Machining factor on the high cycle fatigue life of X80 pipeline steel

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Abstract: Pipeline steel is used to make long pipes transporting fossil oil and natural gas, and the stability of its fatigue life is important to evaluate the pipeline service reliability. However, a fatigue life scatter often occurs at low stress amplitudes in X80 pipeline steel. In this investigation, the possible influencing factors have been studied with observation on the fatigue fracture morphology, microstructure and dislocation pattern, and analysis on the distribution of micro-Vickers hardness. It is found that the remaining transverse machining marks greatly affects the fatigue lives at low stress amplitudes. It is mainly because the critical threshold stress factor is greatly influenced by the depth of transverse machining mark. Removing the remaining transverse marks totally as much as possible can greatly improve the fatigue life scatter.

1. Introduction
Pipeline steel is usually used for making long pipes of fossil oil and natural gas [1, 2]. In recent decades, with increasing demands of fossil oil and natural gas in industries, pipeline steel develops very fast and is widely applied all over the world [3], especially in China. Mostly, the pipe has to survive from its surrounding environmental conditions, such as the large temperature range, corrosion environment and fatigue damage [4-6]. Fatigue fracture is one common kind of failure modes of the pipe, and it is always caused by the alternate load in the pipe which may be induced by the fluctuation pressure of fossil oil and natural gas, or the temperature and other environmental factors, therefore numbers of investigations on the fatigue properties of pipeline steels have been conducted [7-9].

Among all kinds of fatigue tests of pipeline steels, high-cycle fatigue (HCF) test is a widely used to measure the fatigue lives at different stress amplitudes to evaluate the service reliability of the steels [7, 10-11]. The HCF properties consist of S-N curve and fatigue limit, which show the relationship between the fatigue life and the stress amplitude. Commonly, the S-N curve obeys the Basquin law which shows a linear trend of the stress amplitude vs. number of reversals to failure in the double-logarithmic coordinate. According to the previous studies [12-15], the HCF life is mainly determined by the fatigue crack initiation which prefers to originate at inclusion or at persistent slip band (PSB) in the surfaces or subsurface layers of the specimens. For the high-strength steels, non-metallic inclusions are often observed at fracture origin, and the fatigue life is referred to be
inversely related to the size of the inclusion [13-14]. In addition, the machining technique of the specimen surface also affects the fatigue crack initiation according the ASTM standard [15]. All these reasons contribute to the scatter of the fatigue lives, leading to a distribution far away from the linear trend. Compared to the inclusion and the PSB, the defect from the machining technique makes the situation worst to harmonize the test results from the different laboratories.

In this investigation, the factors especially the defect from the mechanical polishing on the scatter of the fatigue lives at low stress amplitudes of X80 pipeline steel have been studied and the improvement on the surface machining technique has also been suggested.

2. Experimental materials and procedures

In this study, an API X80 pipeline steel as-rolled was investigated, and its chemical compositions are shown in Table 1. The mechanical properties are listed in Table 2. It can be seen that this steel is a kind of mild steel with a good combination of strength and ductility. Its yield ratio is about 0.84.

![Figure 1. The dimensions of fatigue specimens.](image)

All the specimens were mechanically polished along the longitudinal direction after finish machining. The dimensions of the fatigue specimens are shown in Figure 1. The cyclic push-pull tests were carried out at high frequency of about 95 Hz at room temperature with the sinusoidal wave form under applied stress ratio of R = -1 up to 107 cycles using an electromagnetic resonant fatigue testing machine (Zwick Amsler 150 HFP5100). After fatigue tests, the fatigue specimens were broke along the fatigue crack plane with a bending machine WB-2000. The fatigue fracture surfaces were observed with scanning electron microscopy (SEM) of JEOL S-3400N. In addition, the microstructure and the corresponding micro-hardness near the fracture surfaces are observed and measured with SEM of JEOL S-3400N and LECO LM300AT automatic hardness testing system, respectively. The load for micro-hardness was 0.2 kgf for 10 seconds. Finally, the dislocation patterns near the fracture surfaces have been observed with transmission electron microscope (TEM) of JEOL JEM-2100F.

3. Experimental results and discussion

3.1. Microstructure

Figure 2 shows the SEM microstructure of the X80 pipeline steel. A kind of banded structures can be observed in Figure 2a since the specimen is cut from a rolled plate. According to the references [16, 17], these banded structures may be related to the prior austenite grains which were rolled. In Figure 2b, a mixed microstructure of polygonal ferrite (PF) and granular bainite (GB) is revealed. GB consists of fine lath ferrite and some island constituents distributed in the matrix. The island constituents commonly named martensite/austenite (M/A) island are transformed from retained...
austenite during accelerated cooling processes. This kind steel has a good combination of high-strength and excellent ductility, since PF can sufficiently provide ductility, and GB was composed of fine lath ferrites, and M/A constituents [18].

3.2. HCF properties

Figure 3 shows the virgin data for HCF lives of X80 pipeline steel at different stress amplitudes. The blue pots represent the broken specimens and the number next the blue pot in the figure is the broken specimen sequence number. The black pots represent the non-broken specimens, and the black number pointed by the black arrow shows the total number of the non-failure specimens. In the Figure 3, it can be seen that the fatigue life increases in a linear trend with decreases the stress amplitude in the range from 320 MPa to 365 MPa. However, these fatigue lives become very scatter when stress amplitudes are close to the fatigue limit 294 MPa, the fatigue lives of some failure specimens such as 4#, 10# and 12# are far away from the linear trend.

For the stress-controlled fatigue test, fatigue crack can initiated from PSB, non-metallic inclusions or defects from surface mechanical machining. PSB forms as a result of the multiplication, growth or tangling of dislocation, which is related to the local deformation of the microstructure. The non-metallic inclusions can be seen as micro-cracks, so the fatigue lives are affected by their sizes and the strength of the microstructure surrounding them. Even though good preparation of fatigue specimens has been done, the defects induced by the surface mechanical machining such as cracks, transverse machining marks, gouges and undercuts may still occur, and they can lead to the short crack propagation since their sizes are small. Therefore all these factors may the reasons for the fatigue life scatter of X80 at the low amplitudes.
3.3. Discussion on the fatigue life scatter
To investigate the mechanism of HCF life scatter at low stress amplitudes for X80 pipeline steel, the fatigue fracture surface morphologies, microstructures, micro-hardness, and the dislocation patterns near the fracture surfaces at different stress amplitudes have been studied in this section.

3.3.1 Fatigue fracture morphologies. The fatigue fracture surface morphologies of specimen 2# and 3# are shown in Figures 4-5, respectively. The stress amplitude for specimen 2# is 350 MPa, and its fatigue life is 195097 cycles. Figure 4a shows its fatigue crack source observed with secondary electron (SE). It can be observed that a smooth plane appears in the fatigue crack source and the crack initiation is from the corner between this plane and the surface. This plane can be inferred as the microscopic geometry defect which may from surface mechanical machining. In Figures 4b-c, second cracks can be observed in the stable crack propagation regions. This phenomenon may be related to the banded structure since the second crack does not cut though the white and smoother band. The unstable crack propagation region can also be seen as the final rupture region in Figure 4d, even thigh the specimen did not break completely at the final cycle. In this region, large plastic deformation can be observed from the distribution of slip bands. The stress amplitude for specimen 3# is 320 MPa, and its fatigue life is 1836466 cycles. Figure 5a-b shows its fatigue crack source observed with secondary electron (SE). It can be observed that the fatigue crack initiation is from a defect in the surface. In Figure 5c, fatigue striations and secondary crack can be observed, which are pointed by the white arrows. In Figure 5d, the fracture surface in the unstable crack propagation region seems flat and a lot of second cracks distribute dispersively in that region.

Figure 4. Fatigue fracture surface morphologies of specimen 2# ($\sigma_a = 350$ MPa, $N_f = 195097$): (a) crack source; (b) and (c) stable crack propagation region; (d) unstable crack propagation region.
Figure 5. Fatigue fracture surface morphologies of specimen 3# ($\sigma_a = 320$ MPa, $N_f = 1836466$): (a) and (b) crack source; (c) stable crack propagation region; (d) unstable crack propagation region.

Figure 6. shows the morphologies along the fatigue fracture profile of specimens at high stress amplitudes. From Figures 6a and d, it can be observed that the fatigue crack initiation from specimens 7# and 3# are induced by the surface local debonding. This is coincident with the fatigue crack source in Figure 5b. From Figures 6b and c, it can be observed that the fatigue crack initiation is induced by the local plastic deformation of the surface for specimens 2# and local crack along the transverse machining mark for specimen 6#.

Figure 6. Morphologies along the fatigue fracture profile of specimens: (a) 7# ($\sigma_a = 365$ MPa, $N_f = 49702$); (b) 2# ($\sigma_a = 350$ MPa, $N_f = 195097$); (c) 6# ($\sigma_a = 335$ MPa, $N_f = 268054$); (d) 3# ($\sigma_a = 320$ MPa, $N_f = 1836466$).
The fatigue fracture surface morphologies of specimen 4# are shown in Figure 7. Specimen 4# is at stress amplitude 305 MPa, and its fatigue life is 369236 cycles. Figure 7a shows its fatigue crack source observed with secondary electron (SE), which initiate from the corner between a smooth plane and the surface like in the Figure 4a. In Figure 7b, there are two regions with different morphologies in parallel distribution. The smooth one is named region A, and the rough one is named region B, which are similar to Figures 4b-c. Compared with Figure 7a, the rough morphology in parallel bands may imply the reason for the distribution for regions A and B. Just like the yellow arrow pointed, the fatigue crack propagates from the microscopic geometry defect in surface to the inner part of the specimen. However, this propagation route is not flat, but in different planes. This single plane is the region A. The region B is the aggregates of the planes. This phenomenon may be related to the glide step. The microstructure is in a banded structure according to Figure 2a and mixed with PF and GB where slip band is the deformation mode as is shown in Figure 7c. When the crack propagates along the slip band as is shown in Figure 7c, a smooth plane can form, which is the region A. However, crack propagation does not always carry on along one slip plane, therefore the coalescence of microcracks along different slip planes should appear, which may form region B. When the maximum loading stress increases with increasing the area of crack planes, the formation of second crack which is shown in Figure 7d can be realized.

Figure 7. Fatigue fracture surface morphologies of specimen 4# ($\sigma_a = 305$ MPa, $N_f = 369236$): (a) and (b) crack source; (c) stable crack propagation region; (d) unstable crack propagation region.
Figure 8 shows the morphologies along the fatigue fracture profile of specimens at low stress amplitudes. From Figures 8a-c, it can be observed that the fatigue crack initiates along the transverse machining mark of specimens 4#, 10# and 12#. Figures 8d-f show the surface local debonding for specimens 9#. It can be seen a lot of small steps in Figure 8d. Figure 8e is the amplified zone in black dash circle in Figure 8d. Some microcracks form along the steps. Then some cracks inflect in perpendicular direction which implies the break of the steps. In the region far away from the fracture surface of specimen (Figure 8f), these phenomena can also be observed. Since the stress amplitude is low enough that the plastic deformation is too small to form the glide step, the steps in the surface near and far away from the fracture surface should not be the glide steps, but the other kind of microscopic geometry defect. In conclusion, the microscopic geometry defect is the mainly reason for the short fatigue life of X80 pipeline steel.

Based on the morphologies on the fracture surfaces and along the fatigue fracture profile, the fatigue crack initiation is from the surface local debonding or the local crack along the transverse machining mark for the specimens which fatigue lives distribute along the linear trend. For the former, it is probably induced by the local plastic deformation. For the specimens which fatigue lives distribute far away from the linear trend, fatigue crack initiation originated from the local crack along the transverse machining mark. Therefore it can be referred that the transverse machining mark rather than the non-metallic inclusion is a factor for the fatigue life scatter of X80 at the low amplitudes in Figure 3, since no non-metallic inclusion can be observed in the fatigue crack source.

3.3.2 Evolution of microstructure and hardness. In general, cyclic hardening or softening appears after numbers of pull-push cycles, especially at a high stress amplitude, because the dislocation multiplication or annihilation during the plastic deformation can strength or soften the metallic materials. Therefore, microstructure evolution and micro-Vickers hardness near and far away from the fracture surfaces at low and high stress amplitude are investigated in this section.

Figures 9-10 show the microstructures of specimen 2# near and far away from the fracture surface with SEM, respectively. The banded microstructure is not clear in Figure 9a. From the amplified region in the red rectangle solid frame (Figure 9b), it can be observed that the matrix mainly consists of PF, and M/A islands which seem in banded distribution. The microstructure in Figure 10a is similar to that in Figure 9a. There are no banded structures. However, banded structures can be observed in
Figure 10b. From the amplified Figures 10c-d of the region in the yellow dash cycle in Figure 10b (SE in Figure 10c means secondary electron; BSE in Figure 10d means backscattered secondary electron.), it can be observed that the banded structures are mixed with PF bands and GB bands. The size of M/A islands in GB band is small, and the number density is high.

Figure 9. Microstructures of specimen 2# near the fracture surface with SEM: (a) the whole; (b) amplified region in the red rectangle solid frame in Figure 9a.

Figure 10. Microstructures of specimen 2# far away from the fracture surface with SEM: (a) the whole; (b), (c) and (d) banded structures.
Figures 11-12 show the microstructures of specimen 10# near and far away from the fracture surface with SEM, respectively. The banded microstructure is not clear in Figure 11a, and clear in Figure 11b. Fine precipitates and tiny inclusions can be observed in Figures 11c-d, respectively. There are no banded structures in Figure 12a. Tiny inclusions can also be observed in Figure 12b.

The diagrammatic sketch of micro-Vickers hardness indentation positions for different microstructures near and far away from the fracture surfaces is shown in Figure 13. Three positions were chosen in a plane perpendicular to the fatigue fracture surface (the upper edge of Figure 13). Some were along the edge pointed by the black arrow at the fatigue source, which were named “E”. Another ones which were along the band structure pointed by the black arrow in the center of the cross section corresponding to the fatigue fracture surface, were named “C”. The last ones which were perpendicular to the banded structure were named “P”. In addition, “N” and “F” were used to named the positions which were near and far away from the fatigue source. The micro-Vickers hardness at
different positions of specimen 2# and 10# are shown in Figure 14. The hardness far away from the fatigue source can be assumed as the hardness of the original material, since it is little affected by the cyclic loading according to the hardness at the positions FC and FP for specimen 10#. Oppositely, the hardness near the fatigue source is more probably affected by the cyclic loading. The result shows that the hardness at the positions NC is higher than those in the center NC and FC for specimen 2#, and an inverse conclusion can be gain for specimen 10#. It means the cyclic hardening and softening exist at the high and low stress amplitudes, respectively. However, these changes are very slight compared with the difference of hardness between specimen 2# and specimen 10#. Therefore no matter at high or low stress amplitude, it is similar for the strengthening effect induced by the plastic deformation of local microstructure near the fracture surface during the fatigue crack initiation and propagation, since the values of hardness at chosen zones have no obvious difference.

Figure 13. Diagrammatic sketch of the indentation positions for different microstructures near and far away from the fracture surfaces.

Figure 14. Comparison of Micro-Vickers hardness at different positions of specimens 2# and 10#.

3.3.3 Dislocation patterns near the fracture surfaces. Dislocation patterns near and far away from the fracture surfaces at low and high stress amplitude are observed with TEM in this section, which can help to prove the existence of glide steps.
Figure 15. Dislocation patterns of specimen 2# near and far away from the fracture surface with TEM: (a) and (b) near; (c) and (d) far.

Figure 15 shows the dislocation patterns of specimen 2# near and far away from the fracture surface with TEM. The dislocation density is very higher in the M/A island than that in the PF grains, which is shown in Figure 15a. In Figure 15b, Dislocation tangleing in the PF grain can be observed. In addition, dislocation tangleing and intersection appear in Figure 15c, and the vein structure begins to form. All of these phenomena imply that the cyclic plastic deformation is too low. In addition, dislocation locking by the precipitate is shown in Figure 15d. At low stress amplitudes, dislocation tangleing can be observed in most regions (Figures 16a and c), and dislocation intersection and formation of vein structure can be observed in local regions (Figures 16b and d). These imply that the cyclic plastic deformation is also small at low stress amplitudes.

Based on the observation on the dislocation patterns, it can be inferred that the cyclic plastic deformation is too small to form the PSBs, which is accordance to the results of micro-Vickers hardness.
3.3.4 The influencing mechanism of the transverse machining mark on the fatigue life. For HCF life of X80 pipeline steel, the machine mark and surface surface local debonding are the main factors on the fatigue crack initiation. Compared with the latter factor, the former performs a dominate role in decreasing the fatigue life at close to the fatigue limit. The reason for this can be explained by the short fatigue crack theory and the model of the plastic zone length of the crack tip based on the stress field strength factors of model I crack [12, 19]. For a certain metallic material, the yield strength is constant. When increasing the loading stress, the plastic zone length of the crack tip increases. Based on this assumption, equation (1) makes it easy to understand the influencing mechanism of the transverse machining mark on the fatigue life according to the short fatigue crack theory [12].

\[
\sigma_{th} = \frac{K_0}{\sqrt{\pi (a + a_0)}}
\]

where, \(\sigma_{th}\) is the critical threshold stress factor, \(K_0\) is the stress intensity factor at the tip of the slip band, \(a\) is the length of short crack, and \(a_0\) is the length of slip band along the crack direction. The transverse machining mark affects \(\sigma_{th}\) through its depth contributes to \(a\) and the length of plastic zone contributes to \(a_0\) at the mark tip. The local slip band from the plastic deformation affects \(\sigma_{th}\) through its length contributes to \(a_0\). At high stress amplitudes far away from the fatigue limit, it needs the large plastic zones to fracture at the crack tip, so the transverse machining mark is not the only factor, and the local slip band becomes another
factor since the latter is positive with the loading stress. Therefore it is not surprise to observe that both the surface debonding and the transverse machining mark can lead to fatigue crack initiation. At low stress amplitudes close to the fatigue limit, it is needed small plastic zones to fracture at the crack tip. The local slip band becomes less competitive with the transverse machining mark. And the depth of the latter makes the $\sigma_{th}$ and fatigue life scatter since the depths of transverse machining marks in the surfaces of the specimens are not totally same.

Based on the analysis above on the influencing mechanism of the transverse machining mark on the fatigue life, it can be concluded that the machining technique that mechanically polished along the longitudinal direction is not enough to eliminate the influence from the machining process, necessary magnified observation on the surface of specimen is needed to ensure that no transverse machining mark remains.

4. Conclusions
In this investigation, the HCF properties of API X80 pipeline steel have been measured, the factors on the scatter of the fatigue lives at low stress amplitudes of X80 pipeline steel have been studied with observation on the fatigue fracture morphology, microstructure and dislocation pattern, and analysis on the distribution of micro-Vickers hardness. Finally the influencing mechanism of the transverse machining mark on the fatigue life is discussed and the suggestion on the improvement is supposed. Several conclusions can be drawn:

1. The microstructure of X80 pipeline steel is a mixed microstructure of polygonal ferrite and granular bainite. The hardness evolution of the microstructure along the fatigue crack is seldom affected by the deformation from the fatigue crack propagation at different stress amplitudes.

2. Both the surface local debonding and the transverse machining mark are the factors on the fatigue crack initiation of X80 pipeline steel, and the latter is the dominate factor on the scatter of the fatigue lives at low stress amplitudes.

3. The fatigue life is shortened at the low stress amplitude because the depth of remaining transverse marks decreases the critical threshold stress factor, so eliminating the remaining transverse marks totally can improve the fatigue life scatter.

Acknowledgements
The authors would like to thank Mrs. Qun Ren for the help of the operation of JEOL S-3400N, Mrs. Jingli Hao for the preparation of specimens for TEM, and Mrs. Ping Lai and Huiping Jia for the operation of LECO LM300AT.

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