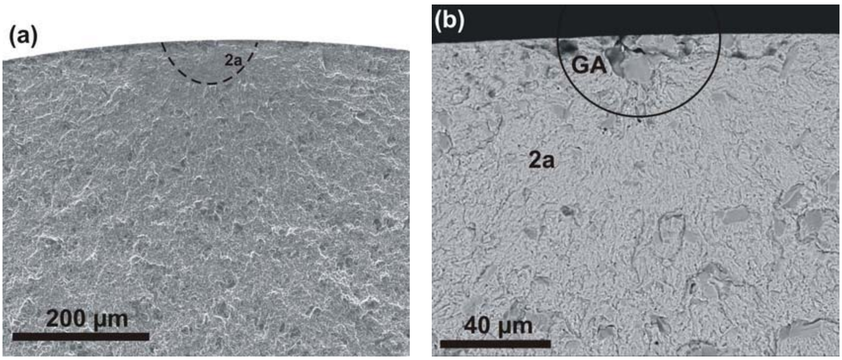

Fig.4: Fractographs of specimen with low surface residual stresses failed at 750 MPa after 9.36×10^5 cycles.

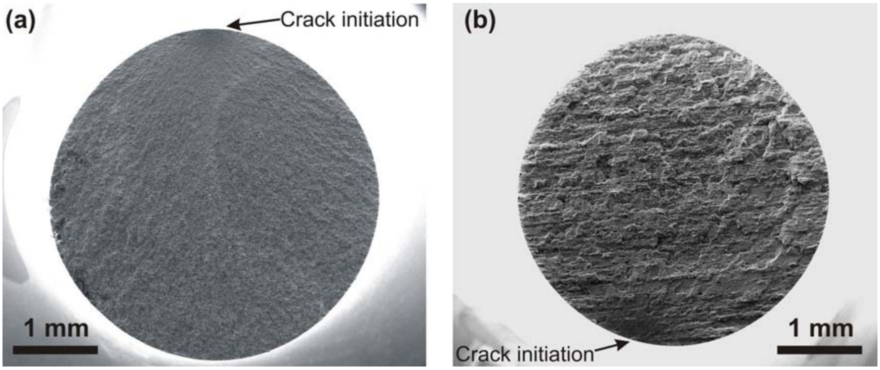

Fig.5: Macroscopic appearance of fracture surfaces of (a) specimens with axis parallel to rolling direction and (b) orthogonal to rolling direction.

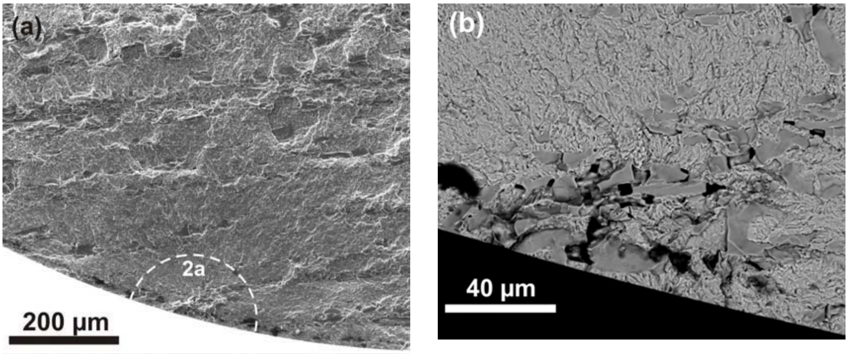

Fig.6: Fractographs of specimen with axis orthogonal to rolling direction that failed at 500 MPa after 2.19×10^6 loading cycles.

It is speculated that the granular area desribed above is formed according to a model introduced by Shiozawa et al. [10], called "dispersive decohesion of spherical carbides". This model suggests that multiple microcracks are initiated by decohesion of small carbides (diameter <1 µm) from the iron matrix. It can be supposed that this process of microcrack formation and subsequent coalescence of these microcracks is rather slow. The border of the granular area and the subsequent area "2a" probably marks the transition from a non-propagating to a propagating fatigue crack. More precisely, the short crack, formed through coalescence of numerous microcracks within the granular area, reached the length of a propagating crack. It can be assumed that before this crack length is attained, fatigue crack growth can only take place by the described microcrack coalescence. Thus, it seems that the formation of the granular area plays an important role in the fatigue process of the studied steel. The sizes of the carbide clusters, the granular area and of stage 2a were evaluated in order to obtain information about the fatigue process. Interesting results were obtained when combining the two datasets for carbide cluster size and granular area sizes observed in fractures of transverse and longitudinal specimens. Fig. 7 presents the relationship to the applied stress amplitude. Obviously, the radii of carbide clusters rather revealed two data sets – one for transversal samples (Fig. 7▲) and one for the specimens with axis parallel to rolling direction (Fig. 7Δ) – but definitely not a correlation with the applied stress amplitude. In contrast, the combined data of granular area size (Fig. 7c ●,○) could be correlated linearly to the applied stress amplitude, further supporting the idea that the size of the granular area represents the critical crack length for starting of a propagating short fatigue crack. It seems that the dimensions of carbide clusters relevant in case of longitudinal loading (i.e. those dimensions that are observed in the transverse section of the steel) were not large enough to form appropriate sized granular area, the size of which is dependent on the applied stress. Consequently, failures of specimens parallel to the rolling direction were not observed below 400 MPa within the maximum tested cycle number of 10^{10} cycles. It can be speculated that with these materials the formation of the granular area might take longer than 10^{10} cycles to attain the critical size for a propagating fatigue crack. In contrast, the carbide (cluster) dimensions relevant for transverse loading (i.e. the dimensions taken from longitudinal sections) were far larger, thus, within the tested cycle numbers appropriate sized granular was able to be formed in transverse fatigue specimens, at least down to stress amplitudes of 300 MPa, below which however the critical granular area size was not attained within $N_{max}=10^{10}$ cycles.

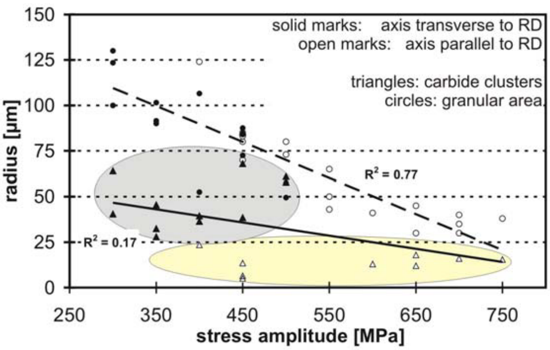

Fig.7: Relationship of applied stress amplitude with the observed crack initiating carbide cluster sizes and sizes of surrounding granular area for specimens with axis transverse and parallel to rolling direction (RD).

Summary

Fatigue behavior of heat treated AISI D2 type tool steel was studied, showing considerable influence of surface residual stresses and material anisotropy. It has been shown that primary carbides and carbide clusters located in the interior and at/just below the surface were origins of fatigue cracks. A granular area was observed around the crack initiating carbides followed by a rather smooth zone. The granular area seems to play an important role in the fatigue process, which can probably be described by the model "dispersive decohesion of spherical carbides" suggested by Shiozawa et al.[10]. A real fatigue limit was not attained in testing up to 10^{10} loading cycles.

Acknowledgements

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