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High-efficiency fast-heating annealing of a cold-rolled dual-phase steel

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ABSTRACT

In the present work, fast-heating annealing was performed on a cold-rolled Fe–0.07C–1.7Mn–0.429Si dual-phase steel. In contrast to commercial conventional continuous annealed steel, 6.6% higher ultimate tensile strength and 14.1% greater elongation were obtained. Transmission electron microscopy observations reveal the incomplete ferrite recrystallization, the formation of bainite and fine fiber-like martensite in fast-heating processed dual-phase steel. High-efficiency simplified fast-heating annealing process demonstrates great potential for large-scale production.

1. Introduction

Dual-phase (DP) steels developed from low carbon steels, with a good combination of strength and ductility for enhanced formability, are widely used in automobile industry due to the urgent need to improve fuel efficiency [1,2]. As martensite phase imparts high strength while the ferrite matrix provides excellent elongation in various low-carbon high-strength steels [3], cold-rolled DP steels [4] and austenite/ε-martensite DP steels [5], the steel mechanical properties depend on the volume fraction, morphology and distribution of martensite in ferrite matrix, which are predominantly controlled by intercritical annealing [6,7]. Recently, fast-heating annealing (up to 500 K/s) attracted much attention due to its advantage of simplified rapid-cycle, which may move up to a new level of energy efficiency [8–11].

In case of conventional continuous annealing (3–5 K/s), ferrite recrystallization first occurs during heating, followed with austenite formation at the intercritical region [12,13]. The austenite preferentially nucleates close to the ferrite–cementite interfaces, and its subsequent growth is mainly governed by the competition between the simultaneous nucleation at ferrite grain boundaries and the growth of prior austenite [14]. In case of fast-heating, however, there would be a strong interaction between ferrite recrystallization and austenite formation, and thus the kinetics of austenite formation is affected, including spatial distribution and morphology of martensite [15,16]. In the present study, we applied fast heating processing to a cold-rolled DP sheet steel and transmission electron microscopy (TEM) was employed to clarify the interactions between these phases in order to explore the possibility of fast-heating routine on large-scale production.

2. Experimental details

An industrial Fe–0.07C–1.7Mn–0.429Si (wt.%) dual-phase sheet steel (DP590) with a cold-rolled ratio of 70% was investigated in the present study. Dilatometer study indicated the start temperature A_{c1} and the end temperature A_{c3} of the pearlite-to-austenite transformation are 1023 K and 1193 K at the heating rate of 500 K/s, respectively (not shown here). Specimen with dimension of $300 \times 260 \times 1.5\ mm^3$ was rapidly heated to 1133 K on a self-designed fast-heating testing bench, soaking for 2 s and then cooled down to 323 K, as shown in Fig. 1. The same DP590 steel commercially manufactured through continuous intercritical annealing was also studied for comparison. Hereafter we refer these two samples as fast heating (FH) and continuous annealing (CA), respectively.

Three tensile samples were cut parallel to the rolling direction for each annealing routine and tensile testing was performed according to the ISO 6892-1:2002 standard [17] at room temperature with a crosshead displacement rate of 30 mm/min (Instron 5581). The gauge length was 80 mm and the width was 20 mm. Microstructural examinations were performed using an optical microscope (Zeiss Axio Imager A2m). Quantitative phase measurements were conducted using a point counting method. TEM analysis was carried out using a JEM-2010F microscope operating at 200 kV using a double-tilt stage. The thin foils were prepared

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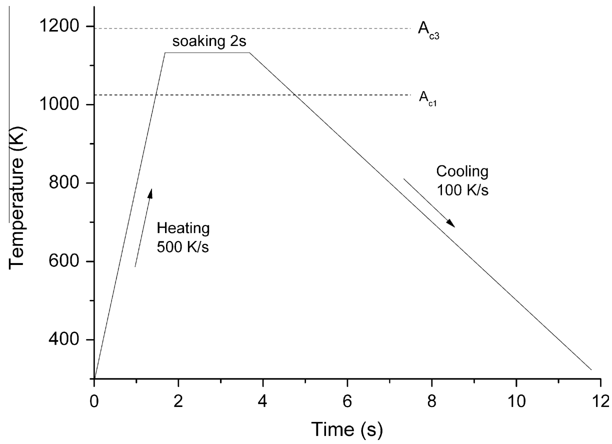

Fig. 1. Fast-heating annealing schedule for the cold-rolled DP590 sheet steel.

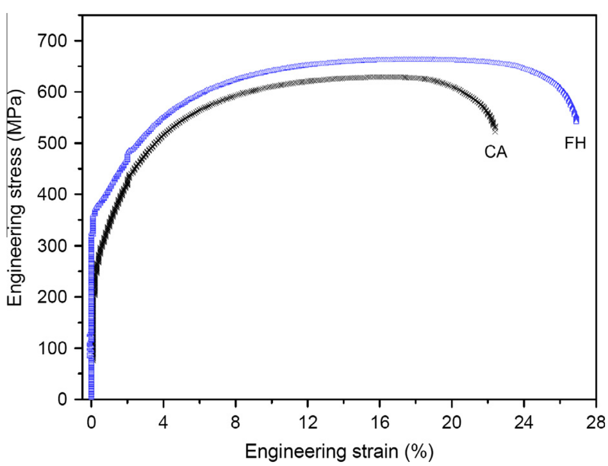

Fig. 2. Two typical engineering stress–strain curves of DP590 sheet steel with different heating routines.

Table 1
Tensile data of DP590 steel with continuous annealing (CA) and fast heating (FH) routines.

Sample	Yield strength $R_{P0.2}$ (MPa)	Ultimate tensile strength R_m (MPa)	Uniform elongation (%)	Total elongation A_{80mm} (%)
CA	277 ± 8.1	625 ± 3.6	16.5 ± 0.2	23.3 ± 0.8
FH	372 ± 3.0	666 ± 2.6	18.0 ± 0.5	26.6 ± 0.5

according to standard metallographic practice and double-jet electrochemical thinning.

3. Results and discussion

Fig. 2 presents two typical tensile engineering stress–strain curves obtained from the DP steel subjected to different annealing routines. Both curves show clear discontinuous yielding characters of DP steel. As listed in Table 1, it is found that the average yield strength $R_{p0.2}$ of FH samples increases from 277 MPa to 372 MPa (34.3% higher), and the ultimate tensile strength R_m improves from 625 to 666 MPa with 41 MPa higher (6.6%). It is interesting that the total elongation A_{80mm} of FH samples is enhanced from 23.3% to 26.6% (with 14.1% increase), and its uniform elongation, from 16.5% to 18.0% (with 9.1% increase). However in most previous reports, the improved strength for steels was always accompanied with decreased elongation [18–20]. The standard deviations of mechanical properties obtained from FH samples show smaller scatter except the uniform elongation.

As discussed in the introduction section, the mechanical properties are closely linked to the dual-phase microstructure [3,8–10]. The starting cold-rolled DP590 steel contains elongated ferrite grains and pearlite colonies [8], and the optical microstructures of two samples with different annealing routines are presented in Fig. 3. Compared with the CA sample in Fig. 3a, the average ferrite grain sizes in the FH sample refined strikingly from about 11.2–4.3 μm (Fig. 3b), and the black secondary phases changed from coarse lath-like martensite (Fig. 3a) to fine fiber-like martensite and bainite characterized based on the following TEM analyses. The improvement of strength is thus related to the ferrite grain refinement based on the classical Hall–Pecht relation [9,21]. Meanwhile, it should be noticed that a quantitative measurement indicates that the content of martensite increases slightly from 23 vol.% (percentage of volume fraction) to 27 vol.%, which could also play a positive role on the strength in the FH samples in addition to the grain refinement effect.

Fig. 4a shows the TEM image of the CA sample, where the ferrite matrix and grain boundaries are almost dislocation-free, and martensite shows coarse lath-like feature. In case of FH condition, dislocations inside the ferrite matrix are clearly visible (Fig. 4b), which may inherit from the cold-rolled state. Similarly the overall duration of fast-heating process is less than 12 s and the time above A_{c1} is only about 3.3 s, so it is not expected to finish the fully austenite formation. Fig. 4c and d shows the fine and fibrous morphology of martensite, which is also favorable to improve the strength. In addition, some portion of bainite (Fig. 4c and d) formed in the vicinity of martensite, which can be well explained by means of insufficient carbon diffusion and redistribution. Upon cooling, the bainite transforms from the carbon-deleted regions [22].

During the conventional continuous annealing with slow heating rates (3–5 K/s), the ferrite recrystallization would fully

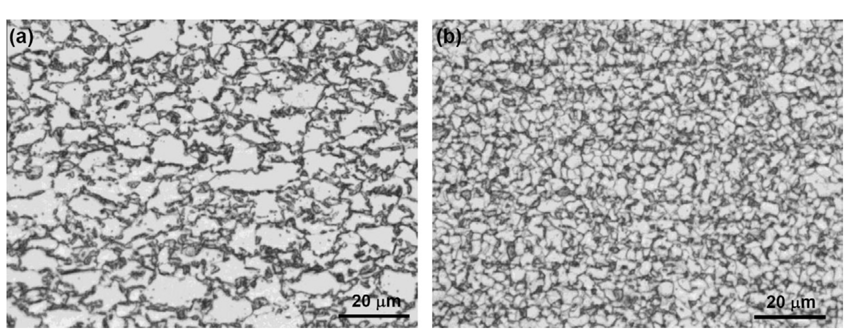

Fig. 3. Optical micrographs of annealed DP590 steel, CA-Continuous annealing (a) and FH- Fast heating (b).