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Deformation behavior in cold-rolled medium-manganese TRIP steel and effect of pre-strain on the Lüders bands

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ABSTRACT

Deformation behavior was studied in cold-rolled 0.2C-1.6Al-6.1Mn-Fe transformation-induced plasticity (TRIP) steel subjected to intercritical annealing. The steel intercritically hardened at 650 °C exhibited excellent mechanical properties, and the excellent ductility was primarily associated with the discontinuous TRIP effect. Moreover at 650 °C, the formation of Lüders bands was associated with TRIP effect and cooperative dislocation glide. The length of Lüders strain was gradually reduced with increasing pre-strain, and was eventually eliminated when the pre-strain was increased to 10%. The increased average stability of retained austenite and increased dislocation density in ferrite induced by pre-strain was responsible for decrease and ultimate elimination of Lüders bands. While in steel intercritically annealed at 600 °C, ferrite and austenite was predominantly deformed, which was responsible for poor work hardening rate and inferior tensile properties.

1. Introduction

Environmental concerns, high oil prices and increased safety standards are the driving forces behind the quest to develop advanced high-strength steels for automotive applications [1,2]. Recently, transformation induced plasticity (TRIP) light weight steels are being considered as a new class of high strength steels with exceptional formability [3–5]. Cold rolled TRIP steel sheets are good candidates for automotive applications. The TRIP effect derives from deformation-induced transformation of retained austenite to martensite [6]. This results in work hardening and hence delays the onset of necking, eventually leading to higher total elongation [7]. TRIP steels are characterized by enhanced ductility at very high strength [8], and the TRIP effect depends on the volume fraction and stability of retained austenite.

Many studies have been carried out on Fe-(5–7)Mn-(0.1–0.2)C (wt%) ternary alloy systems [9], and it was suggested that superior mechanical properties can be obtained with increase in C and Mn-content, which increases the volume fraction of retained austenite. In 1972, Miller [10] reported Fe-5.5Mn-0.1C (wt%) steel with tensile strength of 878 MPa and total elongation of 34%. Luo et al. [11] reported Fe-5Mn-0.2C (wt%) steel with good combination of tensile strength of 850–950 MPa and ductility of 20–30%. Light weight steels generally consist of substitutional elements, such as aluminum and silicon, where their role is to optimize austenite stability by suppressing cementite formation [12]. Moreover, Al in TRIP steels encourages the growth of intercritical ferrite [13], which is a soft phase with good ductility and helps stabilize the austenite phase [14]. For the purpose of obtaining light weight steels with high strength-high ductility combination, Al is added to medium Mn TRIP steels.

The typical or “normal” mechanical behavior of polycrystalline metallic materials in tensile tests is the smooth transition from elastic to elastic-plastic region with a steady increase in stress in the stress-strain curve, prior to necking. In contrast, materials exhibiting yield point phenomenon show an abrupt transition from elastic (or mainly elastic) state to elastic-plastic behavior, with drop in stress in the stress-strain curve. In majority of the cases, this drop in stress is accompanied by a region of almost constant nominal stress with some fluctuation. Within this region of nearly constant stress, plastic deformation takes place locally in deformation bands (generally referred as Lüders bands), starting at nucleation point and spreading out towards the unyielded part of the specimen [15–18]. Formation of Lüders bands is a well known phenomenon in metallic materials, especially in low carbon steels, and has been well studied [19,20]. The presence of Lüders bands on the surface of the sheet during forming is undesirable. Thus, the identification of material and processing conditions for elimination of Lüders bands is important [21]. Lüders strain has been observed in stress-strain curves of intercritically annealed

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TRIP steel [6,22,23], but the yield point phenomenon continues to be unclear. Furthermore, limited studies have been carried out to fundamentally understand and eliminate Lüders bands.

Yield strength, tensile strength, and strain hardening rate can be adjusted by processing methods such as pre-straining of austenitic TRIP-assisted steel [24]. The objective of study described here is to develop a scientific basis for eliminating Lüders bands by pre-straining in a low density Fe-0.2C-6.1Mn-1.6Al (wt%) TRIP steel. Additionally, the microstructural evolution and deformation behavior for different degree of pre-strain was studied to elucidate the relationship between deformation behavior and Lüders bands for different degree of pre-strain.

2. Experimental

The nominal chemical composition of the experimental steel was 0.2C-1.6Al-6Mn-Fe (wt%). A 40 kg experimental steel ingot was cast in a vacuum induction furnace. The ingot was heated at 1200 ℃ for 2 h, hot forged into rods of section size 100 mm×30 mm, then air cooled to room temperature (RT). Subsequently, the rods were soaked at 1200 ℃ for 2 h, hot-rolled to 4 mm thick strip in the temperature range of 1150–850 °C, and finally air cooled to room temperature (RT). The as-hot-rolled strips were subsequently cold-rolled to 1 mm thickness.

To establish an appropriate heat treatment schedule, the intercritical temperature range was determined by dilatometry. Dilatometry samples were cylindrical specimens of diameter 3 mm and length 10 mm. The dilatometric plot of the experimental steel is presented in Fig. 1a. After thermal expansion during the heating stage (30–900 ℃) at the rate of 20 ℃/s, the sample was held at 900 ℃ for 3 min. It is clear that no transformation took place in the sample during fast cooling, at

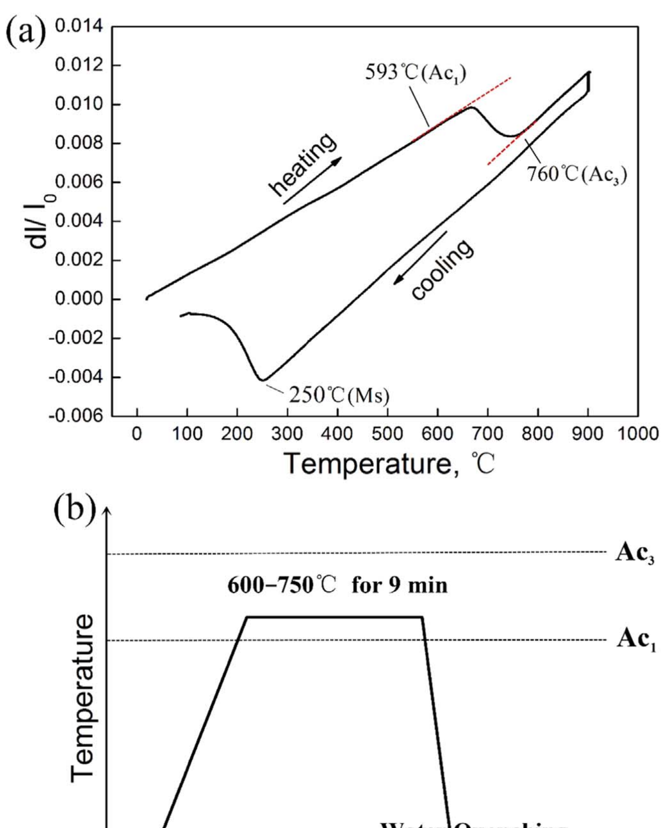

Fig. 1. (a) Dilatometric curve showing intercritical temperature range and (b) heat treatment schedule for the experimental steel.

the rate of 100 ℃/s, until martensite start (Ms) temperature. The intercritical temperature range of the experimental steel was 593–760 ℃. We recently [25–27] demonstrated that austenite reverted transformation (ART) annealing heat treatment (sample was subjected to accelerated cooling in water after austenitization, austenitization followed by intercritical annealing for a long time and finally air cooled to room temperature) used for medium Mn-content steels is not applicable to the experimental steel studied here. A long time annealing makes austenite too stable and weakens the TRIP effect. Thus, we adopted the approach of intercritical hardening of the cold rolled strips. Intercritical annealing was carried out at temperatures in the intercritical range (600–750 ℃) for 9 min, followed by immediate quenching in water to room temperature (as shown in Fig. 1b).

Tensile specimens of dimensions 12.5 mm width and gage length of 25 mm were machined from the heat-treated sheets with 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}. Prior to the tensile tests, the uneven surface of the samples was polished. The samples were etched with 25% sodium bisulfite solution. The microstructure of the experimental steels prior to and after tensile deformation were examined using scanning electron microscope (SEM) and transmission electron microscope (TEM). The volume fraction of austenite was determined by X-ray diffraction (XRD) with CuK$_\alpha$ radiation using direct comparison method [28], involving use of integrated intensities of $(200)_\alpha$ and $(211)_\alpha$ peaks and those of $(200)_\gamma$, $(220)_\gamma$ and $(311)_\gamma$ peaks. The volume fraction of austenite V_A was calculated using equation [29]:

$$V_A = 1.4I_\gamma/(I_\alpha + 1.4I_\gamma) \quad (1)$$

where I_γ is the integrated intensity of austenite and I_α is the integrated intensity of α-phase.

3. Results and discussion

3.1. Microstructure and mechanical properties

The SEM micrographs of cold-rolled samples intercritically hardened in the temperature range of 600–700 ℃ are presented in Fig. 2. Fig. 2a–c describe the microstructure of samples quenched from 600 ℃, 630 ℃ and 650 ℃, respectively. The microstructural constituents consisted of intercritical ferrite (F) and austenite (A) as the dominant phase. When the sample was quenched from 700 ℃, as marked in Fig. 2d, the austenite was significantly decreased because of extensive martensitic (M) transformation.

The variation in the volume fraction of austenite obtained from XRD is summarized in Fig. 3. The austenite content in experimental steel increased from 44 vol% to 72 vol% with increase in intercritical hardening temperature from 600 to 650 ℃, followed by a decrease to 25 vol%, when intercritical hardening was carried out at 700 ℃ because of martensitic transformation. Thus, when the samples were heat-treated in the temperature range of 600–650 ℃, the volume fraction of austenite increased with increase in temperature. With increase in temperature, the austenite stability decreases because of increase in austenite grain size. In the case of samples heat-treated in the temperature range of 700–750 ℃, the higher the temperature, the more is the austenite transformed to martensite on quenching.

Engineering strain-stress plots of steel quenched from different temperatures are presented in Fig. 4a. It can be seen that the ultimate tensile strength (UTS) increased continuously with increase in intercritical temperature and the total elongation (TEL) decreased with increase in temperature after attaining a maximum value at 650 ℃. In the case of samples intercritically hardened at 600 ℃ (henceforth referred as sample 600) and 630 ℃ (sample 630) exhibited low ductility. The reasons underlying the difference in ductility between the samples were elucidated by studying deformation mechanisms (see