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Effects of isothermal transformation conditions on the microstructure and hardness values of a high-carbon Al–Si alloyed steel

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ABSTRACT

In this investigation, the microstructure and hardness of a high-strength high-carbon steel containing 0.80 wt.% silicon and 0.84 wt.% aluminum have been evaluated under different austempering conditions. Austempering was performed at temperatures of 300, 330 and, 360 °C for different times from 1 to 8 h. Observations carried out by light optical microscopy (LOM) and scanning electron microscopy (SEM) revealed that the austempered samples have a microstructure consisting mostly of bainitic ferrite and retained austenite, and this was confirmed by X-ray diffraction (XRD) analysis and microhardness measurements. This simply indicates that it is possible to partially substitute aluminum for silicon and still retard the formation of carbides. There were no significant changes in ferrite plate thickness with increasing transformation temperature. The fraction of retained austenite, however, increased with temperature. A yield strength of 1170 MPa, an ultimate tensile strength of 1370 MPa, and a total elongation of 9.1% were obtained after isothermal transformation at 360 °C for 4 h. Depending on the transformation conditions, hardness values of about 500–660 HV30 were obtained. The hardness increased as the transformation temperature decreased.

1. Introduction

Notable mechanical properties of austempered steels make them ideal replacement materials for quenched and tempered (QT) steels and austempered ductile cast irons (ADI) in an increasingly large number of applications, especially in railway and automotive industries [1–7]. The microstructural features of austempered steels, especially plate size of bainitic ferrite and volume fraction of retained austenite, are most seriously affected by the austempering variables, i.e., austempering temperature and time [8–10]. Owing to the strong dependence of the mechanical behavior of steels on the microstructure, the high state of performance and reliability of austempered steel products is a direct consequence of the proper selection of austempering variables.

Speich and Cohen [11], in a study of the growth kinetics of bainite plates, reported that the growth rate of bainite was time-independent at a given temperature and that edgewise and sidewise growth rates increased with temperature. Chang and Bhadeshia [12] reported that the thickness of bainitic ferrite plates was increased by increasing transformation temperature. A few years later, Singh and Bhadeshia [13] showed that temperature rarely contributes independently to the determination of the plate thickness. They also indicated that the largest effect on the plate thickness comes from the driving force available for nucleation of bainite and the yield strength of austenite at the transformation temperature. These both factors increase as the transformation temperature decreases, leading to thinner bainite plates [13].

Lee et al. [14] showed that the volume fraction of retained austenite initially increases with temperature, reaches a maximum, and then drastically decreases. A similar trend was reported by Putatunda [2]. During austempering process, austenite (γ) initially undergoes a transformation into bainitic ferrite (α_B) and carbon-enriched austenite (γ_{HC}) (Eq. (1)). At longer holding times, this carbon-enriched austenite further decomposes into a mixture of bainitic ferrite and carbide (ε) (Eq. (2)) [15].

$$\gamma \rightarrow \alpha + \gamma_{HC} \quad (1)$$

$$\gamma_{HC} \rightarrow \alpha + \varepsilon \quad (2)$$

The undercooling below the bainite start temperature (B_S) is quite large at very low transformation temperatures, thereby providing a higher driving force for nucleation of bainitic ferrite. This in turn leads to the formation of a greater amount of bainitic ferrite. This is why only a little amount of austenite remains in microstructures obtained following transformation at lower temperatures. The degree of undercooling and hence the driving force for nucleation of bainite decrease as the temperature rises. Therefore, the amount of retained austenite increases gradually by increasing the transformation

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temperature. At very high transformation temperatures, carbon-enriched austenite decomposes into ferrite and carbide in accordance with Eq. (2), leading to a significant reduction in the amount of retained austenite [2,14,16].

The precipitation of carbides from austenite during bainite transformation leads to a remarkable deterioration in toughness of austempered steels [8]. About 1.5–2 wt.% silicon is generally added to these steels to retard carbide precipitation during austempering [17,18]. However, it has been demonstrated that silicon retards bainite formation [19,20]. Silicon also forms an oxide scale that impairs the response of steel to hot rolling [21]. Moreover, high silicon content has been shown [22] to deteriorate the galvanizability of steel. Majority of previous investigations related to carbide-free bainitic steels [2,10,23] have been conducted on alloys containing more than 1 wt.% silicon and only limited work [19,24] has been devoted to alloys with a silicon content below 1 wt.%. The present study was aimed to explore the possibility of formation of the carbide-free bainitic microstructure in a high-carbon steel co-alloyed with 0.8 wt.% Si and 0.84 wt.% Al and examine the effects of austempering time and temperature on the microstructure and hardness of this steel.

2. Materials and methods

The alloy investigated in this work was prepared in a 20 kg medium frequency induction furnace with charge materials of clean mild steel scrap, petroleum coke, Fe–Cr, Fe–Si, Fe–Mn and Fe–Mo ferro-alloys and commercially pure aluminum (99.7%) and cobalt (99.3%). The steel was cast as a cylinder bar of 50 mm in diameter and 600 mm in length. The cast cylinder was electro slag remelted (ESR) to get clean steel. The chemical composition of the manufactured steel, measured after ESR process, is given in Table 1.

The Ac_3 temperature of this steel was calculated to be 788 °C according to the familiar Andrew [25] formula (Eq. (3)).

$$Ac_3(^\circ C) = 910 - 203\sqrt{x_C} - 15.2x_{Ni} + 44.7x_{Si} + 104x_V + 31.5x_{Mo} + 13.1x_W \quad (3)$$

Where x_i is the concentration of element i in wt.%. Samples with dimensions of 65 × 30 × 8 mm were fully annealed at 810 °C for 270 min followed by furnace cooling to room temperature. The TTT diagram of the investigated steel (Fig. 1) was calculated using MAP program MUCG83, developed by Bhadeshia [26]. The TTT diagram was used to design the heat treatment schedules. The samples were initially austenitised at 950 °C for 30 min, and then isothermally transformed in a molten $NaNO_3$–KNO_3 salt bath at 300, 330 and, 360 °C for different holding times before quenching into water maintained at room temperature. The isothermal experiments were conducted at relatively low temperatures within the bainitic range in order to minimize the scale of microstructure and simultaneously reduce the possibility of carbide precipitation. Three samples were heat treated at each condition. More detailed information on the heat treatment process is illustrated in Fig. 2.

Metallographic specimens were prepared using standard techniques and chemically etched in 2% nital solution. Microstructures were then examined using Olympus DP25 light optical microscopy (LOM) and Tescan VEGA-II scanning electron microscopy (SEM) operating at 30 kV. X-ray diffraction (XRD) analysis was performed on a PANalytical X'Pert PRO MPD diffractometer using unfiltered CuK_α radiation (λ = 1.540 Å) at 40 kV and 40 mA and step size of 0.026° (2θ). Scanning

Table 1
Chemical composition of the experimental steel.

Element	C	Si	Mn	Al	Cr	Mo	Co	P	S	Fe
(wt.%)	0.658	0.804	0.993	0.841	0.950	0.197	1.530	0.014	0.006	Bal.

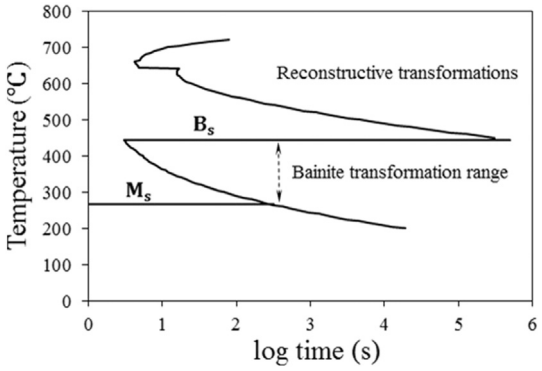

Fig. 1. TTT diagram of the investigated steel, calculated using MAP program MUCG83.

was carried out at a rate of 0.15 °min^{-1} over the range 2θ = 20 − 100 °.

Tensile test specimens were prepared in accordance with B 6 × 30 DIN 50125. Tensile tests were carried out using a 250 kN universal testing machine (Schenck-Trebel Rm 250) with a cross-head speed of 1 mm/min at room temperature. The strength and elongation values reported here are averages from three tests. Macro-hardness Measurements were conducted using a load of 30 kgf on a Vickers hardness testing machine. At least ten readings were taken for each austempering condition and their average is reported here as the hardness value. Micro-Vickers tests were also performed at indentation load of 0.1 kgf. The volume fraction of retained austenite was estimated using MIP image analyzer system. The reported value for each austempering condition is an average of at least five measurements performed over different micrographs.

3. Results and discussions

3.1. Microstructure analysis

3.1.1. Phase characterization

The microstructure of the fully annealed steel is shown in Fig. 3. As it can be seen, a coarse-grained ferrite-pearlite microstructure has been obtained through full annealing.

Typical optical micrographs of samples austempered at different conditions are shown in Fig. 4. One can see that the microstructures consist of bainite (dark-etched areas) and retained austenite (white areas).

In order to insure that the white matrix in Fig. 4 was correctly identified as retained austenite, micro-hardness measurements and X-ray diffraction (XRD) analysis were conducted on samples austempered at 300 °C for 8 h. Micro-hardness measurements done on the white constituent revealed average hardness value of 421 ± 8 HV0.1. Since the hardness value of samples water quenched to room temperature

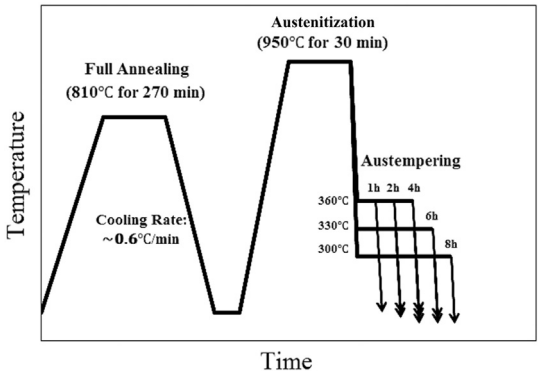

Fig. 2. Schematic illustration of the heat treatment schedules applied to the investigated steel.