Effect of Austenitizing Temperature on Dry Sliding Friction and Wear Properties of Medium-Carbon Quenching and Partitioning Steel Postprint

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Abstract

For medium-carbon Fe-0.4 C-1.5 Mn-1.5 Si steel, the dry sliding friction and wear properties of quenching-partitioning specimens treated at three austenitization temperatures were investigated, using traditional quenching-tempering specimens as a comparison. The results revealed that the two quenching-partitioning specimens subjected to full austenitization at 860 ?C and 1000 ?C exhibited similar retained austenite contents (~14.37 vol.% and ~13.79 vol.%, respectively), with relatively high carbon concentrations within them (1.37 wt.% and 1.38 wt.%, respectively) and strong mechanical stability. Under constant low-load (50 N) and constant low sliding speed conditions (40 mm/s), martensitic transformation was not readily induced during friction, resulting in very low friction and wear resistance for both specimens. Influenced by microstructural refinement, the specimen with lower austenitization temperature possessed higher wear resistance. When the austenitization temperature was reduced to 800 ?C, a specimen with critical quenching-partitioning was obtained. Microstructural analysis indicated that this specimen not only contained a small amount of ferrite (~6.75 vol.%), but also exhibited the highest retained austenite content (~22.28 vol.%), resulting in the lowest microhardness value among the four groups of specimens. However, due to the low carbon concentration (~1.06 wt.%), the retained austenite possessed weaker mechanical stability, and martensitic transformation was readily induced during friction, which not only contributed additional hardening but also the compressive stresses in the material surface layer caused by the volume expansion of martensitic transformation were beneficial for improving wear resistance, thereby resulting in the critical quenching-partitioning specimen exhibiting the best wear resistance. Therefore, under the given friction parameter conditions, the influence of retained austenite on the wear resistance of martensitic steel is primarily determined by whether additional hardening can be induced through phase transformation during the friction process.

Full Text

Effect of Austenitization Temperature on the Dry Sliding Wear Resistance of a Medium Carbon Quenching and Partitioning Steel

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Abstract

The quenching and partitioning (Q&P) process is a promising method to create novel martensitic steels with an improved balance of strength and ductility by retaining a considerable amount of austenite in the martensitic matrix. This microstructure provides suitable conditions to study wear mechanisms, as the effect of retained austenite on the wear properties of martensitic steel remains controversial. Using a traditional quenching and tempering (Q&T) sample with identical composition (Fe-0.4C-1.5Mn-1.5Si) as a reference, this study investigated the dry sliding wear properties of Q&P samples subjected to different austenitization temperatures.

The results show that the volume fraction of retained austenite in Q&P samples with full austenitization at 860°C or 1000°C is nearly identical (approximately 14.37% and 13.79%, respectively), with correspondingly high carbon concentrations (mass fraction) in the retained austenite (1.37% and 1.38%, respectively). Under low loading (50 N) and sliding speed (40 mm/s) conditions, the high mechanical stability of retained austenite makes it difficult to induce martensitic transformation during friction, leading to low friction and wear resistance in both samples. The slightly better wear resistance in the lower austenitization temperature sample can be attributed to microstructural refinement.

When the austenitization temperature was reduced to 800°C, intercritical Q&P samples were obtained. Microstructural analysis revealed the highest volume fraction of retained austenite (approximately 22.28%) along with a small volume fraction of ferrite (approximately 6.75%) in the martensitic matrix, resulting in the lowest microhardness among the four sample types. However, due to the low carbon concentration (approximately 1.06%) and weak mechanical stability of the retained austenite, martensitic transformation was readily induced during sliding wear. This transformation contributed to extra hardening and provided additional compressive stress on the contact surface through volume expansion, thereby significantly improving wear resistance. Consequently, the intercritical Q&P samples exhibited the best wear performance, even surpassing that of the Q&T 860 sample.

Based on these experimental results, the mechanical stability—rather than the amount—of retained austenite in martensitic steel plays the critical role in improving wear resistance under the given friction parameters.

Keywords: medium carbon quenching-partitioning steel, dry sliding wear, wear and abrasion, retained austenite, martensitic transformation

1. Introduction

In 2003, Speer et al. [?] proposed a novel quenching and partitioning (Q&P) heat treatment process based on the thermodynamic theory of constrained carbon partitioning. This process involves rapid cooling of austenitized steel to a temperature between the martensite start (Ms) and finish (Mf) temperatures to form a majority of martensite, holding at this temperature or above to allow carbon diffusion from supersaturated martensite to untransformed austenite, and final quenching to room temperature to obtain a high content of carbon-enriched retained austenite [?]. In Q&P steels, the hard martensite matrix ensures strength while the soft retained austenite improves ductility, resulting in an excellent strength-ductility balance [?, ?, ?, ?, ?, ?]. Additionally, the use of conventional alloying elements such as C, Mn, and Si makes Q&P steels relatively inexpensive and suitable for widespread engineering applications, potentially replacing steels with complex processing routes or high costs [?].

Medium-to-high carbon martensitic steels containing Mn and Si are important wear-resistant materials [?, ?, ?, ?]. Research has shown that microstructural characteristics affect wear mechanisms in martensitic steels. Due to the high content of retained austenite in the martensitic matrix of Q&P steels, their friction and wear behavior must differ from traditional quenching and tempering (Q&T) steels. Regarding retained austenite in martensitic steels, one viewpoint [?] suggests that its presence reduces hardness, leading to decreased wear resistance with increasing soft-phase content. An alternative perspective [?, ?, ?, ?, ?, ?, ?, ?, ?, ?] argues that retained austenite can induce martensitic transformation under frictional stress, producing additional hardening and increasing surface hardness to improve wear resistance. For example, Wei et al. [?] found that metastable retained austenite in TRIP steels significantly enhances wear resistance. A third viewpoint [?, ?, ?, ?, ?] considers the effect of retained austenite content on wear resistance negligible. Arques et al. [?] observed in carbonitrided steels that high retained austenite content, despite low hardness, could improve wear resistance through friction-induced martensitic transformation, while low retained austenite content, though inherently hard, showed limited transformation effects and possible tempering softening. Overall, both conditions yielded similar wear resistance. Thus, the role of retained austenite in the wear resistance of martensitic steels remains controversial.

Previous studies have examined newly developed Q&P and quenching-partitioning-tempering (Q-P-T) steels, but focused primarily on impact wear [?, ?, ?] or vibratory wear [?] in low or high carbon ranges, where applied loads are instantaneous and peak at high values. Moreover, Q&P studies used bainitic rather than traditional Q&T samples as references [?], while Q-P-T studies could not isolate carbide effects [?]. These limitations prevent clear elucidation of retained austenite' s role in Q&P samples. Therefore, this work uses traditional Q&T samples as a reference, applying intercritical and full austenitization Q&P treatments to medium carbon Fe-0.4C-1.5Mn-1.5Si steel by varying austenitization temperature to tailor microstructure. Dry sliding wear tests under constant low loading were conducted to investigate wear behavior and mechanisms, providing theoretical support for understanding friction and wear modes in medium carbon Q&P steels under service conditions.

2. Experimental Methods

A Gleeble 3500 thermal simulator was used to determine characteristic transformation temperatures via dilatometry: the austenite transformation finish temperature (Ac) was 828°C, while Ms and Mf were 303°C and 122°C, respectively. Four specimens (100 mm × 100 mm × 11 mm) were cut by wire electrical discharge machining and heated with the furnace to 800°C (1 piece), 860°C (2 pieces), and 1000°C (1 piece), holding for 10 minutes. One specimen austenitized at 860°C was directly water-quenched to room temperature, then tempered in a salt bath at 420°C for 2 minutes to obtain the Q&T 860 sample. The other three specimens were quenched in a salt bath at 240°C (the quench stop temperature) for 1 minute, transferred to a salt bath at 420°C for 2 minutes for partitioning, and finally water-quenched to room temperature to obtain different Q&P samples: Q&P 800, Q&P 860, and Q&P 1000. The selection of quench stop and partitioning temperatures followed the methodology in reference [?].

Dry sliding wear tests were conducted using a UMT TriboLabTM tribometer in a ball-on-flat linear reciprocating configuration. Samples (20 mm × 20 mm × 10 mm) were ground and polished before sliding against 8 mm diameter tungsten carbide balls. The flat sample remained stationary while the ball reciprocated across the surface. Test parameters were: load 50 N, reciprocating frequency 2 Hz, stroke length 20 mm, sliding speed 40 mm/s, and total sliding distance 144 m. Since test parameters were fixed, wear volume V (mm^3) was used to quantify wear loss. The steady-state friction coefficient and wear volume V were averaged over three tests. For accurate wear volume measurement, cross-sectional profiles of wear tracks were measured using a Bruker Contour GT-1 white light interferometric 3D optical microscope.

Microstructures were examined using an Imager A1m optical microscope (OM) and a LYRA3 TESCAN scanning electron microscope (SEM) after etching with 4% nital. The same SEM was used to observe worn surface morphologies, with an attached energy dispersive spectrometer (EDS) for analyzing wear debris composition. For transmission electron microscopy (TEM), ~0.3 mm thick slices were cut parallel to the worn surface. The worn surface was cleaned and protected with AC paper before grinding from the cut surface. Samples were then electropolished in 5% perchloric acid ethanol solution at approximately -20°C using a twin-jet polisher, followed by ion milling with a Leica EM RES 102 for ~10 minutes to remove surface corrosion products. TEM observation was performed on a JEOL 2100F at 200 kV.

Retained austenite content was measured using a D/Max-2550/PC X-ray diffractometer (XRD) with Cu K radiation at 35 kV and 20 mA, scanning from 40° to 105° at 2°/min. The direct comparison method was applied: correlation coefficients C and C for austenite ( ) and martensite ( ) phases were determined, and the volume fraction of retained austenite (f) was calculated using Equation (1) [**?**]:

where I and I are the integrated intensities of martensite and austenite diffraction peaks, respectively. The (200) , (211) , (200) , (220) , and (311) peaks were selected for paired combinations, and the average retained austenite content was calculated. The lattice parameter a (nm) of retained austenite was determined from the (111) peak position and used to calculate carbon concentration x (mass fraction, %) using Equation (2) [**?**]:

Martensite carbon concentration x (mass fraction, %) was then calculated based on mass conservation.

Microhardness of sample surfaces before and after wear testing was measured using a 402 SXV Vickers microhardness tester with a 9.8 N load and 15 s dwell time.

2.1 Microstructure Analysis

Figure 1 [Figure 1: see original paper] shows OM images of samples after different heat treatments, all exhibiting martensitic matrices. Due to austenitization below Ac , the Q&P 800 sample contained blocky white ferrite with an area fraction of approximately (5.51±0.3)% (Fig. 1b), corresponding to a volume fraction of ~6.75%, consistent with Thermo-Calc predictions. The other three samples, austenitized above Ac , showed no ferrite and typical martensitic structures; retained austenite could not be resolved by OM due to its low volume fraction. The Q&P 1000 sample exhibited more pronounced lath features and larger martensite dimensions due to its higher austenitization temperature.

Prior to wear testing, XRD measurements of (200) , (211) , (200) , (220) , and (311) peaks (Fig. 2 [Figure 2: see original paper]) were used to calculate retained austenite volume fractions of approximately 22.28% (Q&P 800), 14.37% (Q&P 860), and 13.79% (Q&P 1000) using the direct comparison method [?]. No valid austenite peaks were detected in the Q&T 860 sample, indicating retained austenite content below 3% (negligible). The similar retained austenite contents in Q&P 860 and Q&P 1000 suggest that prior austenite grain size differences do not significantly alter transformation kinetics, consistent with findings in low-carbon steels by Huang et al. [?]. The Q&P 800 sample formed a dual-phase structure of carbon-depleted ferrite and carbon-enriched austenite during isothermal holding at 800°C. The enriched austenite had a lower Ms temperature, enabling greater retained austenite formation after subsequent Q&P processing. Calculated carbon concentrations x in retained austenite were (1.06±0.01)% (Q&P 800), (1.37±0.02)% (Q&P 860), and (1.38±0.02)% (Q&P 1000), with Q&P 800 showing the lowest carbon content.

As listed in Table 1 , the Q&T 860 sample exhibited the highest microhardness (546 HV) before wear testing, while Q&P 800 showed the lowest (334 HV). Q&P 860 and Q&P 1000 had intermediate hardness values of 452 HV and 449 HV, respectively. This result aligns with microstructural analysis: increased soft-phase ferrite and austenite content reduces hardness.

2.2 Dry Sliding Friction and Wear Performance

Figure 3a [Figure 3: see original paper] shows the dynamic friction coefficient evolution for all four samples, with steady-state average values listed in Table 2 . All samples exhibited an initial rapid increase in friction coefficient, followed by a sharp decrease and gradual stabilization with fluctuations. This behavior relates to surface changes during friction. Generally, both sample and counterface ball surfaces contain numerous asperities that impede relative sliding. At the onset of friction, small contact areas cause intense surface disruption through asperity fracture, characterizing a running-in period with high friction coefficients and wear rates [?]. Subsequently, accumulated wear debris provides lubrication, reducing friction coefficients. Temperature rise and surface softening during continued wear also contribute to friction reduction. As contact area increases and debris acts as abrasive particles, friction coefficients rise and stabilize, marking the transition to steady-state wear with reduced wear rates. Notably, Q&P 860 and Q&P 1000 samples showed maximum running-in friction coefficients higher than their steady-state values, with similar average steady-state coefficients (~0.56). In contrast, Q&P 800 exhibited steady-state friction coefficients (~0.79) comparable to its running-in maximum—the highest among all samples. The Q&T 860 sample showed steady-state friction coefficients significantly higher than its running-in maximum, though slightly lower than Q&P 800. This demonstrates that bulk hardness alone cannot predict friction coefficients or wear resistance [?, ?].

Figure 3b shows cross-sectional profiles of wear tracks, where X is track width, Y is track length (10 mm), and Z is track depth. Wear volume V (wear loss) was calculated using Equation (3). Table 2 indicates the wear volume ranking: Q&P 800 < Q&T 860 < Q&P 860 < Q&P 1000, showing decreasing wear resistance. Despite containing the highest retained austenite content and some ferrite, Q&P 800 exhibited superior wear resistance over Q&T 860. Comparison of Q&P 860 and Q&P 1000 shows that prior austenite grain refinement provides strengthening and improves wear resistance, yet Q&P 860's performance remains far below Q&P 800.

Figure 4 [Figure 4: see original paper] presents worn surface morphologies after wear testing. Preliminary assessment indicates Q&P 800' s contact surface behavior resembles Q&T 860, consistent with their similar friction coefficients—both maintaining relatively hard surface contact during steady-state. Q&P 860 and Q&P 1000 show similar morphologies, corresponding to their comparable friction coefficients and indicating nearly identical wear mechanisms.

Figures 4a and 4c reveal shallow plowing grooves with micro-pits on Q&T 860 and Q&P 800 surfaces, indicating material removal primarily through plastic deformation with minor brittle fracture (abrasive wear mechanism). Higher magnification images (Figs. 4b and 4d) show more severe plastic deformation in Q&P 800 but more brittle spalling pits in Q&T 860. Both surfaces contained dispersed white particles identified by EDS as iron oxides (Fig. 4i), confirming a mixed wear mechanism dominated by abrasive wear with minor oxidative wear for both Q&P 800 and Q&T 860.

Figures 4e–4h show discontinuous deep/wide plowing grooves with micro-cutting traces on Q&P 860 and Q&P 1000 surfaces, indicating severe plastic deformation. EDS analysis of protrusions on the worn ball surface after testing against Q&P 860 (Fig. 4j) revealed Fe, C, Mn, Si, and O, confirming material transfer from sample to ball (adhesive wear). During wear, actual contact area is much smaller than nominal area, causing stresses at asperity contacts to reach the yield limit and produce plastic deformation. Contact point metals experience plastic flow and possible instantaneous high temperatures, promoting adhesion between the two metals. Since the ball is harder than the sample, adhesive junctions fracture preferentially at the sample surface, leaving pits on the sample and protrusions on the ball [**?**, **?**]. Higher magnification images (Figs. 4f and 4h) show more numerous and larger oxides on Q&P 1000 than Q&P 860, indicating more severe oxidative wear. Thus, Q&P 860 and Q&P 1000 exhibit mixed wear mechanisms dominated by abrasive and adhesive wear with (severe) oxidative wear.

3. Discussion

To further characterize wear features, worn contact surfaces were etched with 4% nital and examined by SEM (Fig. 5 [Figure 5: see original paper]), with microhardness measurements at wear tracks (averaged over three measurements) listed in Table 1.

Figures 5a and 5b show micro-pits of varying sizes on Q&T 860 and Q&P 800 surfaces. Q&T 860 pits contain obvious cracks with minimal surrounding deformation, similar to regions far from pits. Microhardness near these pits (~527 HV) is slightly lower than the pre-wear bulk hardness, indicating limited plastic deformation of martensite during friction and a wear mechanism dominated by brittle spalling (abrasive wear). Q&P 800 pits are smaller and nearly crack-free, demonstrating superior resistance to brittle fracture compared to Q&T 860. This difference likely stems from ferrite' s toughening effect. Surface relief observations show smaller features near pits and larger features in surrounding areas, indicating varying deformation levels. Microhardness near Q&P 800 pits (~430 HV) exceeds both remote regions (419 HV) and pre-wear hardness (340 HV), reflecting work hardening during friction that improves wear resistance.

Figures 5c and 5d show intermittent micro-cutting and micro-pit morphologies on Q&P 860 and Q&P 1000 worn surfaces. Micro-cutting traces run parallel to the sliding direction with 2–5 m widths. High-density micro-cutting indicates easy surface plastic deformation and damage, with wider traces and obvious cracking around micro-cutting/pits in Q&P 1000, confirming its lower wear resistance. Microhardness values near micro-cutting regions (434 HV and 420 HV) are similar to remote regions and lower than pre-wear hardness, indicating no significant work hardening in these samples during friction.

TEM bright-field images of Q&P 800 (ignoring ferrite) show a predominantly lath martensite structure (Fig. 6a [Figure 6: see original paper]). A centered dark-field image obtained with g = (Fig. 6b) reveals a high fraction of interlath retained austenite. Selected area electron diffraction (SAED) patterns confirm Kurdjumov-Sachs (K-S) and Nishiyama-Wassermann (N-W) orientation relationships between retained austenite and lath martensite (Fig. 6b inset). These typical features have been verified in other Q&P steels, differing only in retained austenite width (content) [?].

Post-wear examination of Q&P 800 revealed additional special structures. SAED analysis and centered dark-field imaging (g =) identified a bcc structure (Fig. 6d) lacking blocky ferrite or lath martensite characteristics, identified as twinned (high-carbon) martensite. This martensite appears as fine, clustered, parallel laths with step features, confirmed as deformation-induced [?]. Additionally, high-density parallel friction bands resembling slip lines were observed near the lower right corner of the martensite cluster (arrow in Fig. 6d), indirectly confirming severe plastic deformation during friction.

Using traditional Q&T samples as reference, these results demonstrate that wear performance of Q&P samples with different austenitization treatments is not solely determined by bulk hardness from heat treatment—presence of soft phases does not necessarily degrade wear resistance. Microstructure significantly influences wear mechanisms. Specifically, despite microstructural refinement improving Q&P 860 wear resistance over Q&P 1000, its performance remains less than half that of Q&P 800. Q&P 800 even surpasses Q&T 860 in wear resistance. Notably, Q&P 800 contains the lowest hard-phase martensite content, while Q&T 860 contains virtually no soft phases. Since ferrite is known to be detrimental to wear resistance [?], the divergent wear performance of Q&P 800 versus the other Q&P samples must arise from different deformation and transformation behaviors of retained austenite during friction.

First, Q&P 860' s inferior wear resistance compared to Q&T 860 stems from its ~14.37 vol% retained austenite. Comparison of hardness changes before and after testing (Table 1) and XRD analysis (Fig. 2) showing minimal retained austenite reduction (from ~14.37% to ~12.72%) indicate negligible martensitic transformation at contact surfaces under current test conditions. Retained austenite participated in friction only as a soft phase, undergoing plastic deformation and tearing without transformation-induced hardening. Its low hardness negatively impacted Q&P 860' s wear resistance, resulting in performance far below Q&T 860.

Second, Q&P 1000 showed similar patterns in microhardness and retained austenite content change (from ~13.79% to ~11.91%) as Q&P 860, indicating that its retained austenite also primarily underwent plastic deformation and tearing without significant martensitic transformation. Thus, the wear resistance difference between Q&P 860 and Q&P 1000 arises only from martensite microstructural refinement. As is well known, refined microstructures improve adhesion resistance, reducing adhesive wear in fine-grained Q&P 860 compared to Q&P 1000 (Fig. 4) and enhancing its wear resistance.

During intercritical austenitization, partial ferrite formation increased carbon content in untransformed austenite, lowering its Ms temperature. Using the same quench stop temperature (240°C) as Q&P 860 resulted in less martensite formation and reduced carbon partitioning to retained austenite. Consequently, post-heat treatment retained austenite carbon concentration was only ~1.06%, giving it lower mechanical stability than Q&P 860' s retained austenite (~1.37% C). Carbon mass conservation calculations show similar martensite carbon concentrations (~0.08%) in both samples, suggesting comparable matrix wear resistance. However, unlike Q&P 860, Q&P 800 showed significant microhardness increase (from 334 to 419 HV) and retained austenite reduction (from ~22.28% to ~11.20%) after testing. TEM confirmed twinned martensite near dense friction bands, evidencing pronounced martensitic transformation, surface hardening, and compressive stress from lattice expansion that improved wear resistance and altered wear mechanisms. Thus, Q&P 800' s retained austenite contributed significantly to wear performance, compensating for ferrite' s detrimental effect and even surpassing Q&T 860.

This work demonstrates that for medium carbon Q&P steels, pre-wear hardness does not directly correlate with wear resistance. Under low loading conditions, the stability—not the amount—of soft retained austenite dominates wear performance. Evaluating Q&P wear resistance should consider potential strain-hardening effects during friction rather than relying solely on pre-test hardness.

Wear mechanism changes are also closely related to operating conditions; varying loads and sliding speeds affect retained austenite stability. Excessive loading could induce martensitic transformation in Q&P 860 and Q&P 1000 retained austenite, while high sliding speeds might cause significant surface heating that influences stability. To avoid such complexities, this study focused on low-load, low-speed conditions, which may limit the generalizability of results.

4. Conclusions

1. Q&P 860 and Q&P 1000 samples exhibited the lowest wear resistance due to high mechanical stability of carbon-enriched retained austenite, which resisted martensitic transformation and extra hardening during friction, participating only as a soft phase that contributed nothing to wear resistance. The slight difference between these two samples arose from grain refinement strengthening due to different austenitization temperatures.
2. Although the intercritically austenitized Q&P 800 sample contained more retained austenite and some ferrite, its lower carbon concentration in retained austenite resulted in poor mechanical stability. This facilitated martensitic transformation during friction, providing additional work hardening and stress relaxation that significantly improved wear resistance, even exceeding that of Q&T 860.

References

[1] Speer J G, Matlock D K, DeCooman B C, et al. Carbon partitioning into austenite after martensite transformation [J]. Acta Mater., 2003, 51: 2611
[2] Santofimia M J, Nguyen-Minh T, Zhao L, et al. New low carbon Q&P steels containing film-like intercritical ferrite [J]. Mater. Sci. Eng., 2010, A527: 6429
[3] Santofimia M J, Zhao L, Petrov R, et al. Microstructural development during the quenching and partitioning process in a newly designed low-carbon steel [J]. Acta Mater., 2011, 59: 6059
[4] Wang L, Speer J G. Quenching and partitioning steel heat treatment [J]. Metall. Microstruct. Anal., 2013, 2: 268
[5] Zhao C, Tang D, Jiang H T, et al. Process simulation and microstructure analysis of low carbon Si-Mn quenched and partitioned steel [J]. J. Iron Steel Res., 2008, 15: 82
[6] Speer J G, Rizzo Assuncao F C, Matlock D K, et al. The "Quenching and Partitioning" process: Background and Recent Progress [J]. Mater. Res., 2005, 8: 417
[7] Zhang K, Xu W Z, Guo Z H, et al. Effects of novel Q-P-T and traditional Q-T processes on the microstrucuture and mechanical properties of martensitic steels with different carbon content [J]. Acta Metall. Sin., 2011, 47:489
[8] Weng Y Q, Dong H, Gan Y. Advanced Steels [M]. Beijing: Metallurgical Industry Press, 2011: 67
[9] Wang C Y, Shi J, Liu S, et al. Study on three-body impact-abrasion of steel

treated by quenching-partitioning-tempering process [J]. Chin. J. Mater. Res., 2009, 23: 305
[10] Liu S G, Dong S S, Yang F, et al. Application of quenching-partitioning-tempering process and modification to a newly designed ultrahigh carbon steel [J]. Mater. Des., 2014, 56: 37
[11] Wang C Y, Li X D, Chang Y, et al. Comparision of three-body impact abrasive wear behaviors quenching-partitioning-tempering and quenching-tempering 20Si2Ni3 steels [J]. Wear, 2016, 362-363: 121
[12] Hu F, Wu K M, Hodgson P D. Effect of retained austenite on wear resistance of nanostructured dual phase steels [J]. Mater. Sci. Technol., 2016, 32: 40
[13] Kim H J, Kweon Y G. The effects of retained austenite on dry sliding wear behavior of carburized steels [J]. Wear, 1996, 193: 8
[14] Riviere J P, Brin C, Villain J P. Structure and topography modifications of austenitic steel surfaces after friction in sliding contact [J]. Appl. Phys., 2003, 76A: 277
[15] Zandrahimi M, Bateni M R, Poladi A, et al. The formation of martensite during wear of AISI 304 stainless steel [J]. Wear, 2007, 263: 674
[16] Molinari A, Pellizzari M, Gialanella S, et al. Effect of deep cryogenic treatment on the mechanical properties of tool steels [J]. J. Mater. Proc. Technol., 2001, 118: 350
[17] Wei X C, Li J, Meng H. Tribological characteristic of HSLA TRIP steel containing meta-stable retained austenite [J]. Tribology, 2006, 26: 49
[18] Liu Y, Yu W P, Zhang J D, et al. Effect of deep cryogenic treatment on microstructure and mechanical properties of T8A steel [J]. Trans. Mater. Heat Treatment, 2010, 31(10): 48
[19] Wu H Y. Study on friction-induced martensite transformation of retained austenite and its wear resistance [J]. Hot Work. Technol., 2007, 36: 32
[20] Arques J L, Prado J M. The dry wear resistance of a carbonitrided steel [J]. Wear, 1985, 103: 321
[21] Siepak J. The influence of contact stress on the wear of a carburized steel case with a high content of retained austenite [J]. Wear, 1982, 80: 301
[22] Jatczak C F. Retained Austenite and its Measurement by X-ray Diffraction [M]. Detroit: Society of Automotive Engineers, Inc., 1980: 6
[23] Xiong Z L, Cai Q W, Jiang H T, et al. Research on mechanical stability of austenite in TRIP steels [J]. J. Mater. Eng., 2011, 3: 11
[24] Huang F, Yang J L, Guo Z H et al. Effect of partitioning treatment on the mechanical property of Fe-0.19C-1.47Mn-1.50Si steel with refined martensitic microstructure [J]. Metall. Mater. Trans., 2016, 47A, 1072
[25] Yao Y Q. The research on friction and wear properties of low-carbon low-alloy steels [D]. Nanjing: Nanjing University of Science and Technology, 2013
[26] Wang Y, Lei T Q, Liu J J. Tribo-metallographic behavior of high carbon steels in dry sliding II. Microstructure and wear [J]. Wear, 1999, 231: 12
[27] Sun J S. Wear of Metals [M]. Beijing: Metallurgical Industry Press, 1992: 182
[28] Xu Y L, Qiu H Q, Pang Z G, et al. Wear resistance at high temperature of

H13 steel after RE-N-C-S-V-Nb multi-element penetrating [J]. Surf. Technol., 2015, 44: 60
[29] Wang M S, Zhu Y F. The plastic induced martensite transformation of a medium manganese austenite steel [J]. Phys. Tests. Chem. Analy. A: Phys. Test., 1991, 1:23
[30] Gao C Q. The Friction Wear and Heat Treatment of Metals [M]. Beijing: Mechanical Industry Press, 1988: 93

Note: Figure translations are in progress. See original paper for figures.