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Coupled toughening of a low carbon medium manganese heavy steel plate investigated by electron back–scattered diffraction (EBSD)

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ABSTRACT

Nearly full fine ferrite microstructure ($d=2.7\pm0.5$ μm) and fine ferrite ($d=2.4\pm0.3$ μm) plus ~8.9% retained austenite with relatively good stability were produced. The upper shelf energy increases by ~30 J and the CVN impact energy increases by ~60 J at −100 °C by introducing ~8.9% retained austenite. The high toughness (263 J at −40 °C and 191 J at −100 °C) has been achieved by grain refinement and retained austenite.

1. Introduction

As we know that reducing the weight of engineering components by means of application of high strength steels is an effective way to save resources and protect environment [1,2]. However, the body centered cubic steels show a ductile–brittle transition [3–7], and the brittle damage may be caused with increasing the strength due to the deterioration of toughness. Therefore, it is crucial to develop advanced high strength steels with high toughness. Sufficient grain refinement can enhance strength and toughness and always lower ductile–brittle transition temperature (DBTT) [8–10]. However, the grain refinement in general lowers strain hardening capacity due to the rapid dynamic recovery by the trapped lattice dislocations spreading into grain boundaries [10]. Hence, we must concern this mutually exclusive phenomenon. Efforts have been made in medium manganese steels to improve ductility and strain hardening capacity by utilizing deformation mechanisms of meta–stable retained austenite [1, 11–14]. On the other hand, it has been recognized that the meta–stable retained austenite has also some contributions to toughness [15–17].

Although there are extensive studies on medium manganese steels, very few studies on a high toughness medium manganese heavy steel plate have been carried out. Furthermore, the contributions to toughness between ferrite matrix and retained austenite have not been distinguished. In the present work, the fine ferrite microstructure and fine ferrite plus a considerable volume fraction of retained austenite microstructure were fabricated. The coupled toughening has been obtained.

2. Experimental procedure

The chemical composition of the steel is Fe–0.05C–0.23Si–3.3Mn–0.003P–0.001S–1.4Ni (in mass %). The steel was prepared by high–frequency vacuum induction melting. The steel was reheated to 1200 °C and soaked at this temperature for 2 h, then hot rolled in six passes to a 12 mm thickness plate. The plate was further isothermally treated at 600 °C or 630 °C for 2 h, and then water quenched to room temperature.

Specimens were prepared from the heat treated steels and their surfaces were mechanically polished and then further electropolished in a mixture consisting of ~12.5% perchloric acid and ~87.5% absolute ethyl alcohol to remove surface strain layer. These specimens were examined on a scanning electron microscope (SEM, Zeiss Ultra 55) equipped with an electron back–scattered diffraction (EBSD) attachment. The step size for EBSD analyses is 0.05 μm which is sufficiently small to obtain clear morphology. Moreover, three different regions obtained from the same specimen were analyzed to reduce errors. EBSD data was post–processed by HKL CHANNEL 5 flamenco software. In addition, the EBSD specimens were further analyzed using X–ray diffraction

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(XRD, Cu K_α radiation), and the volume fractions of retained austenite were measured using the integrated intensities of $(200)_\gamma$, $(220)_\gamma$, $(113)_\gamma$, $(200)_\alpha$, $(112)_\alpha$ and $(220)_\alpha$ diffraction peaks [18,19].

Standard round tensile samples with the gauge length of 40 mm and diameter of 8 mm were prepared along the rolling direction, and the tensile properties were measured using a CMT–5105 tensile tester at room temperature at a cross beam speed of 3 mm/min. Standard Charpy V–notch (CVN) impact samples with the size of $10 \times 10 \times 55$ mm^3 were also prepared along the rolling direction, and the impact properties were measured using a JBW–500 impact tester at the temperatures of 15, −40, −60, −80 and −100 °C.

3. Results and discussion

Figs. 1 and 2 provide the EBSD crystallographic analyses for the steels after heat treating at 600 °C and 630 °C for 2 h, respectively. Fig. 1a shows that very few ultrafine austenite grains (the average size is about 240 nm) can be observed and they mainly display at high angle grain boundaries. The reversion transformation start temperature was determined to be about 615 °C by the dilatometric curve at a heating rate of 1 °C/s. Therefore, it can be deduced that the isothermal heat treating temperature of 600 °C may be very close to the equilibrium reversion transformation start temperature, and little austenite phase is formed during isothermal holding at 600 °C. However, when the isothermal heat treating temperature increases to 630 °C which is sufficiently greater than the equilibrium reversion transformation start temperature, the volume fraction and grain size of retained austenite obviously increase, as shown in Fig. 2a, and the XRD measurement shows that its volume fraction increases to ~8.9%. Therefore, two types of microstructure, i.e., nearly full fine ferrite microstructure and fine ferrite plus austenite microstructure, have formed.

On the other hand, the ferrite/austenite interface were highlighted in different colors which represent the interface deviating from the Kurdjumov–Sachs orientation relationship (K–S OR) of $(111)_\gamma//(011)_\alpha$ and $[10-1]_\gamma//[11-1]_\alpha$ [20], as shown in Figs. 1b and 2b. Figs. 1c and 2c show that some austenite grains can maintain approximate K–S OR with all adjacent ferrite grains. This result has been also observed in other alloys [21,22]. Figs. 1d and 2d show that some austenite grains have an approximate K–S OR with one adjacent ferrite grain and hold an incoherent interface with other adjacent ferrite grains. Furthermore, the austenite grains deviating K–S OR with all neighboring ferrite grains cannot be observed. In other words, almost all the austenite grains have an approximate K–S OR with respect to at least one neighboring ferrite grain.

It is interesting to note that the fine ferrite microstructure has been obtained. Moreover, the full–width at half–maximum values of $(200)_\alpha$, $(112)_\alpha$ and $(220)_\alpha$ diffraction peaks are ~0.74°,~0.74° and ~0.98° for as–hot rolled steel, respectively. But these values are ~0.31°,~0.37° and ~0.49° and ~0.31°,~0.36° and ~0.47° for the steel after heat treating at 600 °C and 630 °C, respectively. These results imply that the dislocation density has dramatically decreased [23,24]. The dislocation density in martensite is very high, showing the tangle of dislocations. Hence, the dislocation networks may form due to the climb and annihilation of dislocation at a relatively high heat treating temperature. These dislocation networks may further evolve to sub–grains, leading to the formation of some fine ferrite grains (denoted by the arrows in Figs. 1b and 2b) [25]. On the other hand, some laths may coalesce, also leading to the formation of fine ferrite grains. In addition, grain boundaries were defined where the crystallographic misorientations are greater than 15° between each other, and the

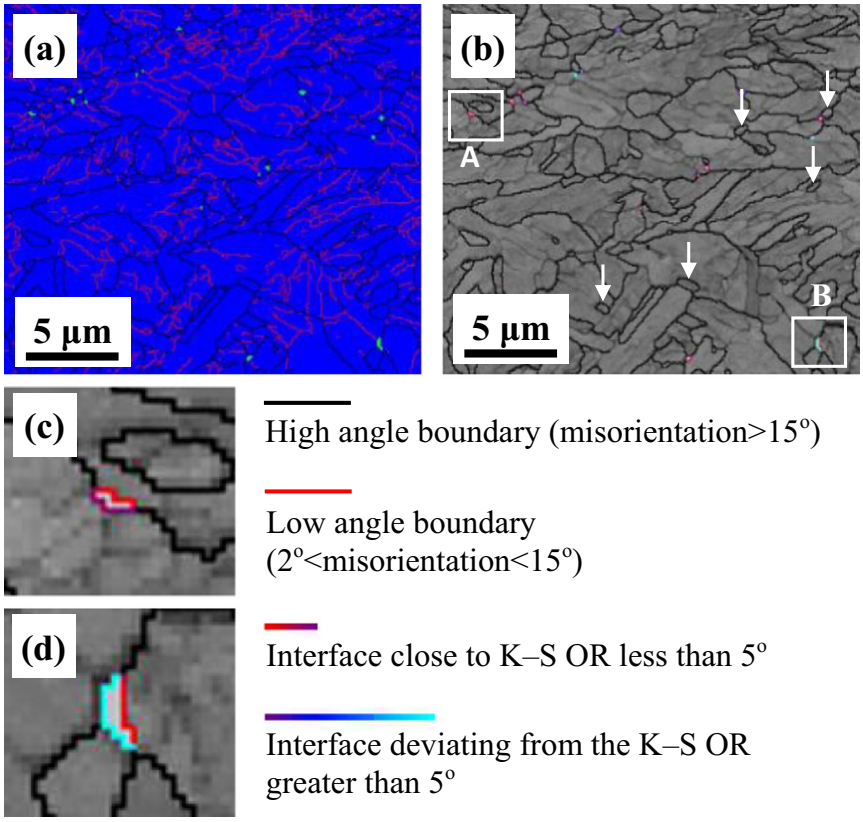

Fig. 1. EBSD analyses on the steel after heat treating at 600 °C for 2 h: (a) the phase map showing the microstructure consisting of ferrite (blue color) and austenite (green color), (b) K–S orientation relationship map showing the ferrite/austenite interface deviating from the K–S OR, (c) and (d) the local magnified maps obtained from the regions A and B in Fig. 1b, respectively. (For interpretation of the references to color in this figure legend, the reader is referred to the web version of this article.)