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Effect of Tempering Temperature on Microstructure and Mechanical Properties of Bainitic Railway Wheel Steel with Thermal Damage Resistance by Alloy Design

Jiaqi Zhu ¹, Zhunli Tan ^{1,*}, Yu Tian ¹, Bo Gao ¹, Min Zhang ¹, Junxiang Wang ² and Yuqing Weng ¹

¹ School of Mechanical and Electronic Engineering, Beijing Jiaotong University, Beijing 100044, China; 16116343@bjtu.edu.cn (J.Z.); 17121354@bjtu.edu.cn (Y.T.); 15121316@bjtu.edu.cn (B.G.); zhangm@bjtu.edu.cn (M.Z.); weng@csm.org.cn (Y.W.)

² Technical Center, Tianjin Will Long Sci. & Tech. Co., Ltd., Tianjin 300385, China; jxwtr@163.com

* Correspondence: tzli@bjtu.edu.cn; Tel.: +86-1381-167-8974

Received: 6 July 2020; Accepted: 28 August 2020; Published: 10 September 2020

Abstract: Thermal damage is one of the principle modes of wagon railway wheels. A new bainitic railway wheel steel with high thermal damage resistance and good combination of strength, plasticity, and toughness was developed. Microstructure and mechanical properties of the new steels in a tempered condition at different temperatures were examined. Microstructures were observed using scanning electron microscope and transmission electron microscope. Mechanical properties were evaluated by tensile, hardness, and Charpy impact tests with a simultaneous comparison to pearlitic railway wheel steel. The characteristic of retain austenite and V(C,N) were measured through X-ray diffractometry and energy disperse spectroscopy. The results indicate that this new bainitic wheel steel presents a submicron-sized carbide-free bainite morphology and preferable integrated mechanical properties when tempered at 280–360 °C. Precipitation strengthening plays an important role for the high strength, since a two-time-strengthening mechanism of the yield strength led by precipitation has been found at 280–360 and 480–560 °C, respectively. Compared with a pearlitic railway wheel steel, bainitic wheel steel tempered at 320 °C has a 10% higher yield strength, five times higher impact toughness, and much better thermal damage resistance, which is a promising railway wheel material for high speed or heavier axle-load service conditions.

Keywords: bainite; railway wheel steel; tempering temperature; microstructure; mechanical properties; thermal damage resistance

1. Introduction

Wear, rolling contact fatigue (RCF), and thermal damage are principle failure modes of in-service railway wheels [1–3]. The present wheels mainly consist of ferrite and perlite microstructure. With the application of higher speed and heavier axle load, these failures happen more frequently and therefore, higher match of strength and toughness, better resistance to wear, RCF, and thermal damage is demanded for wheels to ensure safety and prolonged life [2,4]. Higher wear resistance is commonly achieved by improving strength. The present pearlitic wheel generally has a medium-high carbon content of 0.45–0.80 wt.% [5]. While it is noted that the lack of high toughness and relatively low austenitizing temperature make these kinds of steels vulnerable to thermal damage. Especially when railway wagon braking as well as transient high temperature in the contact zone between wheel and brake shoe can exceed the related austenitizing temperature, resulting in austenite formation in the wheel surface layer. Once the sliding stops, this local austenite rapidly cools down, forming a thin

layer of brittle martensite. This untempered brittle martensite is likely to cause cracking under thermal stress and cycling load [6,7]. Therefore, a new series of railway wagon wheel steels which own both higher toughness and better thermal damage resistance has drawn much attention.

Bainitic steels, demonstrating an excellent match of higher strength and toughness, have been widely studied and utilized in many industrial fields including railway rails and crossings [8–14]. Therefore, related researches are trying to employ a bainitic microstructure as promising candidates for a new generation of railway wheels recently [15]. Sharma et al. tested a bainitic railway wheel steel made from MAS2 steel. This bainitic steel shows better wear resistance than the conventional pearlitic railway wheel steel [1]. Zhang et al. evaluated a kind of carbide-free bainitic (CFB) railway wheel steel with superior mechanical properties, especially high impact toughness. Related results show that it could be used for 30 t axle-load wagon wheels [4,16,17]. Other researches indicate that higher yield strength and toughness of CFB can be obtained through tempering, although the tempering temperature corresponding to optimal properties depends on microstructures as well as the cooling process and sometimes is different from each other [18–21]. So far, the tempering temperature's effect on the microstructure and mechanical properties of CFB railway wheel steels has not been systematically investigated. Furthermore, its resistance to thermal damage has almost not been discussed. According to the forming mechanism of thermal cracks, improving the austenitizing temperature of wheel steel by alloy designing could be one effective method.

In this paper, a new CFB railway wheel steel with a high austenitizing temperature was developed by alloy design through simulation using the Thermo-Calc software. Samples in a similar shape and heat treatment process to the industrial wheel were fabricated. The effect of tempering temperature on the microstructure and mechanical properties were investigated systematically and the role of retained austenite (RA) and $V(C,N)$ in the tempering process were interpreted. Moreover, the improvement of thermal damage resistance was verified via the thermal fatigue test.

2. Alloy Design

The influence of alloy content on austenitizing temperature was calculated by the Thermo-Calc 2020 software (Thermo-Calc Software, Solna, Sweden). Based on our previous work, the basic chemical composition was set as 0.15C-1.8Mn-1.2Si-0.4Cr-0.3Ni (wt.%). The phase content of face-centered cubic (FCC) and body-centered cubic (BCC) was shown in Figure 1. FCC's volume fraction and BCC's become 1 and 0 at the A_{e3} temperature, respectively. The A_{e3} temperature variation with contents of different elements was given in Figure 2. It is shown that there is a nearly linear relationship between the A_{e3} temperature variation and the increased content of each element. The A_{e3} temperature goes up with Si and V increment, but goes down with the improvement of Mn, Mo, Ni content. In these elements, V has the strongest effect pushing the A_{e3} temperature upward. Based on the principle of obtaining higher austenitizing temperature, the chemical compositions of the new railway wheel steel are given in Table 1.

The carbon content of a bainitic wheel steel is about 0.2 wt.%, which will greatly increase the A_{c3} temperature compared to a traditional pearlitic wheel steel CL60 or CL70 with carbon content of 0.6 and 0.7 wt.%, respectively. The strength loss by restricting carbon content can be offset by the addition of solution strengthening elements such as Si, Mn, and Mo [2]. Moreover, Mn and Mo addition will effectively enhance hardenability and promote bainitic transformation, leading to a desired microstructure and uniform hardness. Si plays an important role in carbide-free bainite steel, which can retard the formation of cementite during cooling and tempering, thus allowing austenite to retain between bainite ferrite laths and increasing the tempering resistance [22–26]. Except for the toughness enhancement effect, Ni addition can compensate the latent negative effect of relatively high amount of Si or Mn addition on toughness [23,27,28]. The objective of V addition is to refine bainite plate size and achieve precipitation strengthening by $V(C,N)$ precipitation, simultaneously to enhance the austenitizing temperature [29–31].

Figure 1. Volume fraction of body-centered cubic (BCC) and face-centered cubic (FCC) at different temperatures.

Figure 2. A_{e3} temperature varies with increased content of different elements.

Table 1. Chemical composition of experimental wheel steel wt.%.

C	Mn	Si + Mo	Ni	V	P	S	Fe
0.22	2.2	1.5~2.2	0.4	0.1	0.010	0.006	Bal.

3. Experimental

3.1. Treatment

Tested samples were machined from a forged cylinder ingot with 840 mm diameter and 140 mm thickness, as shown in Figure 3. The continuous cooling transformation (CCT) diagram as well as the austenitizing temperature were measured using DIL805A thermodilatometry (TA Instruments, New Castle, DE, USA) with a sample of $\Phi 4$ mm $\times$ 10 mm. Samples were heated at a rate of 10 °C/s to 930 °C and then held for 10 min, followed by cooling down to room temperature at a rate from 0.05 to 50 °C/s. The A_{e3} temperature of employed bainitic wheel steel is 881 °C, much higher than the traditional pearlitic wheel steel CL60.

All the samples treated by austenitization at 930 °C for 5 h were cooled by water spraying on the surface for 5 min. After that, the samples were air cooled to room temperature and then tempered at 240–640 °C for 5 h.

Figure 3. Structure of the forged cylinder ingot and the sampling place.

3.2. Microstructure Characterization

Samples used for microstructure observation and retained austenite (RA) analysis were obtained at the position in 35 mm depth from the top surface (Figure 3). The microstructure was observed by scanning electron microscope (SEM, Zeiss EVO18, 20 kV, Tokyo, Japan) and transmission electron microscope (TEM, FEI TECNAI G20, 200 kV, Tokyo, Japan). The precipitation was characterized using energy disperse spectroscopy (EDS). SEM specimens were polished and etched with a 3% nital solution. TEM specimens were prepared using a twin-jet electropolishing instrument in 7% perchloric acid at around $-15\text{ }^{\circ}\text{C}$. The volume fractions of retained austenite were measured by X-ray diffractometry (XRD, $\text{CuK}\alpha$ radiation, Rigaku, Tokyo, Japan) with a 2θ ranging from 35 to 95° at a speed of 3° per min, and calculated through the direct comparison method through the integrated intensities of (200), (220), (311) peaks of austenite and those of (200) and (211) peaks of ferrite [15]. The carbon content χ_{C} of RA was calculated by the Nelson-Riley extrapolation method [32]:

$$\alpha_{\gamma} = 0.3556 + 0.00453\chi_{\text{C}} \quad (1)$$

where α_{γ} is the lattice parameter of austenite. The M_{s} of RA was estimated as [19]:

$$M_{\text{s}} = 539 - 423\chi_{\text{C}} - 30.4\chi_{\text{Mn}} - 12.1\chi_{\text{Cr}} - 7.5\chi_{\text{Si}} \quad (2)$$

where χ_{Mn} , χ_{Cr} , and χ_{Si} are the mass ratio of manganese, chromium, and silicon, respectively. The carbon content in the retained austenite is much higher than that in the prior austenite, so the corresponding M_{s} of RA is different from that of prior austenite.

3.3. Mechanical Properties

Samples for tensile and impact tests were gotten from a 35 mm depth position from the top surface. Tensile tests were executed on the universal testing machine using M12 standard specimens with a 1 mm/min loading rate. Charpy U-notch impact toughness was tested using $10\text{ mm} \times 10\text{ mm} \times 55\text{ mm}$ specimens at the temperature from $-60\text{ }^{\circ}\text{C}$ to room temperature ($20\text{ }^{\circ}\text{C}$).

Mechanical properties of the CL70 wheel were tested as a contrast.

3.4. Thermal Damage Resistance

Samples used in thermal fatigue tests were taken from the experimental bainitic wheel steel and commercial pearlitic wheel CL70 and the dimension is shown in Figure 4. Samples were heated to $600\text{ }^{\circ}\text{C}$ and then cooled to room temperature by water cooling. This process was repeated 20 and 50 times, respectively. The morphology and distribution of thermal cracks were observed by an optical microscope (OM) after derusting in a 10% HCl solution and slight polishing of the sample surface.

Figure 4. Dimension of samples for the thermal fatigue test (mm).

4. Results and Discussion

4.1. Microstructure

Figure 5 shows the CCT diagram of the experimental steel. Pearlite or proeutectoid ferrite will not form when the cooling rate is higher than 0.05 °C/s. The microstructure at the ingot surface consists of bainite and martensite, where the fastest cooling rate exists. While inside the ingot with a slower cooling rate, the microstructure is mainly composed of bainite.

Figure 5. The continuous cooling transformation (CCT) diagram of employed steel.

Figure 6 shows a typical microstructure of specimens tempered at different temperatures. The morphologies at a 320 °C tempered state is mainly carbide-free bainite. Fine lath-shaped bainite (LB) and granular bainite (GB) microstructures could be observed clearly (Figure 6a). When the tempering temperature was above 440 °C, the decomposition of RA became remarkable, and some of the bainite ferrite (BF) laths began to merge (Figure 6b). Once the tempering temperature was improved to 640 °C, RA almost completely dissolved and the recrystallization phenomenon could be observed, with plenty of precipitates distributed along the recrystallized grain boundaries (Figure 6c).

Figure 6. SEM microstructures of employed specimens tempered at different temperatures, (a) 320 °C, (b) 440 °C, (c) 640 °C.

TEM images of typical carbide-free bainite micrographs of 320 °C tempered specimens are shown in Figure 7. Obviously, the submicron-sized filmy RA (0.2 μm) was embedded within submicron-sized bainite ferrite laths (0.5–1 μm) (Figure 7a). Si addition can improve the tempering resistance and promote thermal stability of RA. As a result, the filmy RA and blocky RA existed together after 320 °C tempering in the size of less than 1 μm, as shown in Figure 7b,c. Fine BF laths and stabled filmy RA contribute to a high match of strength and toughness [15].

Figure 8 demonstrates the TEM micrograph and EDS analysis result of precipitates in the 480 °C tempered specimen. Much more particles are able to be easily observed, such as those pointed by arrows in Figure 8a. These fine and uniformly distributed particles exhibit an average size of 10–20 nm. The EDS results proved that these particles are V(C,N) (Figure 8b). It is noted that these kinds of particles can hardly be recognized in the 320 °C-tempered specimen.

Figure 7. TEM images of microstructures for the 320 °C tempered specimen. (a) carbide-free bainitic (CFB) lath, (b) bright field, and (c) dark field of film retained austenite (RA) and blocky RA.

Figure 8. (a) TEM images and the (b) EDS results of precipitates in the 480 °C tempered specimens.

The characteristic of retained austenite measured by XRD is listed in Table 2. The retained austenite volume before tempering is 13.7%, much higher than those in the tempered condition. This indicates that the tempering treatment is effective in RA's decomposition and transformation. The RA volume has no significant difference among the samples tempered from 240 to 400 °C, generally keeping within 6.5–8%. However, once the tempering temperature is enhanced to 440 °C, the corresponding RA content decreased sharply to about 2.5%. The thermal stability of retained austenite is mainly determined by its carbon content, and the carbon content maintained about 1.3% between 240~360 °C tempering treatments. The corresponding calculated Ms is nearly −80 °C. If the carbon content drops to 1% or even lower as the sample tempered at 440 °C or above, Ms will be under room temperature. This means that the unconverted retained austenite could also transform through the appropriate treatment.

In summary, the microstructure of designed wheel steel tempered at the temperature from 280 to 360 °C include a fine carbide-free bainite structure with 6.4~7.6% RA existing between bainite lath and inside GB. These RAs demonstrate an impressive thermal stability for their relatively high carbon content of 1.31~1.34%. There is no observation of Inter-lath carbide, which could enhance the damage resistance of the steel [33].

Table 2. The characteristic of retained austenite of the tested wheel steel.

Tempering Temperature/°C	Volume of Retained Austenite/%	Carbon content of Retained Austenite/%	Ms of Retained Austenite/°C
As-quenched	13.7	1.06	20
240	8.3	1.24	−56
280	6.6	1.34	−99
320	7.6	1.31	−86
360	6.4	1.31	−86
400	7.7	1.16	−23
440	2.5	1.03	32

4.2. Mechanical Properties

Mechanical properties of the samples at different tempering temperatures are shown in Table 3. Tensile strength, yield strength, hardness, and plasticity basically maintained a high level between tempering temperature of 240 to 560 °C. The tensile strength was about 1050 MPa, the hardness has a similar tendency, keeping within 33 to 35 HRC. While the yield strength demonstrated two peaks, at ~360 and ~560 °C, indicating that there existed a two-time-strengthening mechanism with the tempering temperature increment. The primary strengthening stage should be attributed to cementite precipitation, at about 360 °C which was not high enough for V(C,N) to precipitate. When the tempering temperature improved to 480 °C, fine V(C,N) could precipitate largely, and the steel's yield strength got 39 MPa enhancement compared with that tempered at 440 °C. The strengthening contribution of V(C,N) can be shown by the Ashby-Orowan equation as follows [34]:

$$\sigma_p = \frac{5.9 \sqrt{f}}{\bar{x}} \ln \left[\frac{\bar{x}}{2.5 \times 10^{-4}} \right] \quad (3)$$

where σ_p represents precipitation strengthening (MPa), f is the volume fraction of precipitates and $\bar{x}$ is the average precipitation size (μm). Here, f and $\bar{x}$ values at 480 °C tempered state can be obtained from Figure 7a. The corresponding calculated σ_p value is 40 MPa, which is close to the testing result. This indicates that V(C,N) precipitation is the main driving force of secondary strengthening stage, and this effect reached a maximum extent at ~560 °C. The elongation increased from 18% to 20% when the tempering temperature changed from 240 to 400 °C, while then fell down to ~17% at 440 and 480 °C. Notably, when the tempering temperature was up to over 560 °C, both tensile strength and yield strength declined obviously and meanwhile, the plasticity and toughness recovered as a

consequence of recrystallization. Therefore, this designed steel shows an excellent match of strength and plasticity when the tempering temperature is below 560 °C.

Table 3. Mechanical properties of the tested wheel steel.

Tempering Temperature/°C	Tensile Strength/MPa	Yield Strength/Mpa	Elongation to fracture/%	Hardness HRC	Impact Energy/J
240	1072	794	18.5	34	138
280	1050	814	18	32	170
320	1055	837	19	33	157
360	1051	843	19	35	144
400	1054	827	20	34	88
440	1090	754	17	33	59
480	1075	793	17	34	51
560	1047	858	19	33	53
640	921	786	21	28	80

As shown in Table 3, samples tempered at 240–360 °C exhibited excellent impact properties and reached a peak impact energy value of 170 J at 280 °C. High impact energy is related to film retained austenite existence, which played an important role in blocking crack propagation. However, when temperature was higher than 400 °C, the impact energy decreased rapidly, appearing obvious brittleness tendency. The valley value appeared at about 480 °C. This steel shows only one style of temper brittleness from 400 to 560 °C, different from the traditional two steps shown in common steels. Si bearing is the main reason for this phenomenon, which enhanced the temper resistance by suppressing the grain-boundary precipitation of cementite and thus retarded the appearance of first temper brittleness.

From Figure 9, the tempering temperature's effect on impact energy was similar to that on the carbon content in retained austenite. An approximately linear relationship between toughness and carbon content in RA was revealed in Figure 9b and the equation is as follows:

$$Ak_{u2} = 357\chi_C - 314 \quad (4)$$

where Ak_{u2} is the impact energy with a U2 notch and χ_C is the carbon content of RA.

Here, the toughness of CFB steel strongly depends on the stability of retained austenite, while this stability is determined by its carbon content. Since the unstable RA may transform into a brittle martensite, which has a negative effect on toughness, on the contrary, the stable RA with high carbon content inside is good to better impact energy. As a consequence, it is an effective way of controlling the carbon content of RA for desirable impact energy to choose the appropriate tempering process.

Figure 9. (a) The value and (b) the relationship of the impact energy and the carbon content of RA.

The mechanical properties comparison between designed bainitic railway wheel steel here and pearlitic CL70 is shown in Table 4. This new steel has a better match of strength and toughness, demonstrating a much higher yield strength and impact toughness, as well as better plasticity.

Table 4. Mechanical properties comparison of 320 °C tempered tested bainitic wheel steels and a commercial pearlitic railway wheel steel.

Microstructure Pattern of Wheel Steel	Tensile Strength/MPa	Yield Strength/Mpa	Elongation to Fracture/%	Hardness HRC	Impact Energy/J
CFB	1055	837	19	33	157
Pearlitic (CL70)	1190	761	16	28	26

Figure 10 shows the low temperature impact energy of the sample tempered at 320 °C. The impact energy decreased continuously with the test temperature. Not like a typical curve, no ductile-brittle-transition phenomenon was observed from the curve in Figure 10. The impact energy was higher than 50 J even at −60 °C, and samples presented excellent low temperature toughness. It has been reported that DBTT tends to increase with the increasing GB volume fraction and its grain size [35]. Therefore, improved hardenability and submicron-size structure of this designed wheel steel may be the primary cause of its higher low-temperature properties.

Figure 10. Low temperature impact energy of 320 °C-tempered wheel steel.

The tempering temperature has a great effect on microstructure and mechanical properties. Commonly, a desirable combination of strength and toughness should be achieved by adopting specific tempering temperatures according to different demands. However, a high match of strength and toughness can also be obtained through a board range of tempering temperature with proper alloy and microstructure design, as described in the article. For railway wheels in service, the crack is fatal to the safety. Therefore, toughness and RCF resistance become the most important properties. Therefore, to ensure a high toughness of the steel, the tempering temperature must be controlled to avoid temper brittleness which starts with 400 °C. Moreover, the tempering temperature should be higher than 320 °C to obtain a higher yield strength and plasticity. Relatively high tempering temperature may be helpful to wear resistance as well, relying on more second-phase particles precipitated.

TEM results indicate that substantial precipitation of V(C,N) requires the tempering temperature higher than 480 °C. Within the proper tempering temperature range of railway wheel, precipitation strengthening has not been fully utilized. Even so, V is still helpful on lifting the A_{e3} temperature and the enhancement of overall properties with even a relatively low content.

4.3. Thermal Damage Resistance

Figure 11 shows OM examinations of samples after thermal fatigue tests. From Figure 11a,b, uniformly distributed microcracks with a consistent propagating direction were found in both pearlitic

and bainitic wheel steels after 20 times cycling. However, the cracks of pearlitic wheel were much more and denser. When the thermal fatigue cycles increased to 50 times, cracks in pearlitic wheel propagated along the length direction and merged into wide cracks. A severe crack coalescence pearlitic wheel can be seen obviously in Figure 11c. While the cracks in bainitic wheel steel only increased in number and propagated limitedly as shown in Figure 11d, no widening was similar to that in pearlitic wheel. This indicates that the thermal damage resistance of the new designed bainitic wheel steel is better than pearlitic wheel. Recycled heating and cooling will cause repeated plastic deformation, leading to initiation propagation of fatigue cracks [36]. The pearlite microstructure consists of a soft ferrite phase and hard cementite phase and cracks are likely to initiate from ferrite whose plastic deformation is restricted by the surrounding cementite [37]. However, in the carbide-free bainite microstructure, the filmy RA in bainitic wheel can prevent cracks from initiation and propagation due to its prominent compatible deformation capability to some extent [38]. Therefore, the crack number in bainitic wheel is less than that in the pearlitic wheel after 20 times cycling and only developed in number but not so extremely merged as in pearlitic wheel even up to 50 times of thermal cycle.

Figure 12 shows the microstructure of bainitic sample after 20 and 50 times thermal cycles. It can be seen that there is no evident change on the microstructure of LB and GB except slight merging of bainite laths during thermal fatigue. This indicates that the employed bainitic steel has a high microstructure stability during thermal fatigue.

Figure 11. OM examinations of (a,c) pearlitic wheel and (b,d) bainitic wheel steel after (a,b) 20 times and (c,d) 50 times thermal cycles.

Figure 12. SEM images of bainitic wheel steel after (a) 20 times and (b) 50 times thermal cycles.

5. Conclusions

A new carbide-free bainitic wheel steel with V addition was developed. The effect of tempering temperature on the microstructure and mechanical properties of this new developed wheel steel were studied. The conclusions were drawn as follows:

1. The new bainitic wheel steel achieved excellent integrated mechanical properties at the tempering temperature range of 280–360 °C.
2. The tested steel has a two-time-strengthening mechanism of yield strength at 280–360 °C and 480–560 °C, respectively. The latter one is led by fine V(C,N) precipitation.
3. This new bainitic wheel steel shows a much higher match of strength and toughness at a proper tempered state than commercial pearlitic railway wheel steel C170, especially more excellent low temperature toughness.
4. The V bearing bainitic wheel steel has better thermal damage resistance than the commercial pearlitic wheel.

Author Contributions: Conceptualization, Z.T. and Y.W.; methodology, J.Z.; validation, Z.T., M.Z. and Y.W.; formal analysis, J.Z.; investigation, B.G.; resources, J.W.; data curation, J.Z.; data writing—original draft preparation, J.Z. and Y.T.; writing—review and editing, J.Z.; visualization, J.Z.; supervision, Y.T.; funding acquisition, Z.T. All authors have read and agreed to the published version of the manuscript.

Funding: This work was supported by the Fundamental Research Funds for the Central Universities, grant number 2019JBM103.

Conflicts of Interest: The authors declare no conflict of interest.

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