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Microstructural evolution and mechanical properties of hot-rolled 11% manganese TRIP steel

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ABSTRACT

Mechanical behaviors of transformation induced plasticity (TRIP) steels largely depend on the amount and stability of austenite. In this investigation, a large volume fraction of austenite (> 65%) was produced in a hot-rolled Fe-11Mn-3.8Al-0.18C TRIP steel by solution treatment in the temperature range of 750–800 °C for 1 h. The hot-rolled alloy exhibited an excellent combination of total elongation of 35–40% and ultimate tensile strength of 880–1100 MPa and this was found to have a similar or higher level of tensile properties compared with other TRIP steels. In the meantime, less cold-rolling work or annealing time was required in the present work. The outstanding properties of the experimental steel were mainly attributed to the enhanced TRIP effect due to the large fraction of austenite. It is shown that the morphology played a more significant role than orientation in the stability of austenite.

1. Introduction

The demands for energy conservation and environmental protection have compelled the automakers to design light-weight auto bodies with enhanced crash resistance. Transformation induced plasticity (TRIP) steels which have attractive combinations of strength and ductility have been regarded as the potential candidates for automotive application. There are basically two types of lightweight steels: austenite-base and ferrite-base steels. Typical Mn and C contents in the low alloy TRIP steels are ~1.5 wt% and ~0.2 wt%, respectively, which are required to achieve the austenite stability [1]. The low alloy TRIP steels referred to as the "first generation" advanced high strength steels (AHSS), comprise ferrite matrix (55–65%) along with bainite (25–35%) and metastable retained austenite (5–20%) [2]. These steels demonstrate advantageous balance of strength and ductility, but commercial TRIP grades are limited mostly by tensile strength of 800 MPa [3]. A variety of studies [4–6] have shown that fully austenitic steels with high Mn (> 18%) content, the representative "second generation" AHSS, with an excellent combination of strength and ductility have the capability to exhibit twin induced plasticity (TWIP) with or without the assistance of TRIP effect. However, the steels are expensive as a result of high alloy levels and have received limited use in the commercial application.

The "third generation" AHSS ought to be less alloyed than TWIP steels, exhibiting intermediate properties between the first and second generation steels. Some publications [3,7,12] consider steels with medium Mn (4–10%) as the potential candidates to achieve the performance of the 3rd generation AHSS. Some of publications [7,8] on medium Mn containing steels proposed that the stability of austenite can be attributed to the significant partitioning of Mn between ferrite and austenite during long (hours) holding time in the intercritical region. Miller reported austenite fraction up to 40% in 6.0% Mn alloys [9]. Similar level of austenite was obtained by Merwin [10] in a 7.1% Mn alloy; and in an 8.0% Mn alloy, austenite fraction up to 48% was reported by Kim et al. [11]. Therefore, Mn is an important element to enhance the volume fraction of retained austenite.

Conventionally, TRIP steels contain about 1.5 wt% silicon which prevents the formation of cementite during the isothermal bainitic transformation [13]. Nevertheless, it was reported that silicon in a concentration higher than 0.5 wt% is detrimental to surface quality [14]. High silicon leads to poor weld ability as well [15]. To avoid these problems silicon can be partially replaced by aluminum [16–18]. Aluminum, like silicon, has the same function in suppressing cementite formation. Furthermore, Al facilitates coating ability by forming an inhibition layer on the steel surface [19]. In view of above analyses, an attempt has been made to develop a Fe-11.02Mn-3.81Al-0.18C austenite matrix TRIP steel. Mn and Al contents of the experimental steel have been judiciously balanced to achieve adequate hardenability and strengthen the austenite phase.

The present work focuses on the microstructure evolution resulting from the heat treatment and TRIP effect. The stability of austenite was discussed as well.

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2. Experimental

A 40 kg experimental steel cast ingot was manufactured using a vacuum furnace. The cast ingot was prepared at 1200 °C for 2 h, hot forged between 1150 °C and 850 °C to 100 mm × 30 mm bars, and then air cooled to room temperature. Subsequently, the bars were soaked at 1200 °C for 2 h, and then hot rolled down to 4 mm in thickness after eight passes within the temperature range of 1150–850 °C. Finally, the as-hot-rolled sheets were air cooled to ambient temperature.

A model proposed by Moor et al. [20] was used to predict the amount of austenite stabilized to room temperature through the enrichment of austenite with Mn, Al and C. It was further deduced that there existed a temperature in the intercritical region resulting in the maximum austenite retention at room temperature. Consequently, the model was instructive to the optimization of alloy composition design for the experimental steel. For the purpose of intercritical temperature range, the phase transformations were studied by means of dilatometry. The sample used for dilatometry was solid cylindrical specimen with a diameter of 3 mm and a length of 10 mm.

In conventional TRIP steels, a two-stage heat treatment is adopted [21]. They are firstly subjected to intercritical annealing (800–900 °C) to form a mixture of ferrite and austenite, and then followed by isothermal annealing (350–450 °C) to stabilize austenite. However, the heat treatment was proven to be not applicable to the experimental steel. Instead, a more convenient two-stage heat treatment was applied to the experimental steel. (1) solution treatment. The as-hot rolled sheets were soaked in a high temperature furnace at the temperature of 700, 750, 800, 850, 900 °C, corresponding to the intercritical region, for 1 h, respectively, and then quenched in the water immediately. (2) Tempering. The quenched samples were tempered at a temperature of 200 °C for 20 min in order to relieve internal stress, and thus the ductility could be improved, the samples were then air cooled to ambient temperature. The attribution of tempering to the excellent mechanical properties was discussed in the later section.

Flat tensile specimens with a width of 12.5 mm and a gauge length of 50 mm were machined from the tempered and untempered sheets with the tensile axis parallel to the prior rolling direction. Tensile tests were carried out at room temperature using a universal testing machine (SANSCMT5000) at a constant crosshead speed of 3 mm min^{-1}. The samples were etched with 25% sodium bisulfite aqueous solution. Microstructures were analyzed by optical microscope (OM), scanning electron microscope(SEM) equipped with electron backscatter diffraction (EBSD), and transmission electron microscope (TEM). The volume fraction of austenite was determined by X-ray diffraction (XRD) with CuK$_\alpha$ radiation using the direct comparison method [22]. This method used the integrated intensities of the $(200)_\alpha$ and $(211)_\alpha$ peaks and those of the $(220)_\gamma$ and $(311)_\gamma$ peaks. The volume fraction of the austenite V_A was calculated as following: [23]:

$$V_A = \frac{1.4I_\gamma}{I_a + 1.4I_\gamma}$$

where:

I_γ—integrated intensity of austenite's diffraction lines,
I_α—integrated intensity of α-phase's diffraction lines.

3. Results and discussion

3.1. Composition design and intercritical temperature

The phase fractions based on an equilibrium thermodynamic analysis [20] predicted by Thermo-Calc are shown in Fig. 1(a), and these predicted equilibrium austenite fractions were used as input to the model. Predictions of austenite composition were also made using Thermo-Calc, and the predicted C, Mn, and Al contents in austenite are shown in Fig. 1(b). The resulting predicted fraction of stabilized austenite as a function of

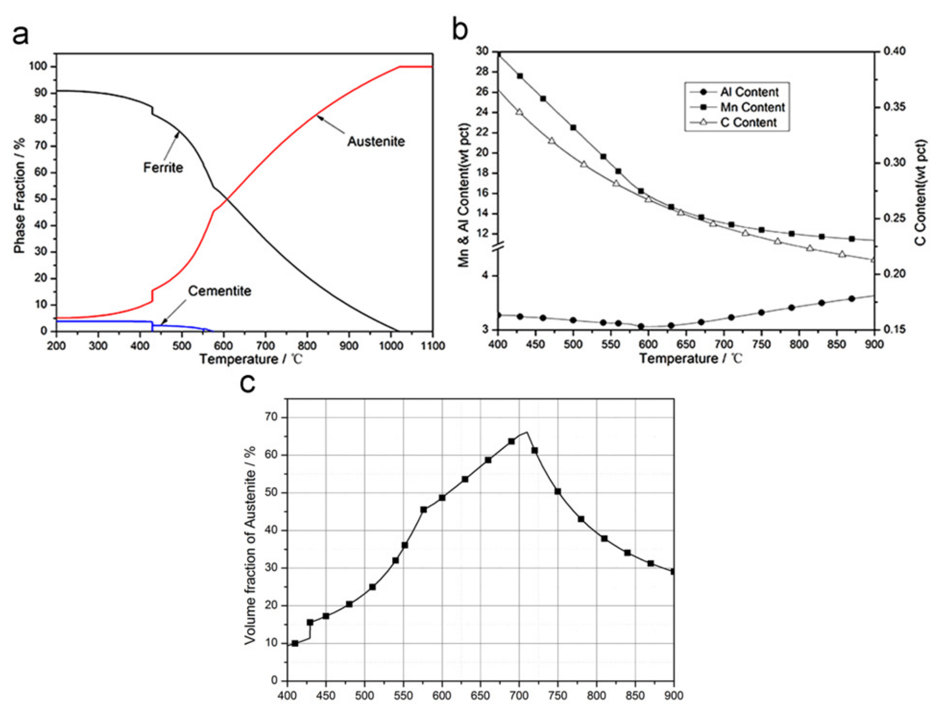

Fig. 1. Schematic illustration of predictive model development for austenite stabilization as a function of temperature. (a) Phase fractions, (b) C, Al and Mn content of austenite in experimental steel, and (c) calculated retained austenite fractions.