Changes in microstructure and mechanical properties during the bending process of NM450 wear-resistant steel

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Abstract

Bending deformation is a typical forming method in the manufacturing process of mechanical equipment. NM450 wear-resistant steel with high strength is vulnerable to bending failure, thus decreasing the bendability. The changes in microstructure and mechanical properties of NM450 wear-resistant steel during the bending process were studied by scanning electron microscopy, energy dispersive spectroscopy, optical microscopy, transmission electron microscopy, electron back-scattered diffractometer methods, and tensile test. The inner and outer arcs around the bending axis underwent compressive stress and tensile stress, respectively, causing an obvious bending deformation of the martensite laths with a high dislocation density. The tensile strength around the inner- and outer-arc of samples exceeded that at the mid-thickness. During the bending process, the microsized TiN inclusions nearby the outer arc were subjected to tensile stress and multiple microcracks/cavities perpendicular to the outer arc were formed. However, the microsized TiN inclusions nearby the inner arc underwent compressive stress, initiating the occurrence of divergent microcracks.

1. Introduction

Owing to its strength and excellent wear resistance, NM450 wear-resistant steel is widely used for large mechanical equipment under tough working conditions [1–3]. Bending deformation is a typical forming method in the manufacturing process of mechanical equipment. However, NM450 steel is vulnerable to bending fracture due to the quite low ductility and toughness caused by its high strength. The bending deformation capacity of low-alloy wear-resistant steel is related to the roughness of the steel surface layer, the hardness at the surface and subsurface layers, the microstructural constituents, and the steel-plate thickness [4, 5]. The bending performance can be improved by lowering the hardness and improving the uniformity of the microstructure at the surface and subsurface layer, which increase the uniformity of the strain distribution along the bending axis [6]. During the bending process, the large size brittle martensite-austenite (MA) constituents or inclusions at the sub-surface layer are broken or separated from the matrix, easily giving rise to the cavities, thus forming the bending cracks [7].

The mechanical properties and wear resistance of low-alloy wear-resistant steel are improved by adding microalloying element Ti. The nanosized TiN refines the grains and improves the mechanical properties of the steel [8, 9], but the microsized TiN inclusions tend to generate stress concentration during the deformation process, resulting the formation of cracks, which reduce the impact toughness and fatigue performance of the steel [10, 11]. The fracture mechanisms of TiN inclusions during the tensile process have been studied in the low-alloy wear-resistant steel [12, 13]. It is found that the microsized TiN inclusions easily cause the cleavage fracture, deteriorating the mechanical properties. However, the study on the changes in the microstructure and fracture mechanism of TiN inclusions in the low alloy wear-resistant steel during bending is still no available.
The present paper experimentally studies the changes in the microstructure and mechanical properties of NM450 steel and the failure mechanism of the microsized TiN inclusions during bending process. The work provides a basis for applications of NM450 steel in the large machinery construction.

2. Experimental procedures

The experimental samples were prepared from a 20-mm-thick NM450 wear-resistant steel plate quenched at 910 °C. The chemical compositions (wt%) of the NM450 steel are given in Table 1.

|   | C    | Mn  | Si  | S    | P    | Cr  | Ti  | Alt | O    | B    | N    |
|---|------|-----|-----|------|------|-----|-----|-----|------|------|------|
|   | 0.20 | 1.58| 0.52| 0.001| 0.010| 0.33| 0.016| 0.032| 0.0004| 0.0017| 0.0028|

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The specimen was bent through 30° with a bend radius of 40 mm. Samples for analysis were taken from the mid-thickness and the inner and outer arcs (subjected to compressive and tensile stress, respectively) at two positions from the bending axis (8 mm and 147 mm). Cylindrical tensile samples were taken at the distance of 8 mm and 147 mm from the bending axis. The locations of the tensile samples are shown in Figure 1. Specimens with a gauge length of 17.5 mm and a diameter of 3.5 mm were prepared according to the China national standard GB/T 228.1–2010. Tensile test was carried out at a tensile rate of 10 mm min⁻¹ on a universal tensile testing machine (SANS CMT5105) at room temperature. The microhardness was tested by a digital hardness tester (XHHD-2000TMSC) under a 0.5 kg force along the thickness direction at the distance of 8 mm and 147 mm from the bending axis. The distance between the neighboring indentations near the surface and the center of the thickness was 0.2 mm, and the other positions in the thickness direction were 0.5 mm.

The TiN inclusions in the thickness direction at the distance of 147 mm from the bending axis were analyzed by optical microscopy (OM, OLYMPUS-BX51M). 293 TiN inclusions were found in a 110 mm² area. The fracture surface of the tensile specimens was observed by scanning electron microscopy (SEM, Quanta 250 FEG) and the inclusions at the fracture surface were analyzed by energy-dispersive x-ray spectroscopy (EDS, INCA-ENERGY). The morphologies of the martensite laths were observed by transmission electron microscopy (TEM, JEM-2100F). For TEM, thin slices were taken from the inner arc, outer arc and the mid-thickness at the distance of 8 mm from the bending axis. Fifty-micrometer-thick TEM foils were prepared by mechanical grinding, followed by twin-jet electrolytic polishing in 10% perchloric acid at 20 V. The microstructure along the thickness direction were observed by SEM and OM, corroded in solutions of 4% nitric acid and saturated picric acid, respectively. After electropolishing and argon-ion polishing, the samples were analyzed by electron backscatter diffraction (EBSD) on a field emission SEM (FESEM, Ultra 55) with an accelerating voltage of 20 kV. Areas of 120 μm × 160 μm were mapped onto a square array with a step size of 0.2 μm. The EBSD data was processed by HKL Channel 5 and ATEX software to obtain quantitative information of the grain-boundary characteristics, grain orientation, and geometrically necessary dislocation densities in NM450 steel.

![Figure 1. Locations of tensile samples of NM450 steel.](image)
Figure 2. Hardness curves and tensile properties of NM450 steel: (a) hardness curves in thickness direction at different positions from bend axis, (b) engineering stress-strain curves in the thickness direction samples at the distance of 8 mm and 147 mm from bending axis, (c) (d) tensile strength and elongation in the thickness direction samples at the distance of 8 mm and 147 mm from the bending axis, respectively.

Figure 3. OM, SEM and TEM graphs at the distance of 8 mm from bending axis: (a) (d) (g) near the outer arc surface, (b) (e) (h) the transition region near the mid-thickness, (c) (f) (i) near the inner arc surface.
3. Results

3.1. Tensile properties
Figure 2 presents the hardness, strength and elongation of the samples at the distance of 8 and 147 mm from the bending axis. The compressive and tensile stress was maximized along the inner and outer arcs (locations A and B, respectively). The tensile strength and hardness of the specimens increase gradually with decreasing the distance from the bending axis. Therefore, the hardness of sample at the distance of 8 mm from the bending axis is higher than that of 147 mm from the bending axis. After bending, the degree of deformation varies along the distance from the bending axis. Therefore, the hardness of sample at the distance of 8 mm from the bending axis, the tensile strength in the outer arc, inner arc and mid-thickness of the samples was 1750, 1640, and 1580 MPa, respectively. At the distance of 147 mm from the bending axis, these values reduced to 1530, 1510, and 1540 MPa, respectively, 220, 130, and 40 MPa lower than those at 8 mm, respectively (figure 2(c)). It is clearly observed that the hardness and tensile strength increase with increasing the amount of plastic deformation which is in agreement with the previous study [14]. Meanwhile, at the distance of 8 mm from the bending axis, the elongation in the outer arc, inner arc and mid-thickness of the samples is 17.98%, 17.62%, 18.19%, respectively. While at the distance of 147 mm from the bending axis, these values increase to 20.17%, 19.84%, 18.47%, respectively. The elongation of the samples along the inner and outer arcs is higher at the distance of 147 mm than that at the distance of 8 mm. But, the elongation along the mid-thickness changes slightly with the distance from the bending axis (figure 2(d)). The increased hardness and tensile strength and decreased elongation in the inner and outer arcs can be attributed to the work hardening behavior due to plastic deformation. Nearby the mid-thickness, the hardness and tensile strength change slightly during bending. If there are large and brittle inclusions in the surface and subsurface layers of the outer arc and inner arc of the bending axis, it is easy to form the bending cracks and thus reduce the bending property of steels.

3.2. Microstructure
3.2.1. Microstructural analysis
Figure 3 shows the microstructure along the thickness direction at the distance of 8 mm from the bending axis. Under tensile stress, the grains of the outer arc are rotated and widened, and the martensite laths tend parallel to the surface of the outer arc (figures 3(a), (d)). The microstructure in the transition region of tensile and compressive stress near the mid-thickness experienced a small strain and was slightly deformed (figures 3(b), (e)). The grains of the inner arc were squeezed under compressive stress, causing the rotation of the martensite laths along the direction perpendicular to the surface of the inner arc (figures 3(c), (f)). According to previous study, the martensite laths were bent irregularly after severe deformation [15]. Further TEM observation revealed an obvious bending deformation of the martensite laths near the inner arc and outer arc, where the dislocation density was higher in the martensite laths than that in the transition region near the mid-thickness (figures 3(g)–(i)). No significant deformation of the martensite laths was observed near the mid-thickness.

3.2.2. Distribution and analysis of TiN inclusions
Figure 4 shows the morphologies, sizes and distributions of the TiN inclusions at the distance of 147 mm from the bending axis. Under OM, the TiN inclusions appeared as the orange–yellow bodies with regular shapes. The TiN inclusions include two types: homogeneous nucleated TiN inclusions (figure 4(a)) and heterogeneous nucleated TiN inclusions with heterogeneous oxide cores (figure 4(b)). The maximum, minimum and average
sizes of the TiN inclusions were 14.42 μm, 1.17 μm, and 4.27 μm, respectively, and about the fraction of 28% of the inclusions were larger than 5 μm (figure 4(c)). Previous studies [16–18] demonstrated the relationship between large-sized TiN inclusions and cleavage fracture; when the proportion of TiN exceeding 5 μm is greater than 19%, the impact energy is remarkably reduced. And, the impact energy is affected not only by the size and proportion of the TiN inclusions, but also by the size and non-homogeneity of the prior austenite grain sizes and the spacing between the TiN inclusions.

The TiN inclusions in the thickness direction at the distance of 8 mm from the bending axis were observed by SEM. As shown in figure 5, the TiN inclusions were broken near the outer arc and inner arc surface but intact in the transition region near the mid-thickness. TiN inclusions are brittle with an elasticity modulus of 600 GPa, much higher than that in the matrix (220 GPa). Therefore, the TiN inclusions were not easily deformed during the bending process and tended to occur stress concentration at the sharp corners, facilitating the initiation and propagation of the cracks [19].

Figure 5 shows the SEM images of the homogeneous and heterogeneous nucleated TiN inclusions. EDS analysis confirmed the presence of MgO or MgAl2O4 cores in the heterogeneous nucleated TiN inclusions (figures 5(g), (h)). As MgO and MgAl2O4 have lower hardness than TiN, they were broken earlier than TiN under stress. Oxide inclusions such as MgO and MgAl2O4 are formed before TiN in the steelmaking process, which promote the nucleation of TiN [20–22]. A small disregistry between the nucleating agent and the nucleated phase indicates a low interfacial energy of nucleation. According to previous studies, inclusion phases are easily nucleated when the disregistry is less than 6.0% [23]. The disregistry values between TiN and the MgO and MgAl2O4 inclusions were 0.053% and 4.88%, respectively [24], so the oxide core could effectively induce the heterogeneous nucleation of TiN. Consequently, the TiN inclusions with oxide cores were slightly larger than the homogeneous nucleated TiN inclusions, which exerted more deteriorative effect on the mechanical properties, especially on the impact toughness of the sample.

Along the outer arc of the bending sample, which was subjected to tensile stress during the bending process, microcracks perpendicular to the tensile-stress direction were formed in the TiN inclusions. As the tensile stress increased, the number and width of the microcracks increased, resulting in the breakage of the TiN inclusions (figure 5(a)). For TiN inclusions with heterogeneous nucleation, the hardness of the oxide core is low, therefore microcracks occur at the interface between the TiN and its oxide core [21]. Under sufficiently high tensile stress,
The heterogeneous nucleated TiN inclusions were broken into two parts (figure 5(d)). The TiN inclusions in the inner arc were subjected to compressive stress, which generated divergent microcracks starting from the inside of TiN. These microcracks stopped at the TiN-matrix interface (figures 5(b), (e)). For heterogeneous nucleated TiN inclusion, the oxide heterogeneous core was crushed by the compressive stress, and several divergent microcracks were initiated at the interface between the TiN and its oxide core (figure 5(e)). As seen in figures 5(c) and (f), the TiN inclusions in the transition region near the mid-thickness, where the strain was reduced, retained their regular shape without fracture. Microcracks in the TiN inclusions did not propagate to the TiN-matrix interface, indicating that the martensite laths prevented the crack propagation \[12\].

Figure 6 shows the microsized TiN fracture mechanism of NM450 steel during bending. The fracture mechanism of TiN inclusion during bending can be divided in four types: (i) under tensile stress, many

![Figure 6. Schematics of the microcrack growth behavior of microsized TiN when subjected to stress: (a) homogeneous nucleating TiN with tensile stress, (b) heterogeneous nucleating TiN with tensile stress, (c) homogeneous nucleation TiN with compressive stress, (d) heterogeneous nucleation TiN with compressive stress.](image)

![Figure 7. Grain boundary misorientation maps at the distance of 8 mm from bending axis: (a) the outer arc, (b) the transition region near the mid-thickness, (c) the inner arc, (d) distribution of proportions of the LAGBs and HAGBs (θ means the angle of boundary, green line 2° ≤ θ ≤ 15°, red line θ > 15°).](image)
microcracks were successively formed on the homogeneous nucleated TiN inclusions in the outer arc near the bend axis (figure 6(a)); (ii) under tensile stress, microcracks occurred at the interface between the TiN and its heterogeneous core, which expanded to the matrix-inclusion interface and finally widened to form cavities (figure 6(b)); (iii) under compressive stress, several divergent microcracks initiated from the centers of the homogeneous nucleated TiN inclusions in the inner arc near the bending axis, which spread to the surrounding matrix-inclusion interface (figure 6(c)); (iv) under compressive stress, divergent microcracks originated from the interface between the TiN and its heterogeneous core in the inner arc near the bending axis, which expanded to the matrix-inclusion interface (figure 6(d)).

3.2.3. EBSD analysis
Figure 7 shows the distributions of grain boundary angles in the samples. High angle grain boundaries (HAGBs) with misorientation angle of \( \theta > 15^\circ \) was indicated by red lines, while the black lines corresponded to low angle grain boundaries (LAGBs) with a misorientation angle of \( 2^\circ \leq \theta \leq 15^\circ \). The LAGB is mainly composed of dislocations. As seen in figure 7, the transition region in the mid–thickness sample contained a larger proportion of HAGBs and a lower proportion of LAGBs than the inner-arc and outer-arc samples (HAGBs: 53.2% in the mid-thickness versus 33.8% and 32.0% in the inner arc and outer arc, respectively; LAGBs: 46.8% in the mid-thickness versus, 66.2% and 68.0% in the inner arc and outer arc, respectively). In summary, the proportion of LAGBs increased and the proportion of HAGBs decreased with increasing the amount of plastic deformation (figure 7(d)). At the start of the bending deformation, the grains began to rotate via dislocation slip, and the dislocation density increased accordingly. With the bending deformation, the HAGBs gradually transformed into the LAGBs [25], the proportion of LAGB was higher in the inner and outer arcs location (which underwent serious deformation) than that in the mid-thickness location (which was only slightly deformed).

In recent years, geometrically necessary dislocation (GND) has been used to quantify the local strain accumulation in metals during plastic deformation. Ma et al [26] found that the geometrically necessary dislocation density (\( \rho^{\text{GND}} \)) increased significantly with increasing plastic strain. The relationship between the \( \rho^{\text{GND}} \) and the local misorientation is shown as follows [27, 28].

\[
\rho^{\text{GND}} = \frac{2\bar{\theta}}{ub}
\]

where \( \bar{\theta} \) represents the average value of local misorientation, \( \nu \) is the step length during EBSD scanning (200 nm), and \( b \) is Burgers vector (0.248 nm). The distributions of \( \rho^{\text{GND}} \) in the outer-arc, mid-thickness and inner-arc samples are shown in figures 8(a)–(c), respectively. The color graduation from blue to red indicates the low to high range of \( \rho^{\text{GND}} \) values. Obviously, the red area is smaller in the mid-thickness sample than in the inner-arc and outer-arc samples. As shown in figure 8(d), the \( \rho^{\text{GND}} \) in the mid-thickness, inner-arc and outer-arc samples peaked at \( 0.91 \times 10^{14} \text{ m}^{-2}, 8.09 \times 10^{14} \text{ m}^{-2}, \) and \( 16.68 \times 10^{14} \text{ m}^{-2} \), respectively. The average \( \rho^{\text{GND}} \) in the mid-thickness, inner-arc and outer-arc samples was \( 9.39 \times 10^{14} \text{ m}^{-2}, 11.85 \times 10^{14} \text{ m}^{-2} \) and...
13.32 × 10^{14} \text{ m}^{-2}$, respectively. Thus, the $\rho_{\text{GND}}$ of the samples increased remarkably as the deformation increased from the mid-thickness to the inner arc and outer arc. Because the GND mainly originated from the LAGBs of the deformed microstructures, the inner-arc and outer-arc samples with more LAGBs than the mid-thickness sample exhibited higher $\rho_{\text{GND}}$ and a larger dislocation strengthening effect than the mid-thickness sample. In a word, the higher strength and hardness of the outer-arc and inner-arc samples than the mid-thickness sample can be attributed to their relatively higher $\rho_{\text{GND}}$ values.

3.2.4. Fracture surface

Figure 9 shows the fracture surface of samples at the distance of 8 mm from the bending axis after the tensile tests. The size and depth of the dimples are related to the plastic deformation characteristics of the matrix. Large and deep dimples indicate high ductility of the steel. The fracture surface of the outer-arc, mid-thickness and inner-arc samples was characterized by dimples of different sizes and a few quasi-cleavage planes, tear ridges, and secondary cracks in different directions, as shown in figures 9(a)–(c), respectively. The fracture surface of mid-thickness sample exhibited the dimples, and many large dimples were surrounded by dense fine dimples (figure 9(b)), indicating high ductility. However, the fracture surface of the inner- and outer-arc samples presented many tear ridges and quasi-cleavage planes besides dimples, which reveals the mixed ductile and brittle fracture (figures 9(a), (c)). TiN inclusion with a MgO core was found on the quasi-cleavage plane in the fiber zone, indicating that the microsized TiN inclusion in the steel initiated cracks during the tensile process (figures 9(d)–(f)). At higher magnification, microcracks were found on the TiN inclusion.

4. Conclusions

The main conclusions of the study are summarized below.

(1) The inner and outer arcs exhibited an obvious bend deformation of the martensite laths with a high dislocation density. At the distance of 8 mm from the bending axis, the tensile strength was higher in the inner and outer arcs than that in the mid-thickness region. Whereas, at the distance of 147 mm from the bend axis, the tensile strength changes slightly in thickness direction. But the tensile strength of the outer arc, inner arc and mid-thickness samples at the distance of 147 mm from the bending axis was 220, 130, and 40 MPa lower, respectively, than that at the distance of 8 mm.

(2) The TiN fracture mechanism was found to depend on the type of applied stress and whether or not to have heterogeneous nucleation core of the TiN inclusions. During the bending process, the microsized TiN inclusions nearby the outer arc were subjected to tensile stress and multiple microcracks/cavities perpendicular to the outer arc were formed. However, the microsized TiN inclusions nearby the inner arc underwent compressive stress, initiating the occurrence of divergent microcracks.
The GND increased from the mid-thickness to the inner and outer arcs of the bending samples, along with the dislocation strengthening effect. Accordingly, the strength and hardness were higher in the outer- and inner-arc samples than in the mid-thickness sample.

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Data availability statement

All data that support the findings of this study are included within the article (and any supplementary files).

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