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Effects of deformation on the microstructures and mechanical properties of carbide–free bainitic steel for railway crossing and its hydrogen embrittlement characteristics

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ABSTRACT

The effects of deformation on the microstructures and mechanical properties of carbide–free bainitic steel for railway crossing were studied in detail, with an aim of analyzing the hydrogen embrittlement characteristics of the deformed bainitic steel. A small deformation could cause a notable increase in the strength and a slash decrease in uniform plasticity. High density of dislocations accumulated during the deformation and the newly formed high strength martensite should be responsible for the changes of mechanical properties. Results from the slow strain rate tests reveal that the hydrogen embrittlement sensitivity is higher for the deformed bainitic steel as compared with the undeformed steel, which also increase with increasing reduction. It should be attributed to the newly formed martensite phase, the accumulated dislocations and the heterogeneous residual stress in the deformed specimens. Fractography observations reveal a changed fracture mode from ductile to brittle with increasing hydrogen charging time.

1. Introduction

Bainitic steels have attracted extensive attention for their excellent mechanical properties, such as high strength and toughness, and have been widely used in industries to manufacture bearings, gears, railway rails and crossings [1–3]. The carbide–free bainite microstructures consisting of bainitic ferrite plate and retained austenite is generally obtained in steel with non-carbide forming elements including Si and Al, and possesses excellent mechanical properties [4–6], which has been widely used in railway, such as manufacturing rail and crossing [7,8]. According to the statistics results from railway interests, however, a series of failure phenomena occurred on the bainite rails and crossings, such as untimely brittle fracture and flake off [9–11], which bring in an unstable service life of the rails and crossings and cause potential safety hazards to railway transportation.

The embrittlement characteristic caused by the hydrogen largely accounts for the above unstable service life. The effects of hydrogen on the properties of bainitic steel had been studied by some researchers. Zheng et al. found that the fatigue failure of carbide–free bainitic steel was accelerated by hydrogen, for hydrogen reduced the energy for the fatigue crack passing through the phase interface, and they also proposed that the addition of Al alloying element in bainitic steel could decrease the hydrogen embrittlement [12,13]. Zhang et al. found the plasticity of the bainitic steel decreased sharply with the increase of hydrogen content with a critical content of 7×10^{-5} wt% [14]. Chatzidouros et al. found that the fracture toughness of the ferritic–bainitic steel was decreased notably when the specimen was in-situ hydrogen charged, and the ferritic–bainitic microstructure was very susceptible to hydrogen-induced reduction of fracture toughness [15]. Arafin and Szpunar found that the bainitic lath boundaries could act as hydrogen trap sites, decreasing the mobile hydrogen content, and that the bainitic lath microstructure was more vulnerable to hydrogen induced crack at high cathodic potential than the ferritic/granular bainitic ones [16].

However, the hydrogen embrittlement characteristic of deformed bainitic steel has not yet been studied. During the service life, the railway rails and crossings undergo the roller compaction from the train wheels and deform inevitably. Then the microstructure of the steel is changed, which also changes the mechanical properties and the hydrogen embrittlement susceptibility. Then, it is worth studying the hydrogen embrittlement characteristic of deformed bainitic

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steel. In the present study, the 1500 MPa level bainitic steel used in practice was chosen as the specimen. The steel was cold rolled to a reduction of 10% and 30%. Then the microstructure and mechanical properties of the deformed material was characterized, and the effect of deformation on the hydrogen embrittlement characteristic was analyzed.

2. Experimental details

The as-received material used in the present work was the rail steel that used in practice, and the chemical composition of the material is listed in Table 1.

The dehydrogenation annealing process was carried out on the material firstly [11]. Then the material was heated to 930 °C and held for 30 min before being air cooled to room temperature. Finally, it was tempered at 320 °C for 1 h. The tempered material with an initial size of 70 mm × 20 mm × 15 mm ($L \times W \times H$) was rolled to reductions of 10% and 30% at room temperature, respectively. The samples were cut along the rolling direction. The tensile testing was carried out on a MTS material testing machine with a crosshead speed of 3 mm/min. Tensile samples with a gauge of 25 mm and a gauge diameter of 5 mm were fabricated. The samples for metallographic observation were mechanically ground, chemically polished and etched with 3% Nital; the microstructures were then examined using an optical microscope. Microstructures were examined using a TEM (Hitachi H-800) operating at 200 kV. The TEM samples were thinned to perforation on a TenuPol-5 twin-jet unit, at a voltage of 27 V with an electrolyte consisting of 7% perchloric acid and 93% glacial acetic acid. The retained austenite content and the carbon content in the austenite were analyzed by a D/max-2500/PC X-ray diffractometer (XRD) with CuKα radiation.

The samples were cathodically charged with hydrogen in a mixed electrolyte solution of 0.5 mol/L H_2SO_4+250 mg/L Na_3AsO_3. The current density for hydrogen charging was 10 mA cm^{-2}. Different charging times were selected, then samples were obtained different hydrogen contents. Uniform distribution of hydrogen was achieved by homogenization heat treatment at 190 °C for 10 h. In order to prevent the hydrogen from escaping from the samples, bright cadmium plating was applied using a 500 ml mixed solution of distilled water, 98% oil of vitriol (25 g), dried $CdSO_4$ powder (25 g), anhydrous Na_2SO_4 (22.5 g), glutin (3 g) and phenol (1.5 g), and 25 mA cm^{-2} current density. A slow strain rate of 6.7×10^{-6} s^{-1} was used to fully reflect the hydrogen embrittlement sensitivity of the steels. The fractured samples were examined via KYKY-2800 scanning electron microscope (SEM).

3. Results and discussion

3.1. Deformation behavior of carbide–free bainitic steel

Fig. 1 shows the optimal microstructures of the undeformed and deformed specimens. It can be seen that with increasing the reduction from 0% to 30%, the typical acicular bainite microstructure, obtained after the heat treatment, deformed and rotated to the direction that paralleling the rolling direction.

TEM observations showed in Fig. 2a revealed that no carbide precipitated within the bainite microstructure, which demonstrated that a carbide–free bainite microstructure was obtained after the heat treatment. Moreover, with increasing the reduction, the width of bainite ferrite (BF) and residual austenite (RA) decreased gradually, and the dislocation density within the BF increased, as shown in Fig. 2b and c.

Table 1
Chemical composition of the studied material (wt%).

C	Mn	Si	Ni	Cr	Mo	Al
0.30	1.58	1.44	0.45	1.13	0.40	0.48

Fig. 3a shows the XRD patterns of the undeformed and deformed carbide–free bainitic steel. Two kinds of diffraction peaks that indicate body-centered cubic (BCC) structure and face-centered cubic (FCC) structure can be seen in the patterns. The intensity of the diffraction peaks indicated the FCC structure gradually reduced with increasing reduction, indicating that a strain induced martensite transformation occurred during the rolling process. The volume fraction of the RA phase can be estimated according to Eq. (1) [17].

$$V_\gamma = \frac{1}{n}\sum_{j=1}^{n}\frac{I_\gamma^j}{R_\gamma^j} \Big/ \left(\frac{1}{n}\sum_{j=1}^{n}\frac{I_\gamma^j}{R_\gamma^j} + \frac{1}{m}\sum_{i=1}^{m}\frac{I_\alpha^i}{R_\alpha^i}\right) \tag{1}$$

where n, m, I and R represent the number of peaks of a phase used in calculation, the integrated intensity for a diffraction peak, and the material scattering factor, respectively. R has been given in Ref. [17] by the equation $R = (1/v^2)[|F|^2 p((1 + \cos^2 2\theta)/\sin\theta\sin 2\theta)]\cdot e^{-2M}$, where v is the volume of unit cell; F is the structure factor and e^{-2M} is the temperature factor which has a negligible effect on the calculation result. Reflections (200) and (220) of the RA phase, and (200), (211), (220) of the BCC phase were used. The volume fraction of RA decreased from ~8% to 2% with increasing reduction from 0% to 30%, as shown in Fig. 3b.

The carbon concentration in the RA is estimated using the lattice parameters of the RA [18] and according to Eq. (2).

$$w(C)_\gamma = (a_\gamma - 3.578)/0.044 \tag{2}$$

where a_γ is the lattice parameters of the RA. The estimated result shows the carbon concentration in RA reached 0.95 wt%.

Fig. 4 shows the engineering stress–strain curves of the bainitic steel with different reductions. The mechanical properties of the steel are summarized in Table 2. It can be seen that the strength and hardness increased, and the plasticity decreased with increasing reduction, which mainly resulted from the increased dislocation density, as shown in Fig. 2. The phase transformation from RA to martensite phase also played a role in the increased strength, as shown in Fig. 3, for the higher strength of martensite phase as compared with the RA phase. The increase in strength via the dislocation strengthening and transformation strengthening from austensite to martensite inevitably causes a decrease in plasticity. It also should be noted that the strength increment mainly occurred at the early stage of deformation.

It also can be seen that the ratio of yield stress/tensile stress is increased from 0.78 to 0.92 with increasing reduction from 0% to 30%, which indicates that the work hardening ability of the bainitic steel is weakened after deformation. It can be seen that uniform strain of 4.1% was obtained in the undeformed specimen before the occurrence of necking in specimen, only 1.3% and 0.9% retained in the 10% and 30% deformed specimens, as shown in Fig. 4, revealing that the uniform deformation ability of the carbide–free bainitic steel was reduced after deformation. The following two reasons should be responsible for this change. For one reason: the carbon content in RA reached 0.95 wt%, suggesting that the martensite phase transformed from RA and via displacive phase transformation was in a high carbon condition. The high carbon martensite is brittleness. Then the micro-voids can be formed near the martensite phase and finally results in local plastic instability. For another one, the accumulated dislocations during the rolling deformation reduced the slip distance of the dislocations and