Article

Relationship between Microstructure and Properties of 1380 MPa Grade Bainitic Rail Steel Treated by Online Bainite-Based Quenching and Partitioning Concept

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Abstract: According to the concept of the bainite-based quenching and partitioning (BQ&P) process, we designed the online heat treatment routes of bainitic rail steel for heavy haul railway. The new heat treatment process reduced the fraction and size of the blocky martensite/austenite (M/A) islands formed during the conventional air-cooling process. The M/A islands are coarse and undesirable for mechanical properties. A new kind of 1380 MPa grade bainitic rail steel with more uniform microstructure and better mechanical properties was produced by the online BQ&P process. We characterized the multiphase microstructures containing bainite, martensite, and retained austenite of 1380 MPa grade bainitic rail steels via optical microscope, scanning electron microscopy, transmission electron microscopy, and X-ray diffractometer. We investigated in-depth the relationship between the microstructure, retained austenite stability, and mechanical properties, particularly the resistance to wear and rolling contact fatigue, of the new 1380 MPa grade bainitic rail steels. Meanwhile, the conventional air-cooling bainitic rail steel was studied as a comparison.

Keywords: bainitic rail steel; microstructure; retained austenite; wear; fatigue

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1. Introduction

Railway transportation is one of the important modes of transportation in China, supporting the development of the national economy. Rail steel is the key component that affects the safety and efficiency of railway transportation. At present, high carbon pearlitic steels are used widely as rail material [1]. The development of high-speed and heavy-haul railways in China results in an increase in traffic density, train speed, and axle load. The current pearlitic rail steels are faced with more and more serious damage problems, such as wear, rolling contact fatigue (RCF), and catastrophic fracture [2]. The failure of pearlitic rail steels could threaten the safety of train operation and increase the maintenance cost of the railway. Therefore, it is necessary to develop a new generation of rail materials with excellent performance.

Compared with conventional pearlitic rail steels, low-alloyed bainitic rail steels have a better combination of strength and toughness, wear resistance, and RCF resistance [3–5]. In recent decades, many research institutions all over the world have been committed to the research and exploration of bainitic rail steels [3–10].

In the 1990s, Clayton et al. [3] studied the rolling contact fatigue of pearlitic and bainitic rail steels and pointed out the potential of bainitic steels for the railway. Bhadeshia et al. [4] reported new carbide-free bainitic rail steels with superior wear and RCF resistance. It is reported that the absence of brittle cementite particles contributed to the improved toughness and RCF resistance. Furthermore, the high toughness of the bainitic rail steels

could improve wear resistance by reducing the generation of wear debris [4]. Yokoyama et al. [5] developed high-strength bainitic rail steels and found that the bainitic rail steels had at least twice the RCF resistance and almost equal wear resistance to those of the heat-treated pearlitic rail steels. The brittle white etching layer was difficult to form on the low carbon bainitic rail steels compared to high carbon pearlitic rail steels. Sawley and Kristan investigated the performance of a low carbon bainitic rail steel (i.e., J6 rail steel) via small- and full-scale tests [6]. They found that the RCF performance of bainitic rail steel appeared to be superior to that of pearlitic rail steel. Furthermore, the increased toughness of bainitic rail steel could increase the critical crack size before a sudden fracture. However, they also pointed out that the wear performance of bainitic rail steel exhibited mixed results from the small- and full-scale tests. Pacyna et al. [7] reported two grades of new bainitic steels with different head hardness, namely RB370 and RB390, where RB390 bainitic rail steel was designed for turnouts and railway frogs and was successfully used for heavy-haul lines. In contrast, Voestalpine Schienen developed a novel high carbon low-alloyed bainitic rail steel with high hardness up to 450 BHN, showing high resistance to the wear and RCF damage [8].

Recently, Kumar et al. [9,10] designed a novel high-strength steel with a fine carbide-free bainitic microstructure for railway crossings applications. The mechanical properties of the newly designed bainitic steel were found to be superior to those of the conventional steels used in railway crossings. The bainitic steel offered controlled crack growth under impact fatigue, which is the main cause of failure in crossings. Hasan et al. [11] designed two new carbide-free bainitic rail steels with minimal alloying additions and processed them to achieve high strength and toughness for application in heavy-haul rail tracks. The results found that the carbide-free bainitic steels provided superior mechanical properties in comparison with the conventional 880 MPa grade pearlitic rail steel and other bainitic steels of similar chemical composition.

Many works show that the strength and toughness, wear-resistance, and fatigue resistance of bainitic rail steels are higher than pearlitic rail steels [3–11]. In addition, bainitic rail steels also have excellent corrosion resistance. Moon et al. discussed [12] the corrosion behavior of newly developed bainitic steels made by isothermal heat treatment. They found that the modified chemical composition, finer microstructures, and compact rust morphology contributed to the better corrosion resistance of bainitic steels.

Since the 1990s, the bainitic rail steels have been developed rapidly in China. The relationships between the alloying, microstructure, process, and performance of bainitic rail steels have been investigated widely via in-lab small-scale and in-field full-scale tests [13–18]. The trial applications of bainitic rail steel on railway crossings and curved tracks have been carried out successfully. It has been found that bainitic rail steels show great potential in improving the service life of rail compared with heat-treated pearlite rails (340~370 HB).

The current bainitic rail steels are mostly produced via the process of natural air cooling plus low-temperature tempering after hot rolling. However, the cooling rate during natural air cooling varies with seasonal changes, resulting in the non-homogeneity in microstructure and performances of bainitic rail steels, especially the fluctuation of wear and RCF resistance [14]. Therefore, it is necessary to develop a new generation of online heat treatment processes for the bainitic rail, aiming to accurately control the microstructure and performances of bainitic rail steel. The optimization of the microstructure and mechanical properties via online heat treatment could, furthermore, improve the resistance to wear and fatigue and prolong the service life of bainitic rails.

In this study, we designed the online heat treatment routes of bainite rail steel for heavy haul railway according to the concept of the bainite-based quenching and partitioning (BQ&P) process [19]. We investigated the microstructure, stability of retained austenite, and mechanical properties, particularly the resistance to wear and RCF, of the new bainitic rail steels. The conventional air cooling bainitic rail steel was also studied as a comparison.

2. Experimental Procedure

The chemical composition of the bainitic rail steel is listed in Table 1. The carbon content of the rail steel is much less than the pearlitic rail steels (generally, 0.7~0.9 wt.%) [1,2]. The bainitic steel is melted, forged, and hot-rolled to rail at a steel mill.

Table 1. Chemical composition of the bainitic rail steel.

Sample	C	Mn	Si	Cr	Ni	Mo
Bainitic rail	0.2	2.2	1.0	1.0	0.5	0.3

Two different heat treatment processes are employed on the rail after hot rolling, namely the conventional naturally air-cooling process and online heat treatment based on the BQ&P concept. The schematic of the microstructural evolution during the two heat treatment processes is shown in Figure 1. In general, vast heat is generated during the bainitic transformation, i.e., phase transformation latent heat. Hence, the rail is cooled at an extremely low cooling rate once the temperature is below the B_s ($\sim 400^\circ\text{C}$), where the bainitic transformation is triggered. The untransformed austenite could transform to martensite when the bainitic transformation is ceased and the temperature is below the M_s (martensitic transformation start temperature, $\sim 340^\circ\text{C}$). This process is also similar to the bainitic austempering process, as indicated in Figure 1a. In contrast, the online BQ&P heat treatment is designed as follows, as indicated in Figure 1b. The railhead surface is cooled to a temperature below M_s ($\sim 280^\circ\text{C}$) at an accelerated cooling rate of $\sim 4^\circ\text{C/s}$ (via controlled spray cooling), but the cooling rate is lower than the critical quenching rate to form fully martensite. Hence, a part of parent austenite is transformed to bainite and martensite during the accelerated cooling step. Then, the accelerated cooling is stopped, which leads to the elevation of temperature on the railhead because of the presence of temperature gradient. The dynamic partitioning could take place during the final slow cooling step. The finite element simulation of the temperature change at the railhead can be found in our previous work [20]. Here, the bainitic rail steels treated by the conventional natural air cooling and online BQ&P processes are called 1280 G and 1380 G rail steel, respectively. Finally, both 1280 G and 1380 G rail steels are tempered at 280°C .

Figure 1. Schematic of the microstructural evolution of bainitic rail steels during natural air-cooling process and on-line heat treatment based on BQ&P concept. (a) Air-cooling process for 1280 G rail steel; (b) on-line BQ&P for 1380 G rail steel; A: austenite, B: bainite, M: martensite.

The specimens for microstructural characterization were cut from a depth of 10 mm from the railhead surface. The microstructure was characterized by optical microscopy (OM, Zeiss Scope.A1) and scanning electron microscopy (SEM, Zeiss EVO18, 20 kV) after mechanical polishing and etching in a 2% nital solution. The ultrafine retained austenite or carbide was investigated by a field emission transmission electron microscope (TEM, JEOL 2010F, 200 kV) using the thin foil samples electrolytically polished at -40°C in 4% perchloric acid solution. The volume fraction and lattice parameter of retained austenite were estimated by X-ray diffractometer (XRD, Rigaku Smartlab, Cu $K\alpha$ radiation) at a step

of 0.01° and a counting time of 2 s/step over the 2θ range of $35\text{--}95^\circ$. The specimens used for XRD measurement were prepared by mechanical grinding and electrolytic polishing in order to remove the deformed layer near the specimen surface. The retained austenite fraction was estimated by collecting the peak intensities of $(200)_\gamma$, $(220)_\gamma$, $(311)_\gamma$, $(200)_\alpha$, and $(211)_\alpha$. The austenite lattice parameter, a_γ , was obtained by Nelson–Riley extrapolation method. The carbon concentration, x_C , of retained austenite can be estimated using Equation (1) [21]:

$$a_\gamma = 3.556 + 0.0453x_C^\gamma + 0.00095x_{Mn}^\gamma + 0.0056x_{Al}^\gamma \quad (1)$$

where the austenite lattice parameter, a_γ , is in Å, and x_C^γ , x_{Mn}^γ , and x_{Al}^γ are the concentrations of carbon, manganese, and aluminum in austenite, respectively, in wt.%.

The specimens for the mechanical properties test (including tensile, U-notch impact, and fracture toughness tests) were prepared according to the standard of TB/T2344-2012. Three specimens were tested for each condition and the average values were recorded. In order to investigate the mechanical stability of retained austenite, the interrupted tensile test was carried out, and the change of retained austenite fraction as engineering strain was measured.

The wear test was carried out under dry wear conditions on the GPM-30D twin-disc wear testing machine (Yihua, China), as shown in Figure 2. The rail specimens for upper discs were cut from the depth of 0–20 mm from the railhead surface. The wheel specimens as lower discs pair were CL70 pearlitic wheel steel treated by normalizing. The CL 70 pearlitic wheel steel has a carbon content of ~0.7 wt.% and has been used in the freight train. The mechanical properties of CL70 steel can be found in Ref [22]. Both the rail and wheel disc specimens have the same outer diameter (60 mm). The other dimensions of the disc specimens can be found in Ref [23]. The twin-disc specimens were contacted in a line during the wear test. In this study, the rotational speed of the rail disc was set to 500 rpm, while that of the wheel disc was set to 497 rpm. In this case, the slip ratio is about 0.6%, as calculated by the methods in Ref [23]. It is reported that the nominal contact stress is ~650 MPa when the train load is 23 tons [17,24]. However, the maximum contact stress can exceed the nominal value during serious operation. Hence, in this study, the load of 2000 N was applied to the disc specimens, which generated a maximum Hertzian contact pressure of ~980 MPa. In order to investigate the RCF cracks initiation and propagation during cyclic contact, the rolling cyclic numbers were set to 1×10^5 , 2.5×10^5 , 5×10^5 , and 7.5×10^5 cycles, respectively.

Figure 2. Photograph of the GPM-30D twin-disc wear testing machine.

After the wear test, the wear mass loss was measured using an electronic scale with a resolution accuracy of 0.0001 g. The worn surface of rail specimens was characterized using SEM (Zeiss EVO18, 20 kV). The longitudinal-section samples were cut along the rolling direction and etched in 2% nital solution after mechanical polishing. The distribution of hardness of the samples was measured by a Vickers hardness tester (HVS-1000) under a

load of 25 gf and a dwell time of 10 s. The relationship between microstructure, plastic deformation, and micro-cracks was investigated using SEM (Zeiss EVO18, 20 kV).

3. Results and Discussion

3.1. Microstructure

Figure 3 shows the OM and SEM images of the 1280 G and 1380 G microstructures at the depth of 10 mm from the railhead surface. The OM image in Figure 3a shows that there are banded structures in the 1280 G samples, which is attributed to the micro-segregations of Mn, Si, Cr, etc., in the steel, as discussed in Ref [17]. However, the banded structures are not found in the 1380 G samples after online BQ&P treatment, as shown in Figure 3c. Forouzan et al. also reported that the Q&P process could minimize the micro-segregation induced banded structures via controlling the quenching temperature [25]. The bainite in the 1280 G sample is mainly granular bainite, as shown in Figure 3b. The granular bainite consists of coarse bainitic ferrite, M/A constituent, and film-like retained austenite. As indicated in Figure 3d, the microstructure of the 1380 G rail steel is also a multiphase of bainite and martensite. However, the bainite in the 1380 G samples exhibits lath or leaf morphologies, with smaller sheaf sizes. The accelerated cooling could reduce the transformation temperature, which leads to the bainitic transformation at a lower temperature range and subsequent refined bainitic microstructure. The martensite formed during the accelerated cooling step could provide the carbon atoms to the adjacent untransformed austenite during the dynamic partitioning [26]. The carbon partitioning from martensite to austenite could result in the formation of “carbon-depleted martensite” and carbon-enriched retained austenite [27].

Figure 3. Microstructure of 1280 G and 1380 G rail steels. (a,b) 1280G; (c,d) 1380G; (a,c) OM images; (b,d) SEM images; B: bainite, M: martensite, BF: bainitic ferrite, M/A: martensite/austenite, RA: retained austenite.

Figure 4 shows the TEM images of the 1280 G and 1380 G microstructures at the depth of 10 mm from the railhead surface. It is found that the width of the bainite lath in the 1280 G sample is approximately 2 μm. In contrast, the width of the bainite lath in the 1380 G

sample is a sub-micron meter. Meanwhile, the film-like retained austenite can be found between the bainite laths.

Figure 4. TEM images of the microstructure of 1280 G and 1380 G rail steels. (a) 1280G; (b) 1380G; BF: bainite ferrite, M: martensite.

In order to investigate in detail the constituents of bainite and martensite in 1380 G rail steel, the TEM analysis was carried out, as indicated in Figure 5. It is found that the microstructure of the 1380 G rail steel is more complex. The multiphase structure is considered to be beneficial to the mechanical properties. The bainite mainly exhibits lath morphology with the presence of inter-lath film-like retained austenite, as shown in Figure 5a,b. The width of retained austenite is tens of nanometers. As indicated in Figure 5c,d, the lower bainite is also observed in the 1380 G bainitic rail steel. The carbide is distributed within the bainitic plates and located along a certain angle. Meanwhile, carbon-depleted martensite is found in the 1380 G steel. The martensite has a straighter lath boundary and smaller lath width, compared with bainite, as shown in Figure 5e. Furthermore, the film-like retained austenite can be found between the carbon-depleted martensite laths, which can be attributed to the carbon partitioning from martensite to the untransformed austenite during the dynamic partitioning process [26,27].

3.2. Mechanical Stability of Retained Austenite

It is reported that the retained austenite with appropriate stability is beneficial to the ductility, toughness, and fatigue of high-strength steels [28,29]. Hence, we compared the stability of retained austenite in the 1280 G and 1380 G rail steels in this section. Table 2 lists the retained austenite amount and its carbon content in the 1280 G and 1380 G rail steels. It is found that the volume fraction of retained austenite is ~7.5% in the untempered 1280 G steel. The presence of retained austenite is attributed to the carbon enrichment in untransformed austenite during bainitic transformation. The retained austenite amount decreases to ~5.6% in the tempered 1280 G steel. In contrast, the retained austenite amount is ~6.6% and ~4.9% for the untempered and tempered 1380 steel, respectively. The carbon content in RA in the 1380 G steel is slightly lower compared with the 1280 G steel. It may be attributed to the formation of carbide during the dynamic partitioning process, which is considered as the competition process of carbon enrichment in retained austenite [30].

Figure 5. TEM images of microstructure in 1380 G rail steels. (a) bright field image showing lath bainite with inter-lath film-like retained austenite, where the inset image is the selected area electron diffraction patterns; (b) dark field image showing retained austenite; (c,d) carbide in lower bainite; (e,f) martensite lath and inter-lath film-like retained austenite.

Table 2. Retained austenite amount (RA %) and its carbon content (C %) in the 1280 G and 1380 G rail steels before or after tempering, determined by X-ray diffraction.

Sample	RA % (Untempered)	C % in RA (Untempered)	RA % (Tempered)	C % in RA (Tempered)
1280 G	7.5 vol.%	1.34 wt.%	5.6 %	1.39%
1380 G	6.6 vol.%	1.14 wt.%	4.9%	1.17%

In order to compare the mechanical stability of the retained austenite in the two rail steels, we carried out the interrupted tensile test and measured the volume fraction of retained austenite after subjecting different strains. Figure 6 shows the change of retained austenite amount versus the engineering strain for the 1280 G and 1380 G rail steels. It is found that the retained austenite amount decreases gradually with the true strain for all the samples. The relation between the fraction of retained austenite and the true strain, ϵ , can be presented by the exponent decay law [19,31]:

$$f = f_0 \exp(-k * \epsilon) \quad (2)$$

where k is the constant, revealing the overall mechanical stability of retained austenite during deformation. A lower k indicates the higher mechanical stability of retained austenite. The fitting lines are shown in Figure 6 as well. It is found that the mechanical stability of retained austenite shows improvement in the tempered 1280 G steel. The retained austenite in the 1380 G steel shows high stability even in the untempered state, which is comparable to the tempered 1280 G steel. The stability of retained austenite could be further improved in the 1380 G steel after tempering. As indicated in Figures 4 and 5, the retained austenite mostly exhibits film-like morphology and has a smaller size in 1380 G steel compared with the coarse blocky M/A in 1280 G steel. Hence, the retained austenite in the 1380 G steel has better mechanical stability despite lower carbon content in retained austenite.

Figure 6. Change of retained austenite amount with engineering strain for 1280 G and 1380 G rail steels. Note: the errors of all the results obtained by XRD are within $\pm 5\%$.

3.3. Mechanical Properties

Table 3 lists a comparison of the conventional mechanical properties of the tempered 1280 G and 1380 G rail steels. It is found that the 1380 G steel has higher strength and toughness compared to the 1280 G steel. The tensile strength of the 1380 G steel is above 1380 MPa and has an elongation of $\sim 15\%$, showing an excellent balance of strength and ductility. Furthermore, the fracture toughness of the 1380 G steel is significantly improved. The increased toughness could increase the critical crack size before a sudden fracture and ensure the safety of train operation [4]. The improved strength and toughness are attributed to the refined microstructure and high stability of the retained austenite in the 1380 G rail steel.

Table 3. Strength and toughness of the 1280 G and 1380 G rail steels.

Sample	R_m , MPa	R_p , MPa	A, %	A_{KU} , J	K_{IC} at -20°C , $\text{MPa}\cdot\text{m}^{1/2}$
1280 G	1335 ± 8	1208 ± 12	14.5 ± 0.5	79 ± 2	77 ± 12
1380 G	1390 ± 5	1263 ± 10	15.0 ± 0.5	104 ± 1	107 ± 4

R_m : tensile strength, R_p : yield strength, A: elongation, A_{KU} : U-notched impact toughness, K_{IC} : fracture toughness.

3.4. Wear Resistance of Bainitic Rail Steel

The wear resistance of the 1280 G and 1380 G bainitic rail steels is compared through a twin-disc wear testing machine. Figure 7 shows the wear mass loss of two bainitic rail

steels with rolling cycles. It can be seen that the wear resistance of the 1380 G rail steel is much better than the 1280 G variant, especially at the late stage of wear. For instance, the wear mass loss of the 1280 G steel (~ 1.03 g) is almost twice that of the 1380 G steel (~ 0.59 g) when the cyclic number reaches 7.5×10^5 .

Figure 7. Change of wear mass loss of 1280 G and 1380 G rail steels with rolling cycles.

The worn surfaces of the 1280 G bainitic rail steel after different rolling cycles are shown in Figure 8. The worn surfaces exhibit typical fatigue wear characteristics at the early stage. The large contact and shear stress cause severe plastic deformation, which promotes the preferential formation of parallel fatigue cracks on the sample surface [32]. Moreover, the cracks could grow to the subsurface layer with the increase in rolling cycles. After 1.0×10^5 cycles, the worn surface exhibits a large number of micro-cracks or peelings, as shown in Figure 8a,b. The cracks and peelings tend to be arrayed after 2.5×10^5 cycles. The parallel micro-cracks distribute at intervals, and the size appears to be larger (Figure 8c,d). The number of micro-cracks and peelings decreases after 5.0×10^5 cycles (Figure 8e,f). However, the micro-cracks start to connect with each other, which promotes peeling. Consequently, the wear debris is found on the worn surface as the rolling cyclic numbers increased to 7.5×10^5 (Figure 8g,h). In this case, the wear mechanism will dominate the wear loss of rail steels without severe fatigue cracking.

Figure 9 shows the worn surface of the 1380 G bainitic rail steel under different rolling cycles. After 1.0×10^5 cycles, a small number of micro-cracks and peelings are also observed on the worn surface (Figure 9a). The number of micro-cracks and peelings decreases obviously when the cycles exceed 2.5×10^5 (Figure 9b–d). This indicates that the wear mechanism will dominate the wear loss of rail steels with the increased cyclic numbers, which is similar to the 1280 G rail steel. Meanwhile, it is found that the number of micro-cracks is much less in the 1380 G rail steel compared with the 1280 G rail steel. In other words, the 1380 G rail steel has a higher resistance to fatigue crack initiation compared with the 1280 G rail steel.

The hardness distributions from the worn surface to the substrate of two bainitic rail steels are shown in Figure 10. It can be seen that the hardness gradually decreases and finally plateaus with the increase in depth. The increased hardness at the top surface is mainly attributed to the plastic deformation under the contact of the rail and wheel, as indicated in Figure 11. Hence, we also find the depth of the plastic deformation layer from the hardness distribution. The depth of the plastic deformation layer of the 1280 G rail steel is ~ 80 μm at the 1.0×10^5 cycles and 2.5×10^5 cycles and increases to more than 100 μm as the cycles increase to above 5.0×10^5 cycles. In contrast, the depths of the plastic deformation layers are all at the range of 50 μm to 60 μm for the 1380 G rail steel. The depth of the plastic deformation layer reflects the resistance to the plastic deformation of the microstructure. It can be seen that the 1380 G rail steel exhibits higher resistance to plastic deformation under wheel/rail contact compared with the 1280 G rail steel.

Figure 8. Worn surface of 1280 G rail steel under different rolling cycles. (a,b) 1.0×10^5 cycles; (c,d) 2.5×10^5 cycles; (e,f) 5.0×10^5 cycles; (g,h) 7.5×10^5 cycles.

Figure 9. Worn surface of 1380 G rail steel under different rolling cycles. (a) 1.0×10^5 cycles; (b) 2.5×10^5 cycles; (c) 5.0×10^5 cycles; (d) 7.5×10^5 cycles.

Figure 10. Hardness distribution of worn samples with the depth from top surface. (a) 1280 G rail steel; (b) 1380 G rail steel.

Meanwhile, it is found that, for the two kinds of rail steels, the hardness of the surface layer is higher at the early stage of the wear stage but decreases slightly with the increase in cycles. This may be attributed to the tempering of the top surface layer caused by the temperature rise during the wheel/rail contact [33].

Figure 11 shows the microstructure of the longitudinal section of the 1280 G and 1380 G rail steels after different rolling cycles. It is found that the microstructure is elongated or fragmented near the contact surface due to the plastic deformation, as shown by the SEM images in Figure 11. This also confirms that the depth of the plastic deformation layer is larger in the 1280 G rail steel compared to the 1380 G rail steel. The micro-cracks are prone to initiate at the surface and grow to the sub-surface in the 1280 G rail steel (Figure 11a). The cracks change the propagation path with the increase in rolling cycles, as indicated in Figure 11b, which could lead to the peeling. In contrast, the micro-cracks are rarely found in the 1380 G samples. This is also consistent with the results in Figures 8 and 9. Meanwhile, we can see that the plastic deformation is not uniform in the 1280 G worn samples, as indicated by the yellow dashed lines in Figure 11b. This is attributed

to the segregation banded structure in the parent microstructure of the 1280 G rail steel (as shown in Figure 3a). Recently, Zhang et al. [17] reported that the cracks propagated along with the segregation band under cyclic load, so segregation promoted the initiation and propagation of fatigue cracks and reduced the service life of the rail. Compared with the 1280 G rail steel, the microstructure in the 1380 G rail steel is more homogeneous and refined (Figure 3c,d), which could suppress the fatigue crack initiation during wheel/rail rolling contact.

Figure 11. Microstructure of the longitudinal section of the 1280 G and 1380 G rail steels under different rolling cycles. (a) 1280 G after 1×10^4 cycles; (b) 1280 G after 7.5×10^4 cycles, where yellow dashed lines show the uneven deformation of segregation bands; (c) 1380 G after 1×10^4 cycles; (d) 1380 G after 7.5×10^4 cycles.

In summary, the 1380 G bainitic rail steel has superior resistance to plastic deformation and fatigue crack initiation, so it exhibits better wear resistance compared with the 1280 G rail steel. In general, it is considered that the higher hardness of pearlitic rail could result in the shallow plastic deformation layer, but its crack propagation is more significant, and the fatigue damage is more severe. Hence, the rolling contact fatigue (RCF) damage is always serious in the case of mild wear [34]. However, in this study, we found that the resistance to plastic deformation and crack initiation is improved simultaneously, which leads to the lower density of micro-cracks and thinner plastic deformation layer in 1380 G rail steel. In this case, the spalling and peeling are rarely formed, which reduced the wear loss mass in the 1380 G rail steel. It is suggested that the fine multiphase structure and high stable retained austenite contribute to the improvement of wear and RCF of the 1380 G bainitic rail steel.

Meanwhile, the online heat treated bainitic rail steels have been used at the curved track of heavy haul railway. Figure 12 shows the profile of the 1380 G worn rail steel after experiencing ~500 MGT (million gross tons). It is found that the most severe position of worn rails has a maximum side wear of ~5.5 mm and vertical wear of ~3.8 mm. Meanwhile, the spalling and peeling are not found on the worn rail surface.

Figure 12. (a) Photograph of a 1380 G bainitic rail steel after the experience of ~500 million gross tons at the curved track of heavy haul railway; (b) the cross-section profile, where the blue and red lines show the original and worn profiles, respectively.

Based on the operation report of the railway, the current pearlitic rail steel (HB330 grade) would be worn to the limit (maximum side wear of ~15 mm) after the passing gross load of ~500 MGT. However, it is speculated that the 1380 G bainitic rail steel could endure the passing gross load of more than 1000 MGT according to the change of cross-section profile (Figure 12b). The rail grinding is not necessary during the service of 1380 G bainitic rail steels since there is not any serious contact surface damage, which would save a great deal of maintenance costs. Hence, although there is a high alloying cost, the life cycle cost (LCC) of 1380 G bainitic rail steels could be reduced by about 50% compared with the conventional pearlitic rail steels. Furthermore, it is also most important that the 1380 G bainitic rail steels have much higher fracture toughness compared with pearlitic rail steels. This could avoid sudden disasters (e.g., train derailment because of rail fracture) during the operation of trains.

4. Conclusions

In this work, we investigated the relationship between the microstructure, mechanical stability of retained austenite, and wear resistance of a novel bainitic rail steel (namely 1380G) treated by the online bainite-based quenching and partitioning (BQ&P) concept. The conventional air cooling bainitic rail steel (namely 1280G) was also studied as a comparison. The main conclusions are given as follows:

- (1) The microstructure in the 1280 G bainitic rail steel is composed of granular bainite and coarse martensite/austenite islands. In contrast, the microstructure in the 1380 G bainitic rail steel is composed of fine lath bainite, lower bainite, carbon-depleted martensite, and film-like retained austenite.
- (2) The retained austenite in the 1380 G bainitic rail steel has higher mechanical stability compared with the 1280 G variant. This is mainly attributed to the small size of the film-like retained austenite in the 1380 G bainitic rail steel.
- (3) An excellent combination of strength, ductility, and toughness was achieved in the 1380 G bainitic rail steel (ultimate tensile strength: 1390 MPa; total elongation: 15%; impact toughness: 104 J; and fracture toughness at $-20\text{ }^{\circ}\text{C}$: $107\text{ MPa}\cdot\text{m}^{1/2}$). The enhanced mechanical properties are attributed to the refined microstructure and higher mechanical stability of the retained austenite.
- (4) The wear mass loss of the 1280 G bainitic rail steel (~1.03 g) is almost twice that of the 1380 G variant (~0.59 g) when the cyclic number reaches 7.5×10^5 . Meanwhile, the rolling contact fatigue cracks are rarely found in the 1380 G rail steel. The improved performances of wear and rolling contact fatigue of the 1380 G bainitic rail steel are attributed to the resistance of plastic deformation and crack propagation.

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