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Influence of microstructure on the erosive-wear resistance of JIS SKH57 high-speed tool steel

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Abstract

Recently, to improve the service life of components used in surface treatment equipment under severe wear environments, high Cr cast irons are increasingly being replaced by cast steels with compositions equivalent to Mo-based high-speed tool steels. The purpose of this study was to investigate the relationship between the erosive-wear resistance and microstructure of JIS SKH57, a representative high-speed tool steel manufactured by self-hardening mold casting. The microstructures of the as-cast, as-quenched, and tempered samples were examined using scanning electron microscopy, X-ray diffraction, and differential scanning calorimetry. It was demonstrated that a slower cooling rate after solidification resulted in the formation of various secondary carbides within the matrix, in addition to primary carbides precipitated along the grain boundaries. Crucially, tempering at 525 °C significantly enhanced the wear resistance of the experimental cast steel, which is primarily attributed to the prominent secondary hardening effect. The erosive-wear mechanism was further discussed in terms of surface hardening, which is associated with increased dislocation density and the volume of

strain-induced martensitic transformation from retained austenite.

Keywords: High-speed tool steel, hardness, erosive-wear resistance, carbide, dislocation density, retained austenite

1. Introduction

Components such as blades and liners used in surface treatment equipment require high wear resistance, so they are commonly made from high Cr cast iron. In such the equipment, the continuous impact of blasting media, including steel grit and shot, causes simultaneous abrasive and erosive-wear on the component surfaces. These wear modes subsequently result in a reduction in component weight ^(Ref.1-2). From a material perspective, the factors influencing these wear mechanisms can be broadly classified into the carbides and the metallic phases within the matrix. During erosive-wear, the impact of steel grit and shot imposes significant external stress on the castings; however, the surface-hardening capacity varies depending on the constituent metallic phases and their phase volume fractions within the matrix. Among these phases, metastable austenite exhibits a superior hardening capacity compared to pearlite, bainite, and martensite ^(Ref.3-4). As for carbides, it has been reported that the wear resistance of high Cr cast iron is significantly influenced by carbide morphology, amount, and distribution etc. The addition of carbide-forming elements changes both the types and amount of primary and secondary carbides ^(Ref. 5). In terms of hardness, titanium-, tungsten-, and molybdenum-based carbides exhibit higher values than chromium-based ones. Consequently, to further enhance the durability of high Cr cast iron, extensive research regarding its microstructure and wear resistance, alongside alloy development, have been actively conducted on high Cr cast irons containing single or combined additions of Ti, W and Mo ^(Ref. 6-8).

On the other hand, high-speed tool steel is a type of alloy steel primarily used for cutting tools and dies. Depending on the application, it is classified into three series of JIS standard steel grades: (1) the W-based series containing 11.5 – 19.0 mass% W, (2) the Mo-based series containing 3.2 – 10.0 mass% Mo, and (3) the powder metallurgy Mo-based series containing 4.7 – 5.3 mass% Mo. In addition to W and Mo, each steel grade series also contains V and Co ^(Ref. 9). These steels are characterized by the coexistence of various types and large amounts of carbides. Furthermore, Co dissolves into the matrix, and suppresses carbide growth. This feature dramatically enhances the wear resistance and maintains the high-temperature strength of the steel. Recently, in response to severe wear environments, high Cr cast irons are increasingly being replaced by cast steels with compositions equivalent to Mo-based high-speed tool steels, particularly JIS SKH55 and SKH57 (hereinafter high-speed tool steels). Despite being more costly than high-Cr cast irons, these steels significantly extend the service life of manufactured components like liners and blades. Consequently, they provide excellent cost performance by drastically minimizing both maintenance costs and replacement downtime.

To date, many studies have focused on the microstructure and mechanical properties of powder metallurgy, sintered high-speed steels, and additive manufacturing ^(Ref. 10–14). The presence of various types of carbides within the martensitic matrix plays an important role in improving hardness and abrasive wear resistance. However, there are limited reports available regarding the erosive-wear characteristics of high-speed tool steels, and the specific factors influencing their erosive-wear behavior remain unclear.

Therefore, this study investigates the relationship between the hardness and microstructure of SKH57 steel in the as-cast, as-quenched, and tempered states. In addition, the effect of microstructural differences on erosive-wear resistance were examined, and the wear mechanisms were thoroughly discussed.

2. Experimental Procedure

2.1 Material preparation

The experimental cast steel (hereinafter referred as SKH57 cast steel) was prepared using a 300 kg high-frequency induction furnace. After the raw materials was melted, the melt was heated to 1600 °C, transferred to a ladle, and then poured at 1572 °C into a sand mold for a type-A Y-block ingot in accordance with JIS G307. The chemical composition of the SKH57 cast steel, measured by optical emission spectrometry and carbon sulfur analyzer, is shown in Table 1.

Table 1 Chemical compositions of SKH57 cast steel (mass%).

C	Si	Mn	Cr	Mo	W	V	Co	Fe
1.24	0.41	0.17	4.08	3.21	9.75	3.28	10.15	bal.

To prevent decarburization and quench cracking, the obtained Y-block ingot was heat-treated at 1210 °C for 1 h in a vacuum atmosphere, followed by forced-gas cooling in a high-pressure nitrogen atmosphere, as shown in Figure 1. Subsequently, tempering was performed for 1 h in the temperature range of 300 °C to 650 °C. For comparison, 25Cr cast iron (main composition: Fe – 24.73 mass% Cr – 0.58 mass% Si – 0.59 mass% Mn – 0.06 mass% Mo – 2.54 mass% C), quenched from 1050 °C, were tempered under the same conditions.

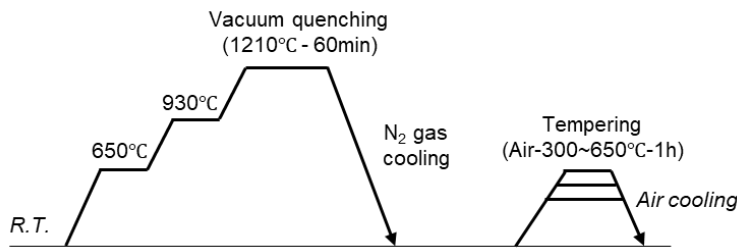


Figure 1 Heat treatment conditions of experimental cast steel.

2.2 Erosive-wear test

The erosive wear test samples with dimensions of 35 mm x 40 mm x 15 mm (width x length x thickness) were cut from the Y-block ingots using by wire-cutting. The erosive wear test was conducted using a direct-pressure air blasting machine. Each sample was fixed to the test table inside the chamber, and the distance from the 6 mm diameter nozzle tip to the sample surface was adjusted to 100 mm. Iron grit (Fe-1.0C) with a hardness of 850 HV and an average particle diameter of 1.2 mm was used as the erodent.

For each test cycle, 25 kg of iron grit was accelerated from the nozzle tip at a pressure of 0.5 MPa and impacted the sample surface at a normal angle. The duration of each cycle was approximately 0.85 s. After every 5 cycles, the weight loss of each sample was measured using a high-precision electric balance with an accuracy of 0.1 mg. The samples were subjected to 20 cycles of wear tests; however, the test was terminated prematurely if a through-hole formed due to the grit impact. After testing, the surface appearance of each sample was photographed.

2.3 Hardness measurement

The macrohardness was measured out using a Vickers hardness tester (Akashi, AVK-A) under a load of 20 kgf (196 N) and a dwell time of 15 s. In addition, the hardness distribution in the depth direction from the worn surfaces after the wear tests was measured using a micro-Vickers hardness tester (Shimadzu, HMV-G21 FA) under a load of 0.3 kgf (2.94 N) and a dwell time of 15 s. Each value in the hardness distribution represents the average of 10 measurements. To evaluate the matrix hardening after the wear tests, the hardness of the matrix region, excluding primary carbides, was measured 10 times under a load of 0.2 kgf (1.96 N) using the same tester, and the average value was determined.

2.4 Microstructural characterization

Samples for microstructure observation were sequentially wet-ground using SiC emery paper and buff-polished with an alumina suspension, followed by etching with Vilella's reagent (5 mL HCl, 1 g picric acid, and 100 mL ethanol) for 10 s. Microstructures were performed using an optical microscope (OM, Keyence, VHX-2000) and a scanning electron microscope equipped with an energy-dispersive X-ray detector (SEM-EDS, JEOL, IT-100LA,) operated at an acceleration voltage of 15 kV. Phase identification was carried out using X-ray diffraction (XRD, Empyrean, Malvern Panalytical) using Cu target. The XRD measurement was conducted at a 45 kV and 40 mA, with a 2θ scanning range from 35° to 65° and a step size of 0.013° . The amount of retained austenite was measured using an XRD instrument (Rigaku, Ultima IV) equipped with a rotating / oscillating sample stage and using Mo target, scanning a 2θ range from 24° to 44° . The amount of retained austenite was calculated from the integrated intensity ratios of the (200) and (220) diffraction peaks of martensite, and the (220) and (311) diffraction peaks of austenite. In addition, the full width at half maximum (FWHM) of the martensite (211) peak of the SKH57 cast steel was measured using a portable X-ray residual stress analyzer (Pulstec, μ -X360) using Cr target. Based on the FWHM results, the dislocation

density was calculated according to Williamson-Hall method ^(Ref. 15). The FWHM measurements were carried out at a diffraction angle of 156.396° with 30 kV and 1.0 mA. Thermo-Calc software (database: Fe-DATA Ver.4) was used to calculate the volume fraction of equilibrium phases for the experimental cast steel.

To investigate the temperature at which carbide reaction occurs during tempering, differential scanning calorimetry (DSC) was performed using a thermal analyzer (Netzsch, STA 449 F1 Jupiter). A Pt / Rh pan with an Al_2O_3 liner was used. To prevent oxidation, argon gas was flowed at 10 mL / min, with an empty pan used as a reference. The sample, with dimensions of 5 mm in diameter and 1.0 mm in thickness, was heated to 1250 °C at a heating rate of 10 °C / min.

3. Results and Discussion

3.1 Hardness and microstructure

Figure 2 shows the hardness variation of SKH57 cast steel in the as-cast, the quenched (hereinafter AsQ), and after tempering within a temperature range of 300 °C to 650 °C (hereinafter T). For comparison, the hardness of 25Cr cast iron heat-treated under the same conditions are also presented in the same figure. The hardness of the as-cast SKH57 cast steel sample was 600 HV, while that of the AsQ sample was 700 HV, almost same as the 25Cr cast iron. However, upon tempering, SKH57 cast steel exhibited significant secondary hardening, distinct from 25Cr cast iron. In the 25Cr cast iron, hardness in the temperature range of 300 °C to 500 °C was slightly higher than that of the AsQ sample. At temperatures above this range, the hardness drastically decreased. Meanwhile, in the SKH57 cast steel, the hardness gradually increased with increasing tempering temperature and reached a maximum hardness of 825 HV at around 525 °C, which was significantly higher than that of AsQ sample. Furthermore, at tempering temperatures of 550 °C or higher, SKH57 cast steel showed tempering softening resistance, characterized by a much more moderate decline in hardness compared to 25Cr cast iron.

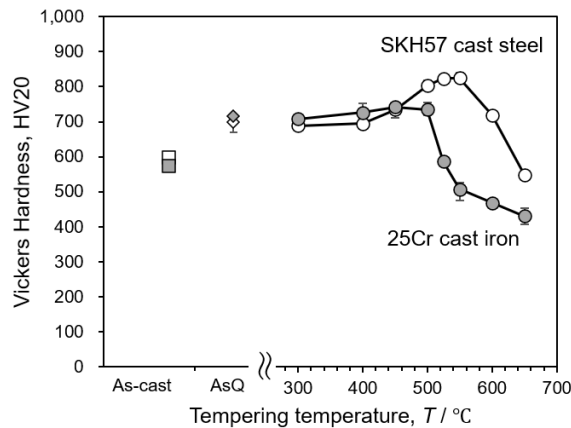


Figure 2 Change in hardness of SKH57 cast steel and 25Cr cast iron as a function of tempering temperature.

It is well known that an excessive amount of retained austenite in martensitic steels decreases hardness and leads to significant dimensional changes over time (Ref.16, 17). Although oil quenching combined with sub-zero treatment using liquid nitrogen or dry ice can eliminate retained austenite, such method requires special equipment and increase the microstructural risks associated with quench cracking and deformation in

the castings. As described in the experimental procedure, forced-gas cooling in a nitrogen atmosphere was performed after quenching to reduce these risks. However, the forced-gas cooling provides a lower cooling rate than oil quenching, which may result in a trace amount of retained austenite. Therefore, the measurement of retained austenite in all SKH57 cast steel samples was performed using the XRD method. The obtained results are shown in Figure 3. The retained austenite amount was approximately 11 vol.% in the as-cast sample and 13 vol.% in the AsQ sample. In the tempered samples, the retained austenite amount decreased almost linearly with increasing tempering temperature, which is attributed to the decomposition of retained austenite into ultra-low carbon ferrite and carbides during tempering.

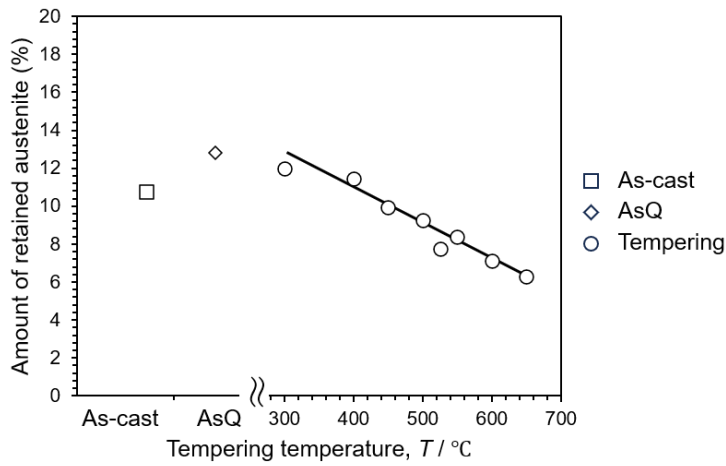


Figure 3 Measurement of retained austenite amount in SKH57 cast steel samples.

Figure 4 shows the SEM microstructure of SKH57 cast steel in the as-cast state, together with the corresponding EDS elemental composition maps of the observed area. For reference, a rapid-cooled sample used for chemical composition analysis is also shown; this sample was cut from a 35-mm-diameter round bar ingot cast into a metallic mold and subsequently water-cooled. Unlike the rapid-cooled sample, the as-cast sample exhibits many fine carbides dispersed in the matrix. This difference is due to the lower cooling rate after solidification. Due to the high content of various strong carbide-forming elements, such slower cooling rate promotes the precipitation of fine carbides (hereafter referred to as secondary carbides) within the matrix, following the completion of primary carbide formation. The primary carbides exhibited two distinct morphologies: massive and network-like. The main composition of the massive carbides was (V, Cr)-rich, whereas that of the network-like carbides was (W, Mo, V)-rich. Meanwhile, the secondary

carbides possessed a nearly globular morphology. Since Co was scarcely detected in any of the carbides, it is concluded that Co was completely dissolved in the matrix as a solid solution.

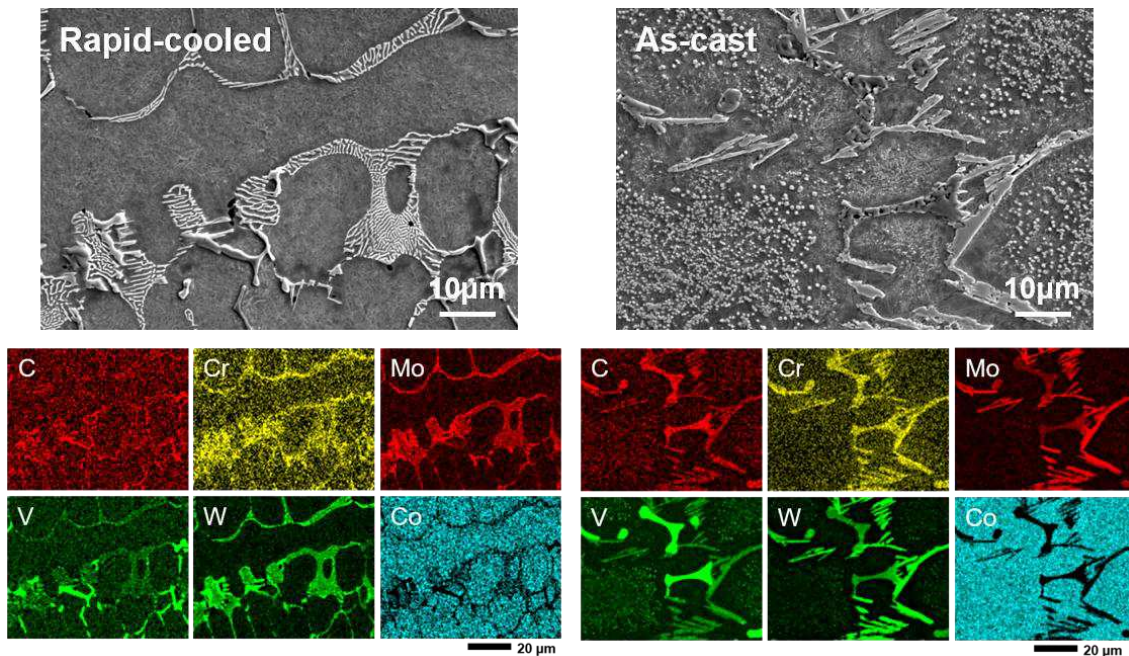


Figure 4 Microstructures of Rapid-cooled and as-cast samples analyzed by SEM-EDS. The Rapid-cooled sample was cut from a 35-mm diameter block ingot for chemical composition analysis.

The microstructures of the AsQ, T525°C, and T650°C samples of SKH57 cast steel are shown in Figure 5. In all samples, numerous secondary carbides were distributed throughout the matrix, similar to the as-cast condition. Detailed characterization of the network-like primary carbides confirms that (V, Cr-rich) and (W, Mo-rich) carbides were closely intertwined. Compared with the as-cast sample, the AsQ sample exhibited a clear trend toward finer secondary carbides dispersed in the matrix. Specifically, these carbides in the peripheral region of the matrix are finer than those in the central zone. In the tempered samples (T525°C and T650°C), the secondary carbides were densely dispersed. According to the XRD phase identification results shown in Figure 6, four types of carbides were identified in the cast steel: M_7C_3 , M_2C , M_6C and MC . Among these, M_7C_3 is a metastable carbide and appears only in the rapid-cooled and as-cast samples, whereas the remaining carbides persist in all samples. This indicates that no carbide transformation occurred during tempering. Although quantitative analysis of these carbides is not difficult, relatively high peak intensities for M_6C and MC were observed

in the T525°C and T650°C samples.

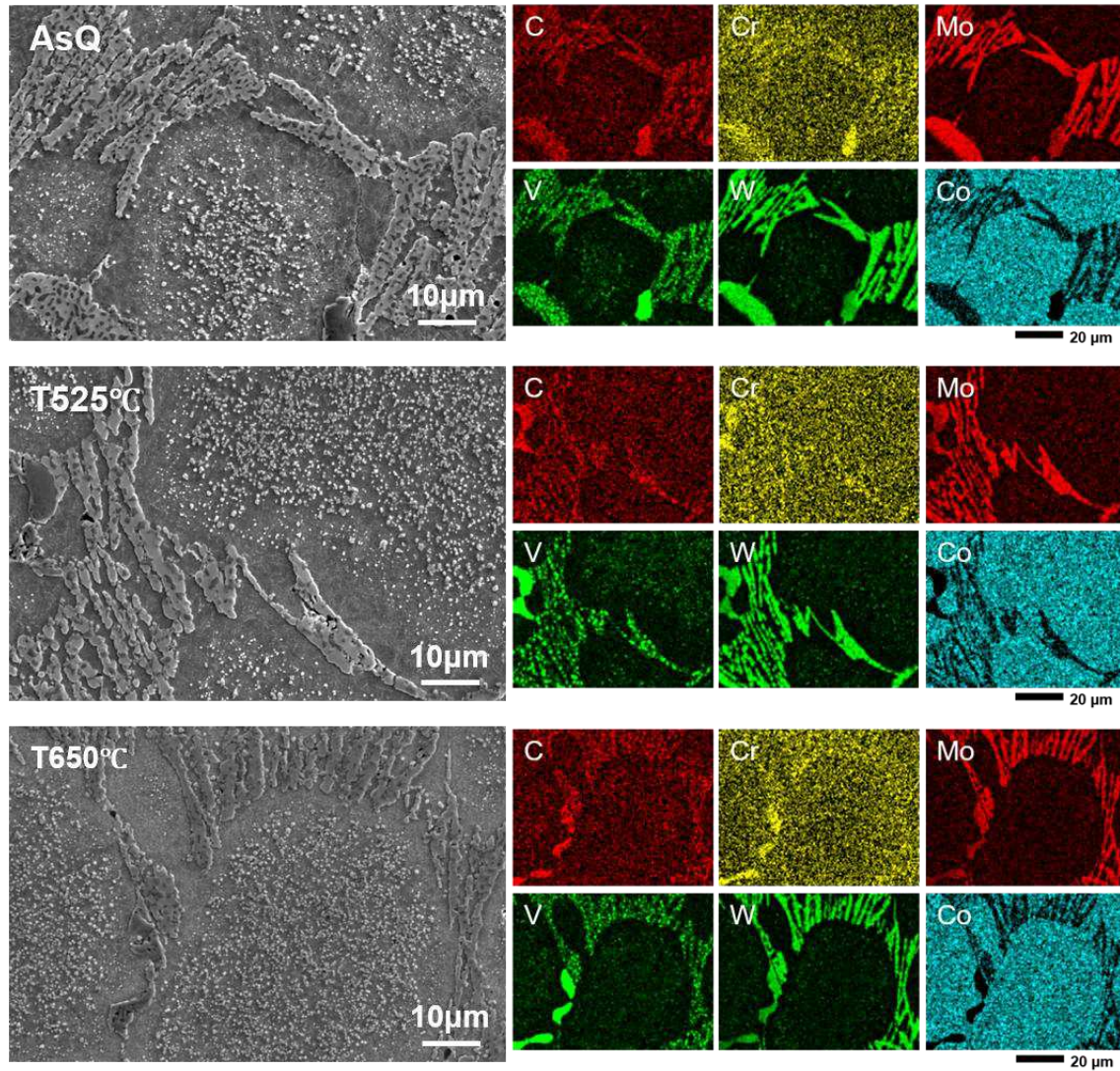


Figure 5 Secondary electron images and EDS composition maps showing the microstructure of AsQ sample, two samples tempered at 525 °C and 650 °C.

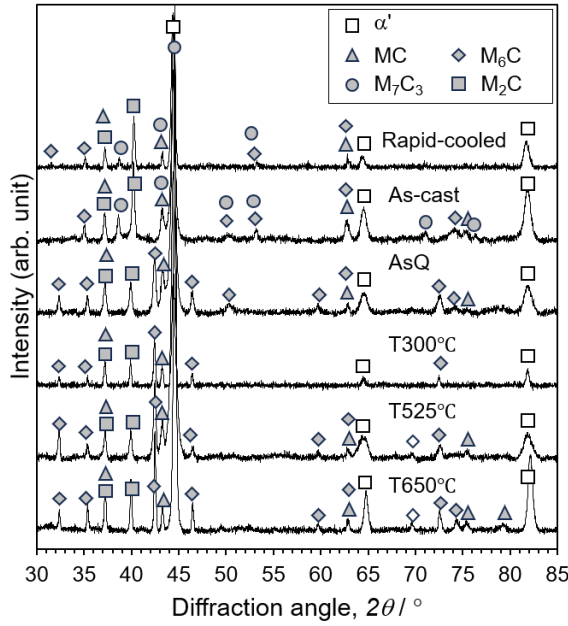
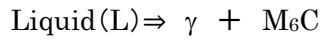
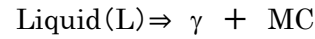


Figure 6 Phase identification for SKH57 cast steel samples by XRD.

In order to clarify the thermally stable carbides at the quenching temperature of 1210 °C, the equilibrium phase diagram was calculated using Thermo-Calc software. In addition, the change in the volume fraction of these carbides as a function of temperature was determined from the temperature-dependent phase volume fraction diagram. As seen in Figure 7, following the onset of solidification, the austenite phase (γ) along with MC-type and M_6C -type carbides formed sequentially according to the following eutectic reactions:



From the equilibrium phase diagram, it was found that these carbides coexist with austenite within the temperature range of 1150 K to 1500 K. According to the phase fraction diagram, at 1483 K (~1210 °C), which is the heat treatment temperature set in this study, the amount of MC-type carbides is slightly higher than that of M_6C -type carbides. Below 1130 K, γ transforms into the ferrite phase (α), and $M_{23}C_6$ -type carbides precipitate in addition to MC and M_6C . Furthermore, the amount of alloying elements dissolved in the matrix at 1483 K was determined based from the isothermal phase diagram. Furthermore, the solid solution amounts of alloying elements in the matrix at 1483 K were calculated. The results indicate that in the equilibrium state, Co completely

subsequently precipitate as fresh secondary carbides during tempering. However, whether the precipitation behavior proceeds through an *in-situ* or separate nucleation remains unclear. Further detailed investigations using transmission electron microscopy (TEM) will be carried out in future work. On the other hand, the prominent endothermic peak observed at 865 °C corresponds to the transformation of martensite to austenite; however, this transformation will not be discussed further.

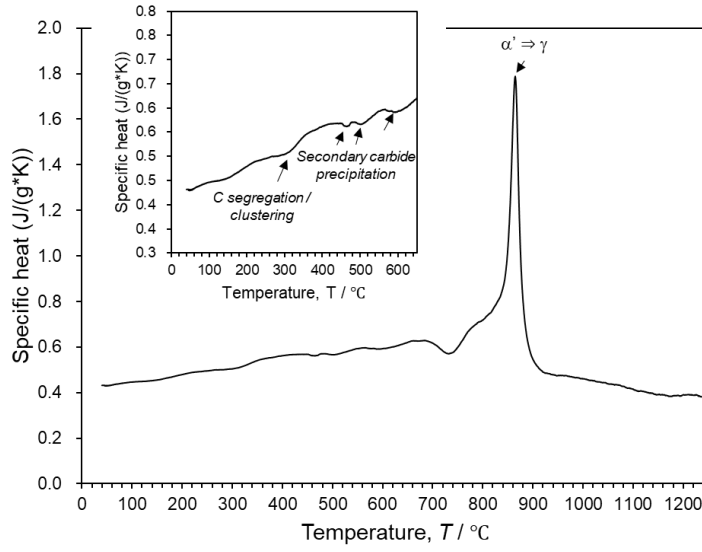


Figure 8 Specific heat curve of SKH57 cast steel obtained by DSC analysis.

Martensitic carbon and alloy steels are known to possess an extremely high dislocation density because the severe lattice distortion, accompanied plastic deformation occur during martensitic transformation. Compared with direct TEM observation of microstructure (Ref. 18) and electrical resistivity measurement (Ref. 19), the XRD method is widely used to evaluate dislocation density due to its ease of sample preparation and measurement. In the case of ultra-low carbon martensitic steels, the relationship between the full width at half maximum (FWHM, β) in XRD analysis and the dislocation density (ρ) can be closely approximated by the following equation.

$$\rho \propto \beta^2$$

Previous studies have indicated that a large FWHM corresponds to a higher dislocation density (Ref. 15, 20). However, martensitic steels containing carbon in solid-solution exhibit a body-centered tetragonal (BCT) structure characterized by a and c lattice parameters, where the a – axis shrinks and the c – axis elongates with increasing carbon content.

This leads to the splitting of XRD diffraction peaks and peak broadening. Furthermore, in high-alloy steels such as the SK and SKH series steels, the overlap between the martensite (110) peak and the M_7C_3 carbide (151) peak can also lead to an overestimation of the calculated dislocation density. To circumvent these issues, the present study evaluated the FWHM of the martensite (211) peak to indirectly assess and relatively compare the dislocation densities. Figure 9 displays these FWHM measurements obtained via XRD. The FWHM values of the martensite (211) peak reveal a clear contrast, with the AsQ sample exhibiting greater broadening than its as-cast counterpart. Upon tempering, the FWHM in the temperature range from 300 °C to 400 °C was almost the same as that of the AsQ state. However, further increases in tempering temperature induce a dramatic transition: the FWHM values increased remarkably and reached a maximum at 525 °C. The FWHM measurements exhibit excellent correlation with the hardness variations presented in Figure 2, from which the evolution of dislocation density can be classified into three distinct stages. In stage I, the AsQ sample exhibits a higher dislocation density than the as-cast sample, because the quenching treatment promotes the redistribution of alloying elements and a faster cooling rate. In stage II, the dislocation density increases in the tempering temperature range of 300 °C to 525 °C. In stage III, dislocation recovery including rearrangement and annihilation occurs in the temperature range from 525 °C to 650 °C. Regarding Stage II, W. Ding et al. provided an explanation for the increase in dislocation density ^(Ref. 21). During tempering of low-carbon steel (Fe – 1.5 mass% Mn – 0.05 mass% Nb – 0.1mass% Ti – 0.07mass% C), these interfacial vacancies separate from the carbide, diffusing from the interface to the matrix to accumulate in the direction of stress. The aggregation of supersaturated vacancies inevitably causes vacancy discs to collapse, ultimately giving rise to dislocation loops. Particularly, the dislocation density has been reported to reach its maximum within the temperature regime where secondary hardening occurs. In the present work, the broadening of the martensitic peak provides clear evidence that fresh secondary carbides precipitate during tempering, which is responsible for the secondary hardening of the SKH57 cast steel.

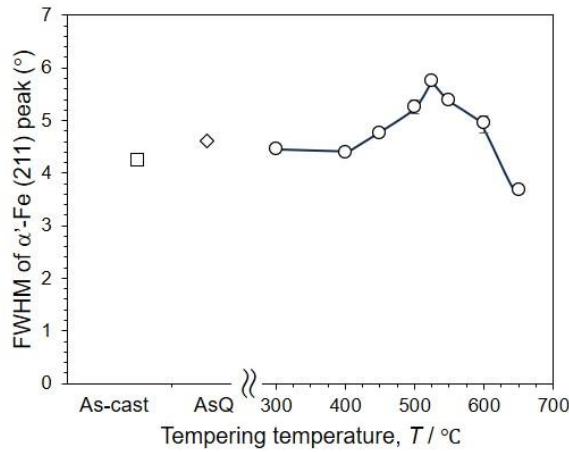


Figure 9 Change in the full width at half maximum (FWHM) of α' -Fe (211) peak of SKH57 cast steel as a function of tempering temperature.

3. 2 Erosive-wear resistance

Figure 10 and Figure 11 show the appearances of the four types of samples, and their weight loss measured every 5 cycles during wear tests. The T650°C sample exhibited the largest wear loss, leading to early penetration during the 12th test cycle. In contrast, the wear loss of the other samples almost increased linearly with the number of test cycles. The erosive-wear resistance of the SKH57 cast steel in ascending order is as follows: T650°C < as-cast < AsQ < T525°C. The trend agrees well with the hardness results shown in Figure 2. In other words, the erosive-wear resistance of the martensitic cast steel is strongly dependent on its hardness.

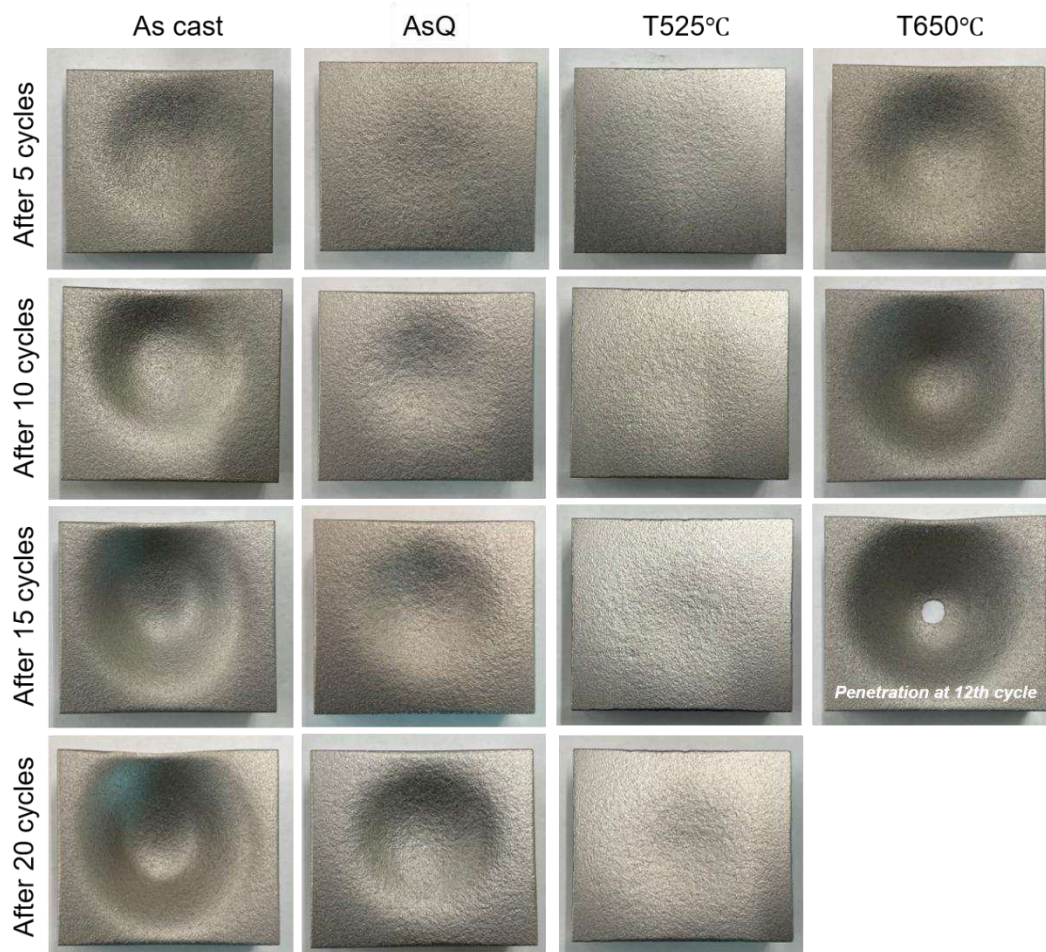


Figure 10 Appearances of the four types of samples during erosive-wear test.

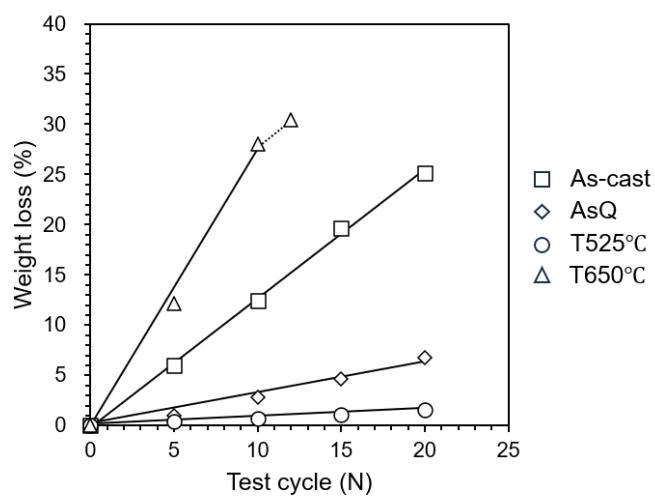


Figure 11 Weight loss results of the experimental cast steels obtained from the erosive-wear tests.

Figure 12 shows the surface appearances and low-magnification SEM images of each sample after wear testing. Grit impact induced severe plastic deformation on the surface of all sample, resulting in the formation of distinctive wave-like patterns. Furthermore, because the grit had an angular shape, they plowed the samples surface during impact, and several debris particles were observed in localized surface regions before detachment. In a relative comparison, the surface roughness of the T525°C sample, which has the highest hardness, exhibited a smoother surface, and its debris size was also smaller. To further elucidate the wear mechanism, Cross-sectional microstructures beneath the most severely worn regions were characterized. Representative backscattered electron images are also shown in the same figure. In the cross-sectional microstructure of the as-cast sample, numerous cracks induced by grit impingement were observed inside the massive primary carbides immediately beneath the surface, as indicated by white arrows. Meanwhile, the network-like primary carbides, as indicated by black arrows, underwent deformation. A similar trend was observed in both the AsQ and T650°C samples. In regions without primary carbides, matrix cracking was observed beneath the surface, especially in the T625°C sample. By contrast, in the T525°C sample, which exhibited the highest hardness, although the grit impact caused cracking of the primary carbides, plastic deformation of the matrix and crack initiation within it were relatively limited. These results suggest that the erosive-wear resistance of the present cast steel may depend not only on the carbides but also on the hardness and hardening capacity of the matrix.

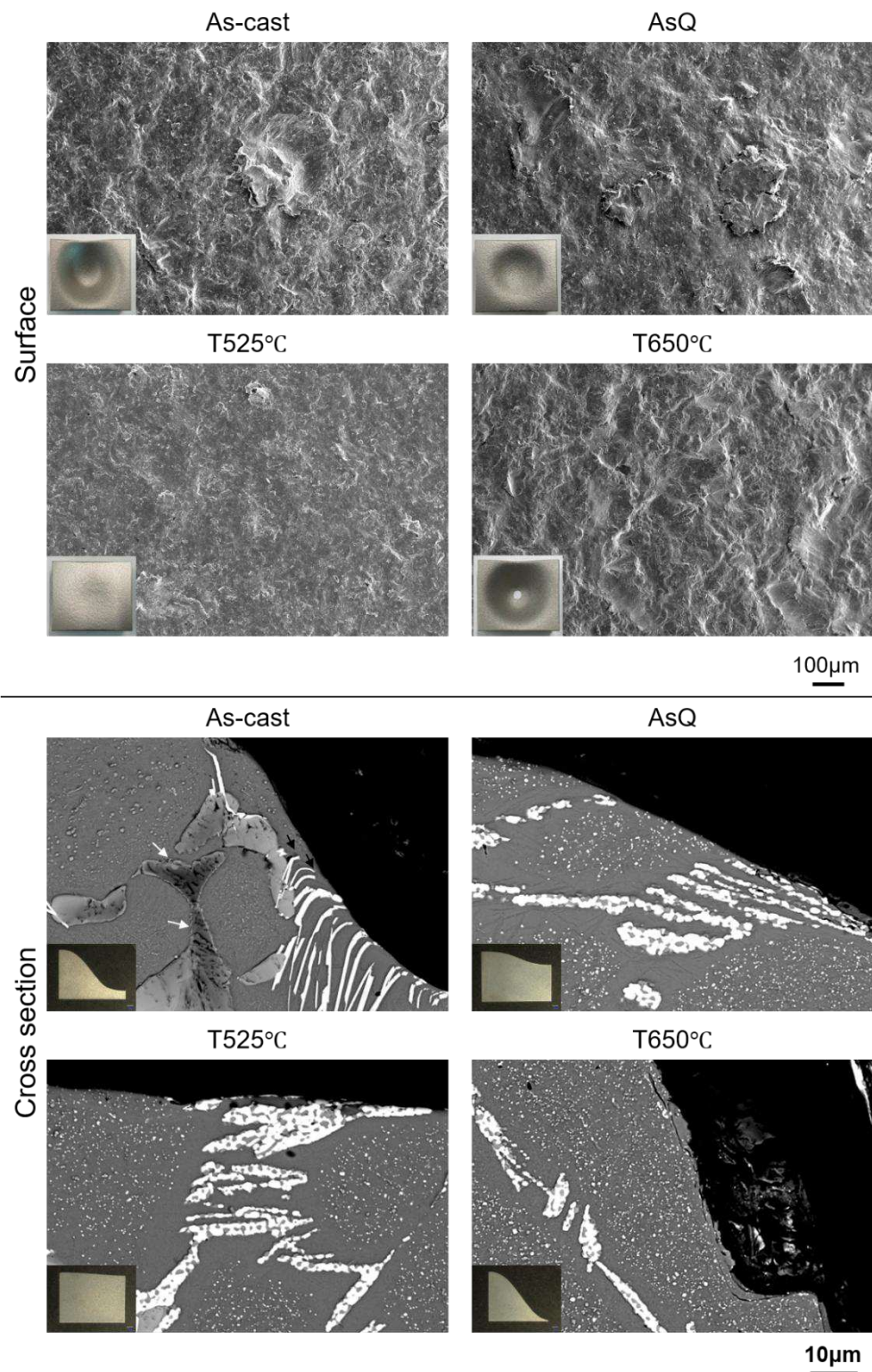


Figure 12 Worn surfaces and cross-sectional microstructures of samples after erosive-wear tests.

Figure 13 shows the hardness distribution along the depth from the worn surface for each sample, measured under a load of 2.94 N (0.3 kgf). Each value in the figure represents the combined hardness of the primary carbides and the matrix, which consists of martensite and retained austenite containing secondary carbides. Compared with the bulk regions, an increase in hardness at a depth of 0.05 mm from the surface, indicative of a surface-hardening phenomenon, was observed exclusively in the T525°C and AsQ samples. In contrast, the as-cast and T650°C samples exhibited little difference in hardness between the near-surface and bulk regions.

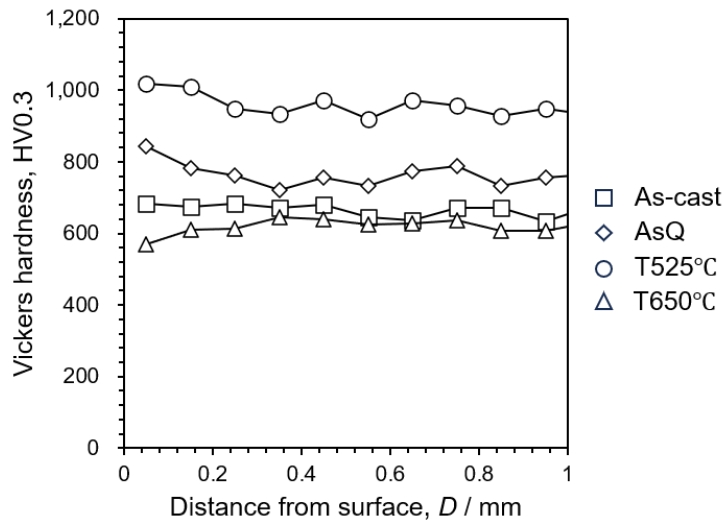


Figure 13 Hardness distribution from the worn surface in thickness-depth direction after erosive-wear tests.

To evaluate the matrix hardening capacity under each heat treatment condition, the matrix hardness containing secondary carbides was measured at 0.05 mm below the worn surface and in the center part of each sample under a load of 1.96 N (0.2 kgf). The results are shown in Figure 14. In the center part of the samples, the matrix hardness increased in the following order: T650°C sample < as-cast sample < AsQ sample < T525°C sample. The increment in matrix hardness for the AsQ sample was approximately 120 HV, whereas the increases for the T525°C and T650°C samples were around 55 HV, which were comparable to that of the as-cast sample.

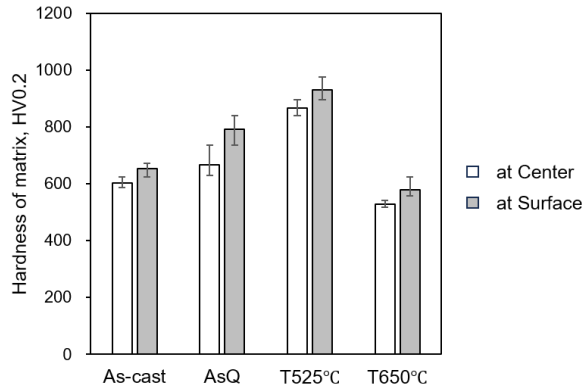


Figure 14 Increase in hardness of each sample due to erosive-wear tests.

In our previous work on high-Cr cast irons whose matrix consisted of a mixture of retained austenite and martensite, it was found that continuous high-velocity steel-grit impact induced a strain-induced martensitic transformation from retained austenite near the surface, alongside instantaneous self-tempering of the initial martensite caused by frictional heat (Ref. 22, 23). While metastable retained austenite enhances surface-hardening capacity and thereby improves erosive-wear resistance, a predominant martensitic matrix accelerates the wear rate due to its self-tempering effects. Furthermore, carbides within the matrix exert a minor direct contribution to this hardening behavior, which is primarily governed by phase transformation and dislocation accumulation. Building upon these findings, the present study extends the investigation to SKH57 martensitic cast steel. Given its unique matrix features, both the changes in retained austenite amount, as well as dislocation density near the surface after the wear test were evaluated via XRD analysis.

Figure 15 shows a comparison of the amount of retained austenite in the four samples before (initial state) and after the wear test. The initial retained austenite data were extracted from the results previously shown in Figure 3. The results demonstrate that grit impingement caused strain-induced martensitic transformation in both the as-cast and AsQ samples, with the transformed amount being approximately 3.5 vol.%. Conversely, the retained austenite amount in the T525°C and T650°C samples exhibited negligible changes after the test, indicating that no strain-induced martensitic transformation occurred.

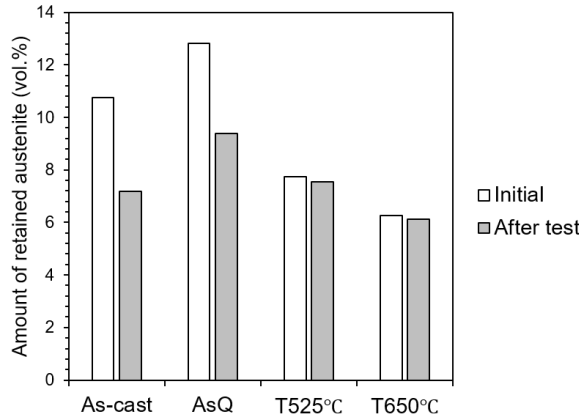


Figure 15 Change in amount of retained austenite due to erosive-wear tests.

Figure 16 shows the FWHM of the martensite (211) peaks for the four types of samples before and after the wear tests. For all specimens, prominent broadening of the martensite diffraction peak was observed following the wear tests. Furthermore, the increase in FWHM values after the test followed the order of T525°C sample < as-cast sample < AsQ sample < T650°C sample, which was the same sequence observed before testing (initial state). However, the increase in FWHM for the T525 °C sample was smaller than the that of other samples. As reported by Takaki et al., a dislocation density of $10 \times 10^{15} \text{ (m}^{-2}\text{)}$ is the typical saturation limit for deformed steels, and exceeding this threshold leads to dislocation annihilation via the mutual interaction of dislocations with opposite signs (Ref.20). In the present study, it is inferred that the near-surface dislocation density of the T525°C sample reached this critical threshold, thereby triggering partial dislocation annihilation through dynamic recovery.

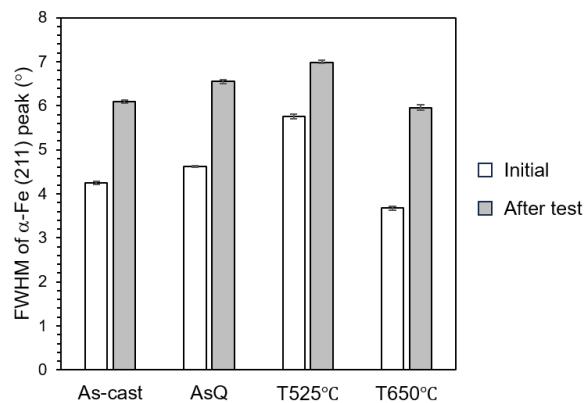


Figure 16 Change in FWHM of the martensite (211) peak resulting from erosive-wear tests.

Based on the above results, it can be concluded that the introduction of high-density dislocations near the surface during the wear test is the primary cause of the surface hardening. A higher degree of matrix hardening degree associated with the increased dislocation density effectively suppresses the plastic deformation of the matrix itself. Consequently, even if micro-cracks occur within the carbides during testing, their detachment is hindered due to the robust constraint provided by the surrounding matrix, which simultaneously suppresses the initiation and propagation of cracks within the matrix.

4. Conclusions

The microstructure, hardness, and erosive-wear resistance of SKH57 high-speed tool steel manufactured by a self-hardening mold casting process were investigated. The main findings are summarized as follows.

1. Due to the high-speed tool steel containing a large amount of various carbide-forming elements, primary carbides precipitate along the grain boundaries in the as-cast state, while secondary carbides are also formed within the matrix.
2. Unlike the tempering behavior of high Cr cast iron, significant secondary hardening appears at around 525 °C. This secondary hardening is attributed to the precipitation of fine secondary carbides and an increase in dislocation density. Above 525 °C, the steel exhibits high resistance to tempering softening, and the retained austenite decomposes with increasing tempering temperature.
3. The dislocation density within martensitic matrix plays an important role in suppressing detachment of primary hard carbide from sample surface under erosive-wear testing. Secondary hardening, associated with carbide precipitation as well as increased dislocation density, results in the superior wear resistance of the high-speed tool steel.

Appendix

Considering the overlap between the martensite (110) peak and the carbide (511) peak, the dislocation density in this study was evaluated solely based on the FWHM of the martensite (211) peak as a relative index. Supplementary Figure S1 displays the corresponding dislocation densities. The dislocation density was estimated under the assumption that micro-strain is the dominant contributor to peak broadening, rather than crystallite size refinement. Based on the calculated dislocation densities, the values for the as-cast, AsQ, and T650°C samples more than doubled compared to their initial states under the current wear test conditions. Conversely, while the T525°C sample displayed the highest dislocation density at $9.51 \times 10^{15} \text{ (m}^{-2}\text{)}$, its increase in dislocation density due to the wear test was found to be smaller than those of the other samples. It is inferred that the T525°C sample approached this critical threshold, implying that a partial dislocation annihilation might have occurred via recovery.

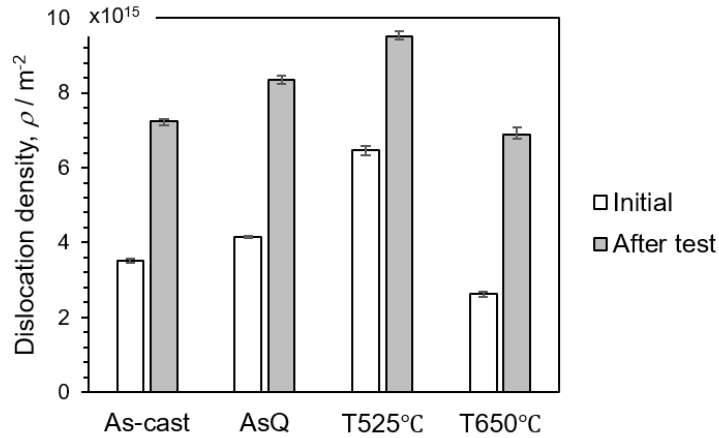


Figure S1 Estimated dislocation densities of the samples evaluated from XRD peak broadening. (Note: The dislocation density values presented in the figure are intended for relative comparison and qualitative trend analysis, bearing potential margins of error inherent to the single-peak approximation.)

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