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A study of properties of ultrafine bainitic steel under low austempering temperature

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Abstract:

The mechanical properties and wear resistance of the ultra-fine bainitic steel austempered at various temperatures were investigated. Scanning electron microscopy (SEM) and X-ray diffraction were used to analyze the microstructure. The worn surfaces were observed via laser scanning confocal microscopy and SEM. Results indicated that, under low austempering temperatures, the mechanical properties differed, and the wear resistance remained basically unchanged. The tensile strength of the samples was above 1800 MPa, but only one sample austempered at 230°C had an elongation of more than 10%. The sample weight loss was approximately linear with the cycles of wear and nonlinear with the loads. The samples showed little difference in wear resistance at different isothermal temperatures, whereas the thickness of their deformed layers varied greatly. The results are related to the initial hardness of the sample and the stability of the retained austenite. Meanwhile, the experimental results show that the effect of austempering temperatures on the wear resistance of ultra-fine bainitic steel can be neglected under low applied loads and low austempering temperature.

Keywords:

ultra-fine bainitic steel; austempering temperature; mechanical properties; wear resistance

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1 Introduction

Ultrafine bainitic steel, developed by Bhadeshia and their co-workers [1], has a multiphase structure comprising extremely fine bainitic ferrite and carbon-rich retained austenite formed by austempering at 125°C–350 °C [1-4]. The former provides ultra-high strength (1.2−2.6GPa)[5-8], and the latter offers sufficient toughness for ultrafine bainitic steels (30−50MPa • $m^{1/2}$) [8-10]. This combination of strength and toughness has attracted the attention of scholars.

In addition to its mechanical properties, the wear resistance of ultrafine bainitic steels is also of significant interest. In recent years, extensive research has been conducted on the wear resistance of ultrafine bainitic steels, such as two-body abrasive wear [11,12], three-body abrasive wear [13,14], dry sliding wear [15], rolling/sliding wear [16,17], repeated frictional sliding [18,19], etc. Zhang et al. [20] studied the wear resistance of carbide-free bainitic steels, finding that carbide-free bainite exhibited higher wear resistance under high wear loading because of the increased hardness of the sample surface by the strain-induced martensitic transformation from retained austenite. Bakshi et al. [13] studied the wear resistance of three microstructures under a three-body abrasive. They found that nanostructured bainite was more resistant to abrasion than the other two types of microstructure, including fine pearlite and martensite. This is, perhaps, because of austenitization and the martensite transformation of the surface layer during abrasion. Hasan et al. [16] found that carbide-free bainitic rail steel had a higher wear resistance than pearlitic rail steel and that the wear resistance increased with increasing retained austenite content and decreasing bainitic ferrite lath thickness. The ultrafine bainitic structure, which is formed at low austempering temperatures, has, therefore, displayed satisfactory wear resistance.

Prior studies [11,19] have shown that, the lower the austempering temperature, the better the wear resistance of the ultra-fine bainitic steel. Because this finding was obtained at a higher temperature gradient (Ms+50°C to Ms+150°C, Ms: martensite start temperature), it is unclear whether the wear resistance of the ultra-fine bainite obtained under a lower temperature gradient (Ms to Ms+50°C) would also display these

properties. Therefore, a new composition of high-carbon, low-alloy ultra-fine bainitic steel was designed and austempered between Ms+5°C and Ms+50°C for this study. Moreover, the wear resistance and wear mechanism of the new bainite steel under two-body wear were studied. Additionally, the effect of austempering temperatures and applied loads on the wear resistance was investigated.

2 Experimental section

The chemical composition of the experimental steel used in this paper was 0.70C-2.47Si-1.18Mn-0.87Al (wt. %). The steel was melted into an ingot of 20 kg by a vacuum medium frequency induction melting furnace. After forged, the steel billets

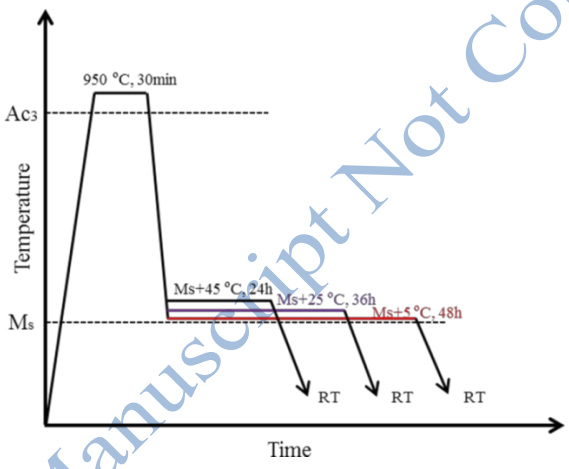

Fig. 1. Schematic diagram of austempering treatments.

were homogenized at 1200°C for 2 h; subsequently, the steel billets were rolled into a thickness of 6 mm, and finally cooled to ambient air temperature. Cylindrical samples (φ4×10 mm) were applied to measure the austenite formation temperature (Ac1 and Ac3) and Ms temperature. The austenite formation temperature (Ac1 and Ac3) and Ms temperature were measured using a cylindrical sample (φ4 × 10 mm). The corresponding temperatures were 782°C, 853°C and, 185°C, respectively. The rolled steels were austenitized at 950^{o}C for 30 min, followed by cooling to 190^{o}C, 210^{o}C, and 230^{o}C for 48 h, 36 h, and 24 h respectively, as shown in Fig. 1.

A ZEISS ULTRA 55-type field emission scanning electron microscope (FE-SEM), operating at 20 kV, was used to observe the microstructural morphology and phase

distribution of the steel specimens. X-ray diffraction (XRD) was used to measure the volume fraction of the retained austenite (RA). The instrument was operated at 40 kV and a current of 150 mA with Cu-Kα radiation. The 2θ scanning angles were varied from 45° to 95° with a stepping angle of 0.05° and a counting time of 1.2 s per step using 10 × 10 mm specimens. The RA content was measured by measuring the integrated intensities of the $(200)_\gamma$, $(220)_\gamma$ and $(311)_\gamma$ peaks, and with $(200)_\alpha$, $(211)_\alpha$ [21]. The film-like RA content could be estimated using the following Eq. (1) [22]:

$$\frac{V_r^f}{V_r^b} = \frac{0.15\ V_{BF}}{V_r - 0.15 V_{BF}} \quad (1)$$

where V_r^f, V_r^b, V_{BF}, and Vr represent the volume fraction of film-like retained austenite, blocky retained austenite, bainitic ferrite, and retained austenite, respectively.

Tensile properties were tested using an MTS 810 electronic universal testing machine with a crosshead speed of 1.0 mm•min^{-1} at room temperature. Samples were prepared with a gauge length of 25 mm and width of 6 mm. The true stress-strain curves were converted based on the engineering stress-strain curves, and strain hardening was characterized by the instantaneous work-hardening coefficient (n), which was calculated from the true stress-strain curve as shown in Eq. (2):

$$n = \frac{d \ln \sigma}{d \ln \varepsilon} \quad (2)$$

where σ and ε are the true stress and true strain, respectively.

The equipment used in the two-body abrasion testing was an ML-100 abrasive wear testing machine. The rotation rate of the disk and a constant feed of samples were 60 rpm and 3 mm/rev, respectively, as shown in Fig. 2. Applied loads used in the wear test were 5 N, 10 N, and 15 N. Cylindrical samples (φ3 × 10 mm) cut from the austempered samples were used to wear the experiments. The medium used for the abrasive wear test was 200 grit SiC paper (particle size: 75 μm). The wear test was stopped every 200 cycles, and weight loss of the samples was recorded using an analytical balance (accuracy of 10^{-4} g). The total number of wear cycles was 600 rotations. Worn samples under each test condition were tested at least five times.

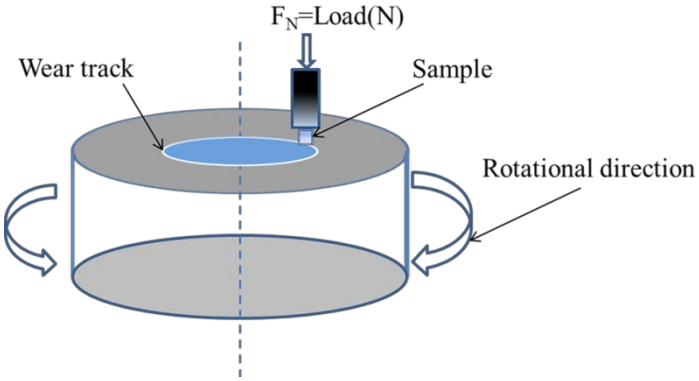

Fig. 2. Schematic diagram of wear tests.

3 Results and discussion

3.1 Microstructure characterization

Fig. 3 shows the SEM images of the microstructures under different isothermal temperatures. The microstructure for all heat treatments consisted of bainitic ferrite (BF) and RA. Precipitation of carbide was not observed in the microstructure because of the addition of Si [23]. According to the formula [24], $L_T=\pi t/2$, L_T is the mean linear intercept normal to the length of the plates, and t is the true thickness of the plates. The mean plate thicknesses (t) are 49.3 ± 2.5, 58.2 ± 3.3, and 63.4 ± 3.3 nm, austempered at 190°C, 210°C, and 230°C, respectively (defined as 190-sample, 210-sample, and 230-sample). Untransformed austenite presented as blocky (prior austenite boundaries and between bainitic sheaves) and film-like (between BF plates) structures at ambient-temperature. The film-like retained austenite (FRA) was of great benefit when improving the toughness of the ultra-fine bainitic steel. Compared with FRA, the blocky retained austenite (BRA) was more inclined to transform into brittle martensite, which reduced the toughness of the ultra-fine bainite steels under residual stresses or external forces [25]. The BRA dimension of the 190-sample was smaller than that of the other two samples, as shown in Fig. 3.

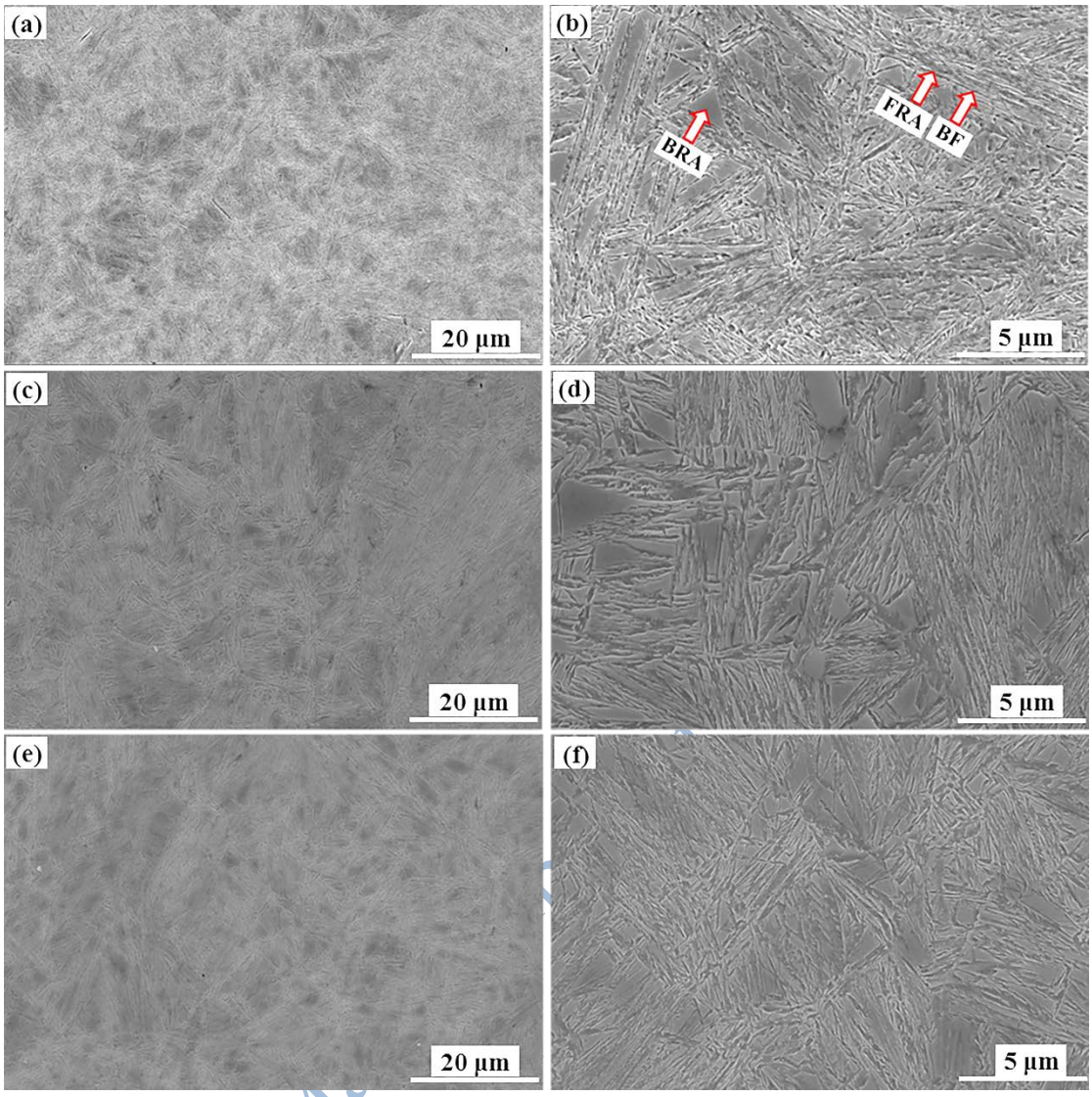

Fig. 3. SEM image at: (a, b) 190 ^{o}C, (c, d) 210 ^{o}C, and (e, f) 230 ^{o}C.

3.2 Mechanical properties

Fig. 4(a) represents the engineering stress-strain, and the measured tensile properties are enumerated in Table 1. The 230-sample displayed the lowest yield ($\sigma_{0.2}$) and tensile (σ_b) strength (1338 MPa and 1846 MPa, respectively) and the highest elongation (10.93%). The yield and tensile strength increased with decreasing isothermal temperature, except for the tensile strength of the 210-sample, whereas the total elongation was the lowest in the 190-sample. Fig. 4(b) shows the true-strain and hardening index (n) curves. The 190-sample and 210-sample broke suddenly at low

strain; the hardening index at that point was higher than for the 230-sample. Necking occurred in the 230-sample before fracture.

It is well-known that nanoscale BF provides high strength for ultra-fine bainitic steel. The strength is primarily controlled by the bainitic plate thickness (i.e. $\sigma=115 / (2t)$[26]). As the isothermal temperature decreases, the BF becomes thinner. BF thicknesses in the 190 and 230 samples were 49.3 ± 2.5 and 63.4 ± 3.3 nm, respectively. Dislocation density in the microstructure was also a factor affecting strength. After paraequilibrium nucleation, the ferrite plate grew through a shearing mechanism. During this process, austenite produced high-density dislocations to accommodate strain. The lower the bainite transformation temperature, the higher the strength of the parent austenite[27]. Dislocation density in the microstructure of the 190-sample was higher than that in the 230-sample. In summary, the ultimate tensile strength (UTS) of the 190- and 210-samples was higher than that of the 230-sample.

It is known that the amount, morphology, and distribution of RA are significant factors influencing the mechanical properties of steel. The volume fractions of RA were measured by XRD before and after the tensile test, as shown in Fig. 5. The RA contents of the three samples before and after stretching were > 20 and 6~7vol%, respectively. It can be seen that the austenite to martensite transformation (TRIP effect) occurred in all samples during the tensile test. The question then arose as to why the elongations of the 190 and 210 specimens were so low. This was attributed to the stability of the RA. The dimension and morphology of the RA had an important influence on the properties of the experimental steel. First, as the austempering temperature decreased, the thickness of the RF and FRA decreased. The possibility of FRA producing a TRIP effect during the tension was reduced. Second, large amounts of BRA were transformed into martensite at low strain [28]. Third, BRA was liable to crack. These reasons could explain why the UTS of the 210-sample sample was higher than that of the 190-sample.

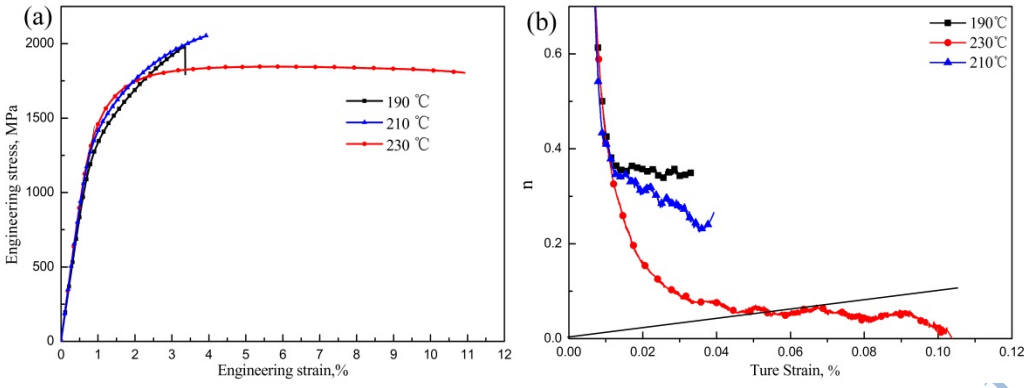

Fig. 4. Tensile properties and hardening index curves: (a) engineering stress-strain curves, and (b) true-strain and n curves of three samples in ambient temperature.

Table 1. Mechanical properties of the three samples.

sample	YS($\sigma_{0.2}$) / MPa	UTS(σ_b) / MPa	TEL / %	Hardness(HV1.0)
190-sample	1523±15	1985±15	3.4±0.17	673±20
210-sample	1430±13	2053±13	3.9±0.19	645±21
230-sample	1338±12	1846±12	10.9±0.6	593±17

YS, UTS and TEL stand for yield strength, ultimate tensile strength and total elongation, respectively.

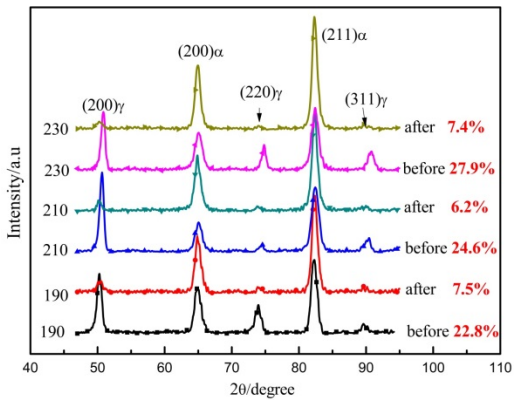

Fig. 5. X-ray diffraction (XRD) patterns of the three samples before and after tensile experiments.

Typical SEM images of the tensile fracture surface of the 190-sample and 230-sample are shown in Fig. 6. The macroscopic fracture morphology of the samples was a river-like pattern, as shown in Fig. 6(a) and 6(b). The fracture surface was composed of a number of cleavage facets and some tearing edges, such as in Fig. 6(c), (d), and (e). There was a small necking zone on the fracture surface of the 230-sample, as shown in Fig. 6(b), which was not found in the 190-sample. The phenomenon was consistent with the work-hardening exponential curve (Fig. 4(b)). There were many dimples in the necking zone, as seen in Fig. 6(f). This was another reason why the elongation of the 230-sample was greater than that of the other samples.

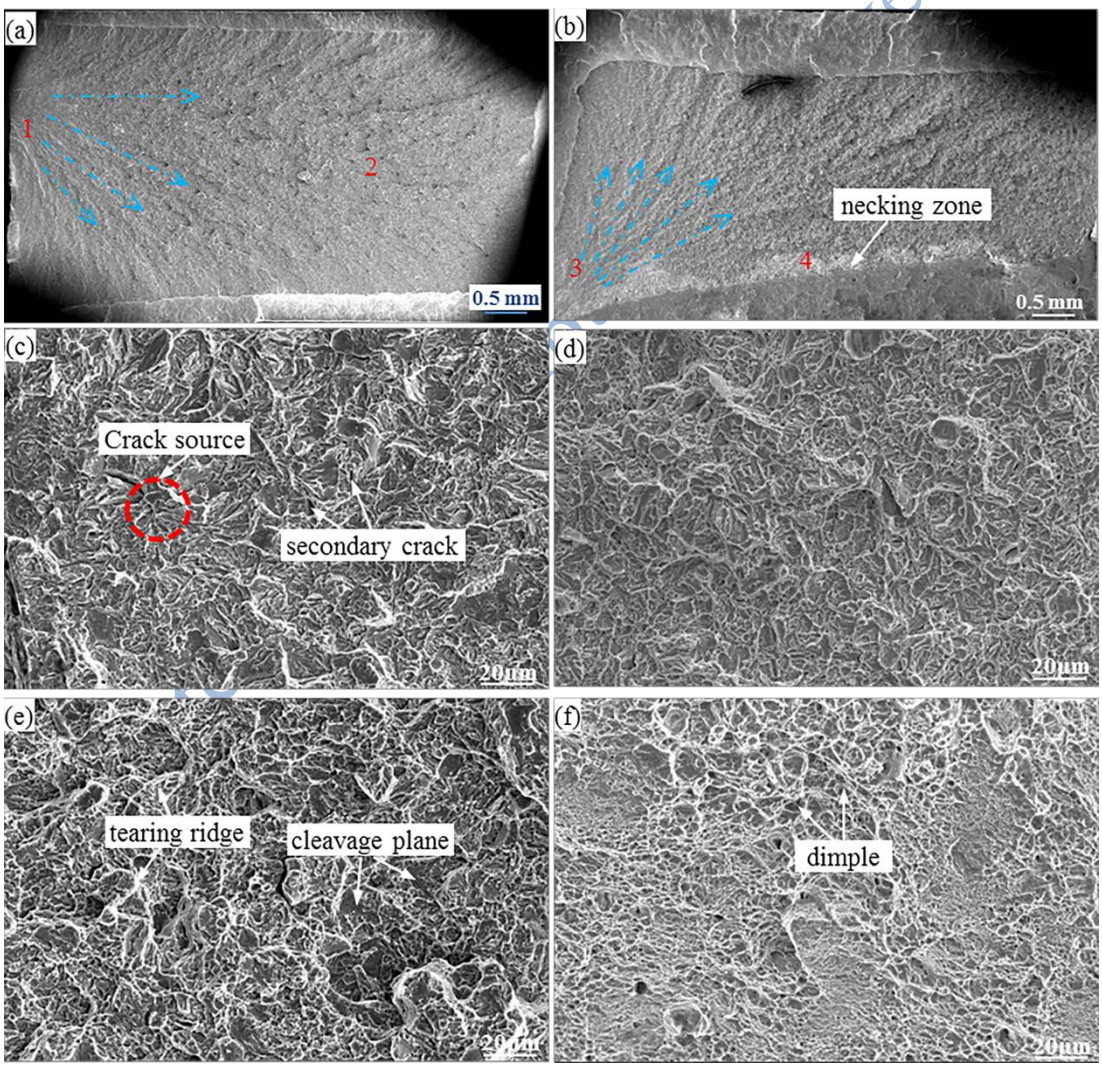

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Fig. 6. Typical SEM micrograph of tensile fractures of the sample at (a) 190°C, and (b) 230°C; (c) enlargement of 1 in (a); (e) enlargement of 2 in (a); (d) enlargement of 3 in (b); (f) enlargement of 4 in (b).

3.3 Wear resistance

The experimental results of the abrasive wear test are shown in Fig. 7. Under any of the experimental parameters of this study, the wear loss of the 190-sample was lower than that of the 210- and 230-samples. The wear losses of the 210- and 230-samples were essentially the same under low applied loads and low wear cycles. As the number of wear cycles or the applied load increased, the wear loss of the 210-sample was lower than the wear loss weight of the 230-sample. Weight loss of the samples was approximately linear with the cycles of wear, whereas the weight loss of the tested samples increased nonlinearly with increasing applied load under the same number of cycles, as shown in Fig. 7. Wear resistance of the tested samples increased with decreasing isothermal temperature. When the load was 5 N, the weight loss every 200 cycles (Δ) was almost unchanged under the same austempering temperature. Weight loss increased with increasing load under 200 cycles, but the Δ of the investigated samples decreased with increasing applied load.

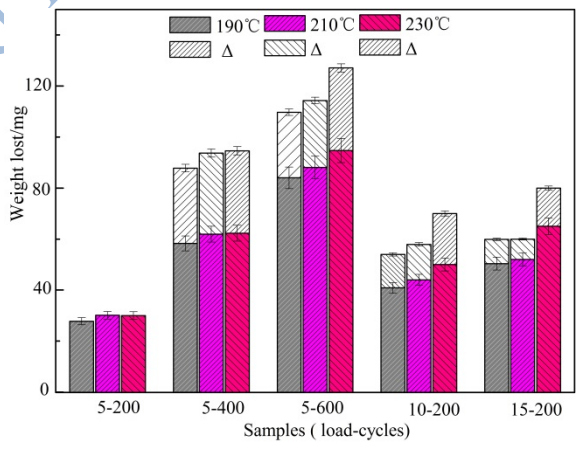

Fig. 7. Weight loss of the samples under different heat treatments and different wear conditions. (5-200): the former represents applied loads (5 N), the latter represents cycles of wear (200 cycles); 5-400, 5-600, 10-200 and 15-200 are named the same way as 5-200;

Δ represents the weight loss of each 200 cycle)

Fig. 8 and Fig. 9 show the SEM images of the worn surfaces of the experimental samples under different cycles and loads. Lengthwise cutting and continuous grooves were observed on the worn surface in all samples. The grooves were formed by SiC particles cutting the surface of samples. The direction of the grooves fell along the sliding route. The phenomenon indicated that the primary damage mechanism in this study was micro-plowing abrasion. It was noteworthy to observe that these processes endured very few material losses, as shown in Fig. 7. With increasing wear cycles and loads, the width of the grooves in the 190- and 230-samples increased, but the width of the grooves in the 190-sample was narrower than that of the 230-sample. The 230-sample under 200 and 400 cycles under 5 N presented some deep scratches, owing to the relatively low initial hardness (593±17 HV1.0) of the samples, which resulted in some SiC particles being deeply pressed into the surface by the load. Straight grooves were most likely to be observed in other samples, because of the relatively high initial hardness (673±20 HV1.0) of the 190-samples and the more-hardened layer of the 230-samples for 600 cycles, under 10 and 15 N.

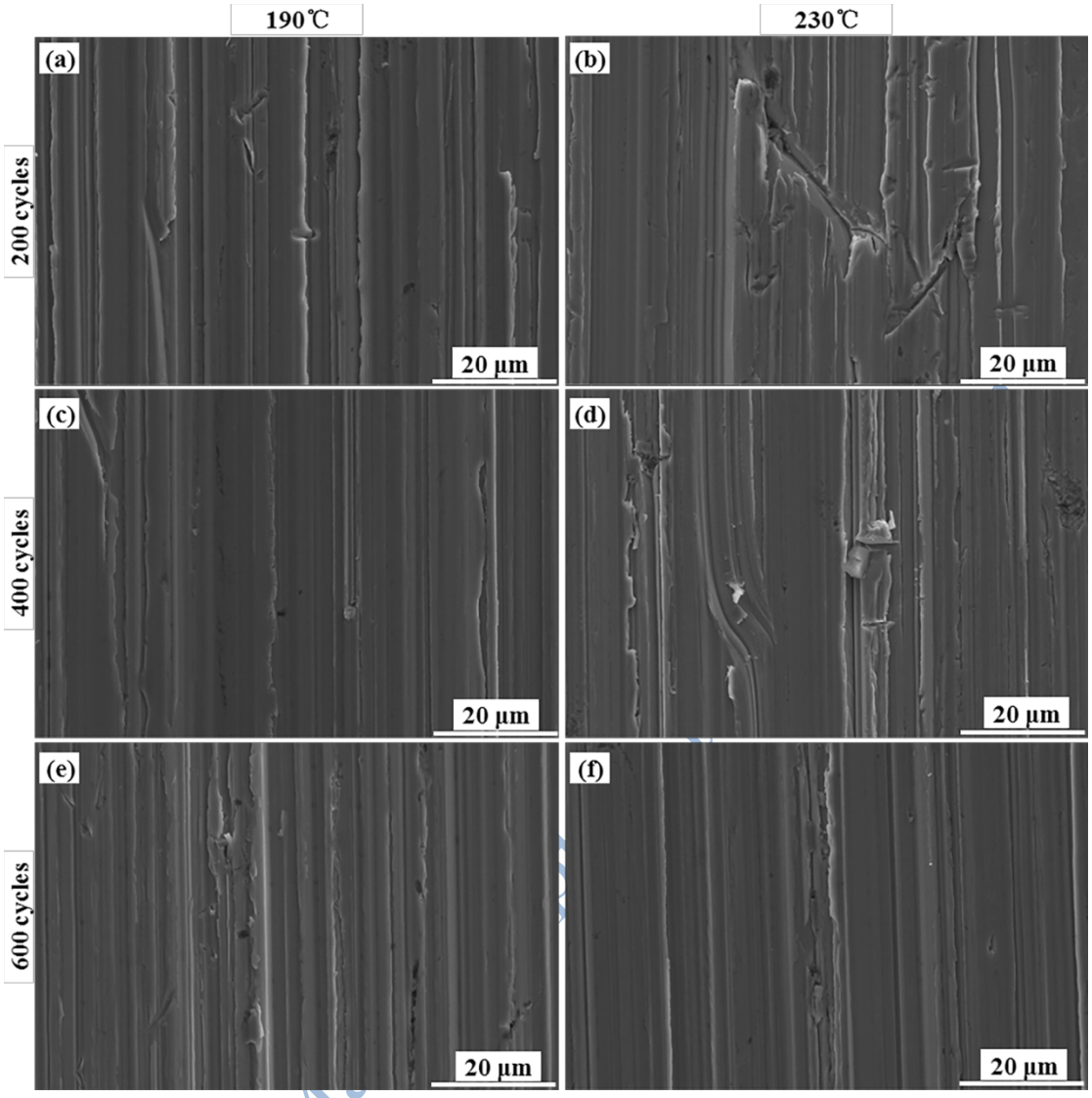

Fig. 8. SEM micrograph of the wear surface of samples for 200 cycles, 400 cycles, and 600 cycles under 5 N.

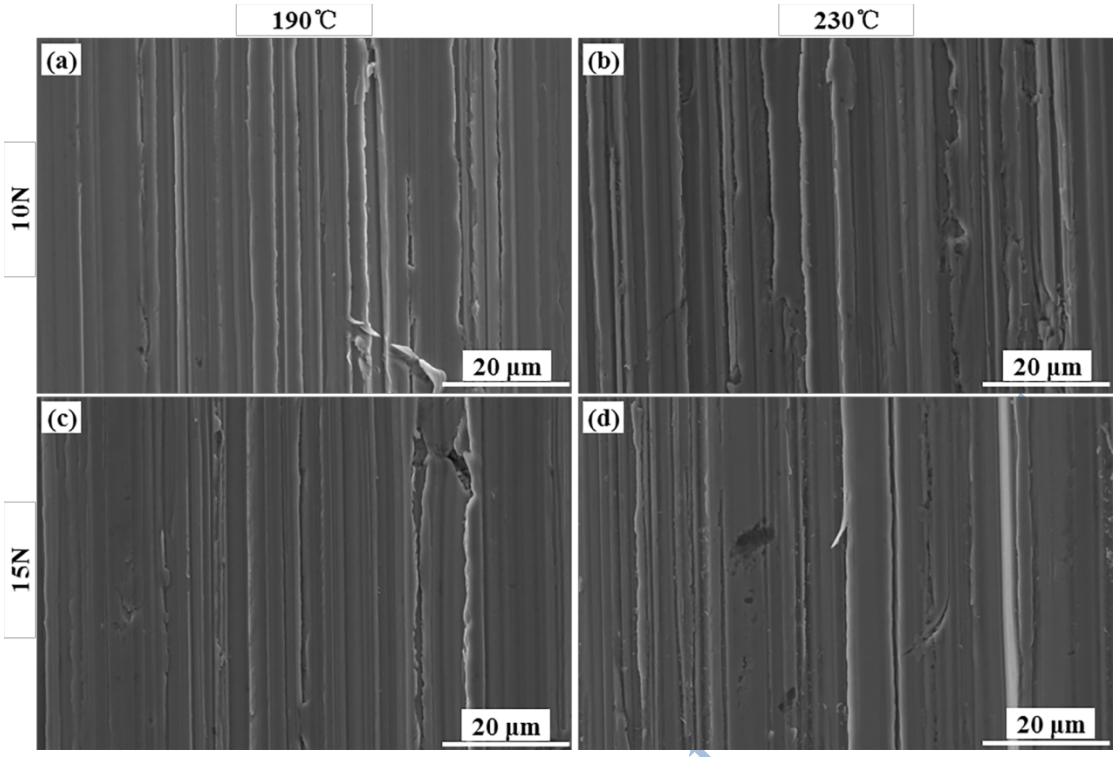

Fig. 9. SEM micrograph of the wear surface of samples for the 10 N and 15 N under 200 cycles.

Fig. 10 and Fig. 11 illustrate the line profiles of the wear surfaces with different samples by laser scanning confocal microscopy. The groove surfaces of the 190-samples having high initial hardness were “narrow and deep.” With increasing wear cycles, the groove surfaces changed from “narrow and deep” to “wide and shallow,” and then back to “narrow and deep.” For samples having lower initial hardness, such as the 230-sample, the groove surfaces were “wide and shallow.” With the increase in wear cycles, the surfaces of the grooves tended to be “narrow and deep.” Similarly, the groove surfaces tended to be “narrow and deep” as the load increased. These results were consistent with those obtained by Narayanaswamy[29].

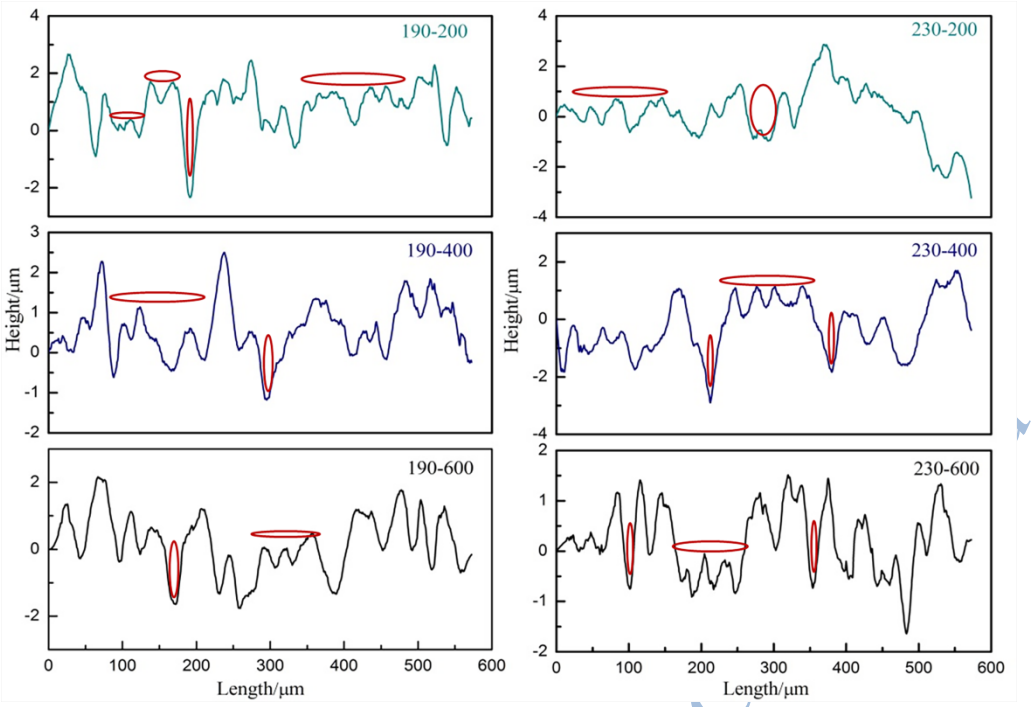

Fig. 10. Profiles of the wear surfaces of samples for 200, 400, and 600 cycles under 5 N.

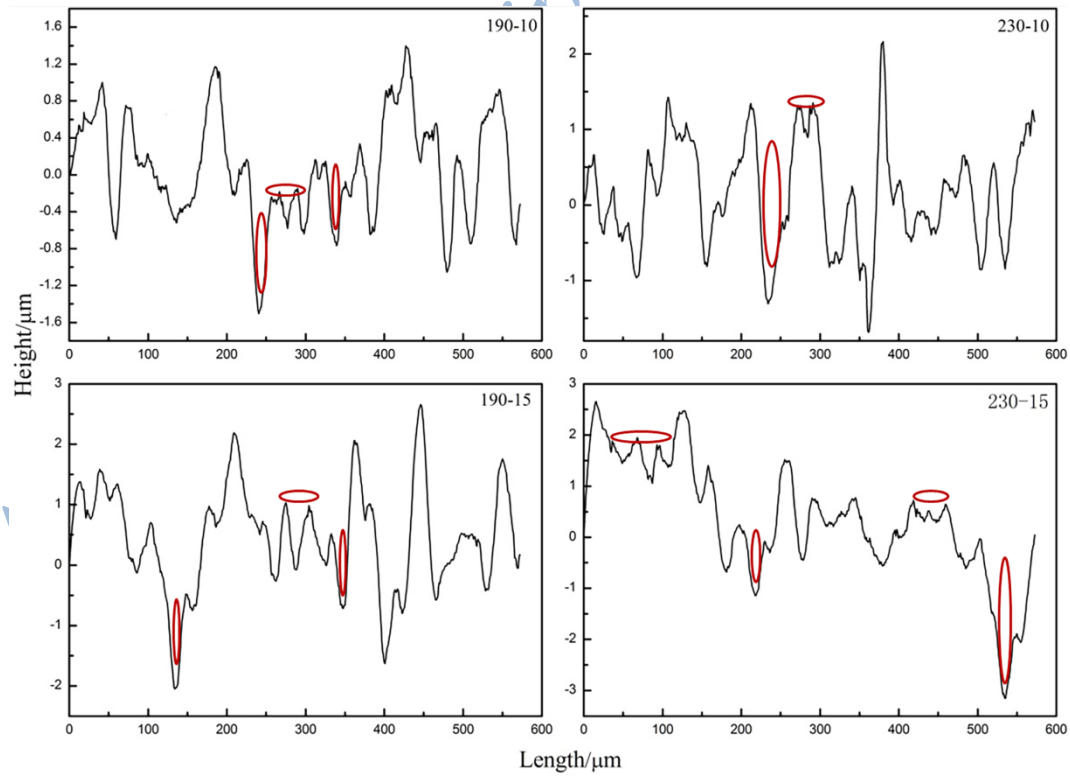

Fig. 11. Profiles of the wear surfaces of samples for the 10 and 15 N under 200 cycles.

Cross-sections of the worn surfaces of some samples under typical processes were observed by SEM, as shown in Fig. 12. A severely deformed layer was clearly seen between the worn surface and the violet dashed line. The deformed microstructures were rearranged and deflected in the sliding direction. Furthermore, in the deformed layer, the FRA became thinner and the BRA became slender. However, BRA was not observed to become FRA, which was inconsistent with the research of Singh[19].

Generally, sliding abrasion involves high strain levels on the abrasive surface, resulting in severe deformation on the sub-surface of the samples during wear[30,31]. However, the thickness of the deformation layer varies. For example, in the 190-sample, the thickness of the deformation layer was close to zero with 200 cycles under 5 N (Fig. 12(a)). Under austempering at 230°C, the thickness of the deformation layer increased from 3 to 5 μm as the wear cycles increased from 200 to 600 (Figure 12(b), (d), and (e)). When the number of wear cycles remained constant, the thickness of the deformation layer of the 190- and 230-samples increased with increasing applied loads (Fig. 12 (a), (g) and (b), (h)).

A non-etching layer beneath the surface was also observed, as shown in Fig. 12 (a)−(h). This layer indicated the onset of wear matching with the location where, in theory, the maximum tangential traction stress occurs[17]. It was impossible to obtain martensitic transformation after reaustenitization at the mating surfaces during abrasion. Therefore, the non-etching layer became a typical reflection of intense, repeated deformation, grain fragmentation, and mechanical homogenization [32].

Compared with the 190-sample, the initial hardness of the 230-sample was lower. The pressure, therefore, propagated more deeply in the 230-sample under the same load, leading to a more significant deformation layer, as shown in Fig. 13. Li et al. [33] found that the effect of RA stability was an essential factor in wear resistance. They reported that the RA stability decreased with increasing austempering temperature. When the mechanical stability of RA was high enough, it could delay the transformation of austenite to martensite. Therefore, compared with the 190-sample, it was easier for RA in the 230-sample to transform into martensite, thereby improving the surface hardness and enhancing the wear resistance.

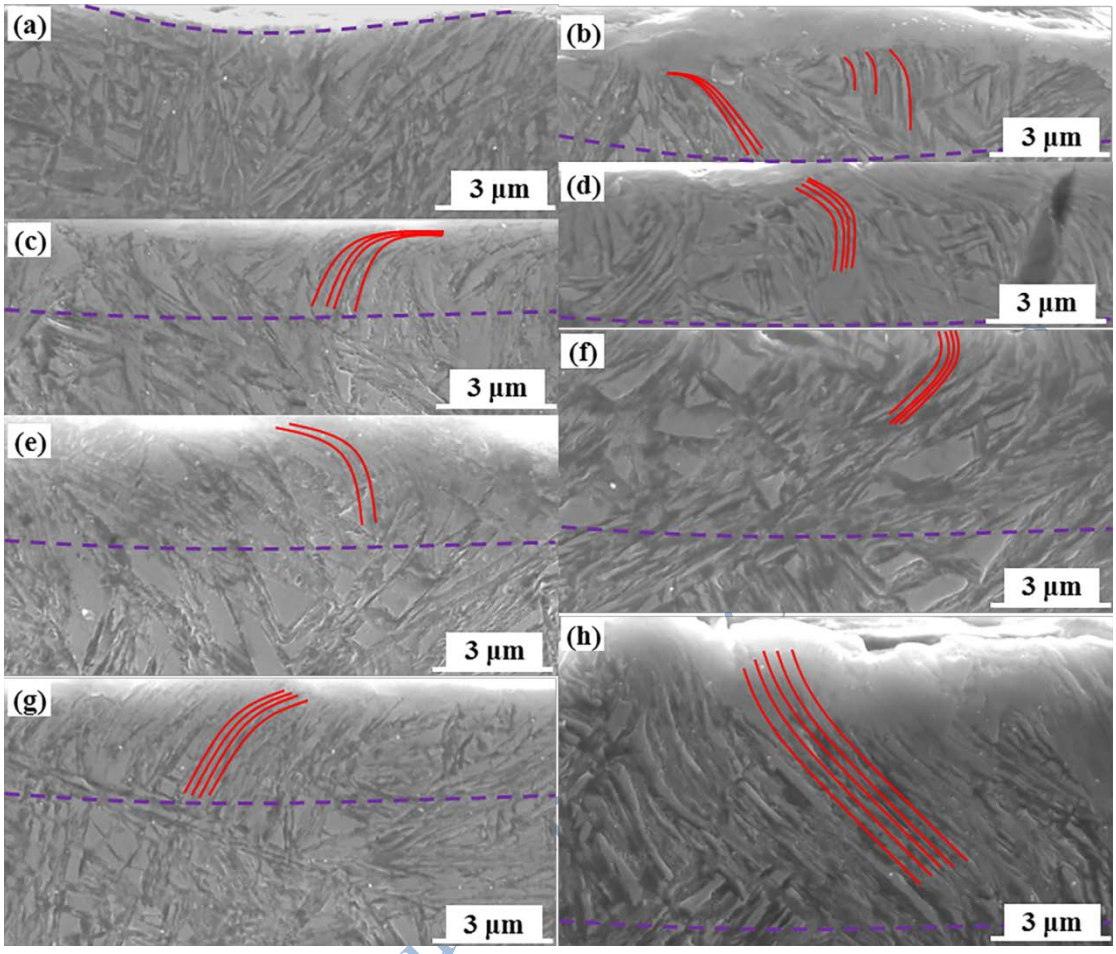

Fig. 12. SEM micrograph of wear sub-surface of (a) 5-200, (c) 5-400, (e) 5-600, (g) 15-200 at 190°C, and (b) 5-200, (d) 5-400, (f) 5-600, and (h) 15-200 at 230°C.

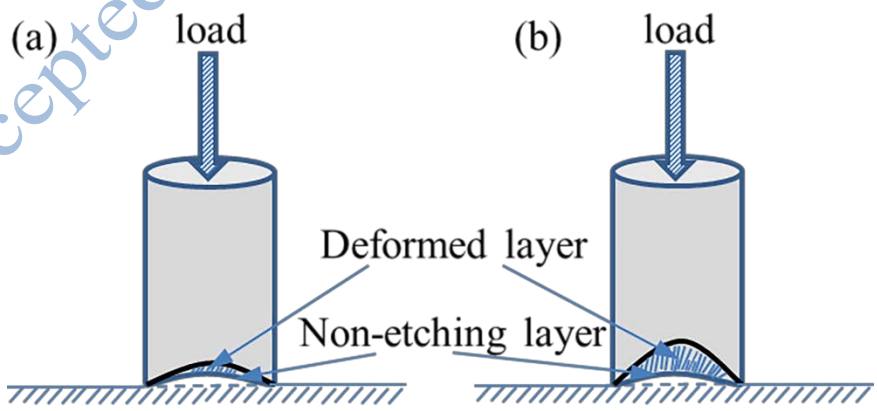

Fig. 13 Schematic diagram of wear mechanism: (a) hard sample, (b) soft sample.

Conclusion

This paper investigated the properties of ultrafine bainitic steel by isothermal transformation at temperatures as low as between Ms +5°C and Ms + 50°C, and the following conclusions can be drawn from the analysis of results:

(i) The tensile strength of samples austempered at 190°C–230°C was better than 1800 MPa, but the only sample austempered at 230°C had an elongation of more than 10%. The samples under different temperatures had high wear resistance because of the high initial hardness (593−673 HV).

(ii) The wear surface in ultrafine bainitic steel mainly comprises grooves, and micro-plowing is the main wear mechanism. As the austempering temperature decreases, the change in wear resistance is not significant. The sample weight loss was approximately linear with the cycles of wear, whereas the sample weight loss was nonlinear with the increase in applied loads.

(iii) The thickness of the deformed layers was, at most, 3 μm in the 190-sample, whereas the thickness of the deformed layers in the 230-sample was approximately 2.5 times that of the 190-sample. This phenomenon is related to the initial hardness of the sample and the stability of the RA.

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