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Fast-heating for intercritical annealing of cold-rolled quenching and partitioning steel

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ABSTRACT

In this study, fast-heating (300 °C/s) was applied to achieve the intercritical annealing of a cold-rolled quenching and partitioning (Q&P) steel with a wide range of soaking temperature (770–850 °C) and soaking time (0–120 s) for austenitization. The dilatometry and microstructural analysis revealed that in contrast to the conventional heating rate (5 °C/s), a fast-heating rate led to an accelerated transformation and grain refinement of the prior austenite in the Q&P samples. The microstructures of the Q&P treated samples subjected to different intercritical annealing conditions were studied in detail by various material characterization techniques including electron microscopy, electron probe micro-analysis, and X-ray diffraction. The working hardening behavior and the mechanical stability of the retained austenite were discussed on the basis of the typical stress-strain curves. The statistics of the ultimate tensile strength vs. total elongation of each sample under the orthogonal annealing conditions suggest that, for the investigated steel, the fast-heating process improved the strength with approximately 90 MPa on average within the elongation ranged from 17 to 27%.

1. Introduction

In the past few decades, much effort has been made to improve the fuel efficiency and serviceability of steels for applications in the automotive industry. Quenching and partitioning (Q&P) steel, a new application of the transformation-induced plasticity (TRIP) effect, possesses good strength and ductility [1–3]. Being a multiphase high-strength steel, its final mechanical properties are mainly affected by the volume fraction, morphology, and distribution of the microstructure constituents, which may include martensite, bainite, retained austenite, and ferrite matrix at the same time. These microstructure constituents are known to be predominantly controlled by the conditions for the austenite transformation [4,5]. In other words, the austenitization parameters such as the heating rate, annealing temperature, and soaking time are the key factors for an optimal microstructural design.

Recently, the application of fast-heating annealing to cold-rolled steels has attracted much attention because of its advantage of simplified rapid-cycle, which leads to an improvement in the energy efficiency [6,7]. At the same time, due to the delay of recrystallization at fast-heating, a non-recrystallized matrix can be introduced during the austenitization, which may change the mechanism of austenite nucleation and growth. In the conventional heating process (3–10 °C/s), the austenitization starts in a nearly recrystallized matrix and nucleation occurs preferentially at the interfaces of ferrite and cementite [8]. During the fast-heating process, a mass of grain boundaries and unreleased distortion energy of the deformed structures are reserved, which provide a large number of austenite nucleation sites and result in a remarkable grain refinement [9,10]. The fast-heating of ultralow carbon steels results in grain homogenization and grain refinement because of a uniform distribution of carbides [7,11,12]. For some fast-heated high-strength steels, the initially formed austenite grains always inherit the morphology of the non-uniform cementite, and exhibit inhomogeneous band structures [13–15]. From the limited literature available on the mechanical properties of the fast-heated high-strength steels, it is known that the fast-heating increases the strength but decreases the elongation to some extent [16,17]. These results were mostly obtained from a set of designed annealing tests and failed to highlight the advantages of fast-heating annealing clearly.

Despite the availability of a large number of reports on the microstructural investigations of the recrystallization and austenitization by fast-heating process, the exact mechanism by which fast-heating affects the final microstructure and mechanical properties of the annealed steels, especially those of the high-

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strength steels is not known yet. Thus, the present paper discusses the effect of fast-heating on the final microstructure and mechanical properties of Q&P steels systematically and comprehensively, and brings insight into the strategies of fast-heating annealing on large-scale production.

2. Experimental

The chemical composition (wt%) of the high-strength steel used in this study is as follows: $C=0.21$, $Si=1.40$, $Mn=1.30$, balanced with Fe. The thickness of the cold-rolled steel sheet was reduced by 70%, and the final thickness was 1.2 mm.

To investigate the austenitization process of the steel samples, the dilatometry study was conducted using a Gleeble 3500 thermomechanical simulator equipped with a dilatometer (Dynamic Systems Inc., Poestenkill, NY). The samples were machined to 10 mm × 70 mm × 1.2 mm sheets and were tested along the longitudinal axis coinciding with the rolling direction. Conventional-heating (5 °C/s) and fast-heating (300 °C/s) were applied during the heating process. After soaking at 790 and 830 °C for 120 s, the samples were heated to 1000 °C for complete austenitization. To analyze the microstructure evolution, the heating of the samples was interrupted at various heating conditions by water quenching.

The Q&P treatments were performed on a customized heat treatment installation with 300 mm × 250 mm × 1.2 mm sheet blanks. The schematic representation of the annealing process is shown in Fig. 1. At the two heating rates (5, 300 °C/s), the cold-rolled strips were heated to five different temperatures ranging from 770 to 850 °C for soaking times ranging from 0 to 120 s. Each sample was then quenched to 260 °C and reheated to 430 °C for a duration of 90 s, and finally cooled down to room temperature.

Microstructural morphology was investigated using a scanning electron microscope (SEM, EVO MA25, ZEISS). Transmission electron microscopy (TEM) analysis of the samples was performed using a JEM-2100F microscope. A JXA-8500F electron probe microanalyzer (EPMA) was used to investigate the carbon distribution in the microstructures. At room temperature, the volume fractions of the retained austenite (RA) and the average carbon content were determined by an X-ray diffractometer (XRD) using a Cu K radiation by estimating the intensities of the peaks corresponding to the $(200)_\gamma$, $(220)_\gamma$, $(311)_\gamma$, $(200)_\alpha$, and $(211)_\alpha$. The austenite lattice a_γ was obtained by the Nelson-Riley extrapolation method [18], and the carbon content in the RA was calculated using the empirical expression, $\alpha_0=3.5467+0.0467C_{RA}$, where α_0 is the lattice constant of the RA, as determined from the $(220)_\gamma$ austenite peak and C_{RA} is the carbon content [5]. The mechanical properties were measured on an INSTRON 5581 tensile testing machine with a crosshead displacement of 2 mm/min, gauge length of 50 mm, gauge width of 12.5 mm, and thickness of 1.2 mm at room temperature.

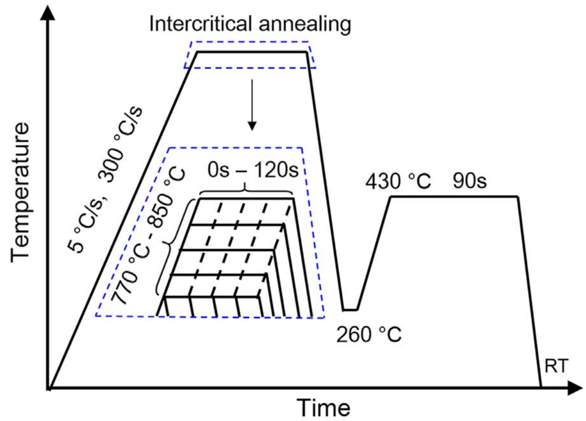

Fig. 1. Schematic heat cycle of Q&P process used in present study.

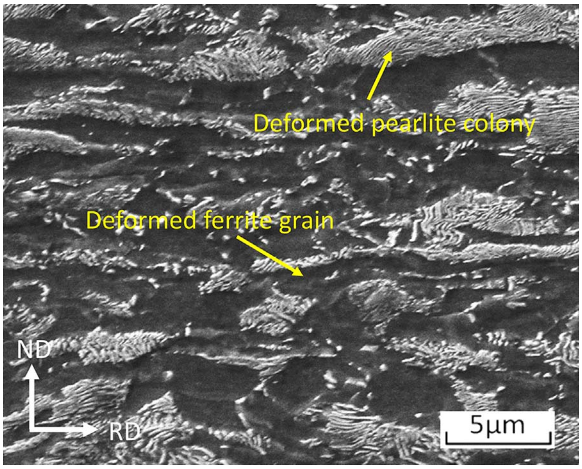

Fig. 2. The microstructure of cold-rolled steel.

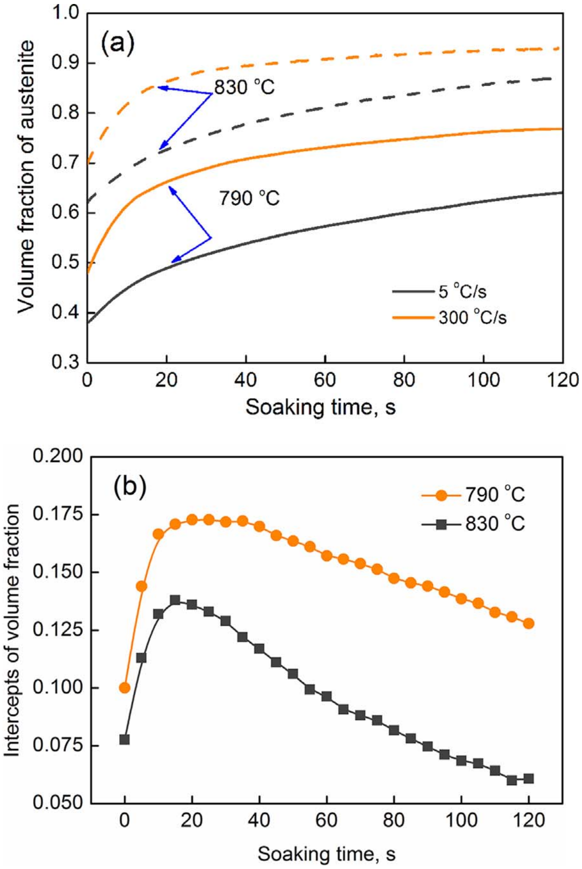

Fig. 3. (a) Austenite formation kinetics and (b) intercepts of the volume fraction of austenite during intercritical annealing at 790 °C and 830 °C.

3. Results and discussion

3.1. Initial microstructure

As observed from Fig. 2, the cold-rolled sample consisted of deformed ferrite (DF) grains and deformed pearlite colonies,