Materials Science & Engineering A 674 (2016) 151–157

Contents lists available at ScienceDirect

Materials Science & Engineering A

journal homepage: www.elsevier.com/locate/msea

Heat treatment, microstructure and mechanical properties of a C–Mn–Al–P hot dip galvanizing TRIP steel

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ARTICLE INFO

Article history:
Received 30 May 2016
Received in revised form
25 July 2016
Accepted 27 July 2016
Available online 28 July 2016

Keywords:
Hot Dip Galvanizing TRIP Steel
Mechanical Properties
Isothermal bainitic transformation
Microstructure
Heat treatment

ABSTRACT

Heat treatments of a hot dip galvanizing TRIP (Transformation induced plasticity) steel with chemical composition 0.20C-1.50Mn-1.2Al-0.07P(mass%) were performed in a Gleeble 3500 laboratory equipment. The heat treatment process parameters were varied to investigate the effect of intercritical annealing temperature as well as isothermal bainitic transformation (IBT) temperature and time, on the microstructure and the mechanical properties. The microstructure was investigated using scanning electron microscopy, transmission electron microscopy and x-ray diffraction, while mechanical properties were evaluated by tensile testing. Furthermore, to generate a better understanding of the phase transformations during heat treatment, dilatometry trials were conducted. The desired microstructure containing ferrite, bainite, retained austenite and martensite was obtained after the heat treatments. It was further found that the IBT is critical in determining the mechanical properties of the steel, since it controls the fraction of bainite. With increasing bainite fraction, the fraction of retained austenite increases while the fraction of martensite decreases. The mechanical properties of the steel are excellent with a tensile strength above 780 MPa (expect in one case) and elongation above 22%.

1. Introduction

TRIP (Transformation Induced Plasticity) steels generally have high strength in combination with excellent ductility [1,2]. Their high energy absorption in high strain rate deformation [1,3] and excellent fatigue properties make them attractive in automotive applications [4,5]. The mechanical properties are brought about by the multi-phase microstructure consisting of ferrite, bainite, retained austenite and possibly a small fraction of martensite, where the retained austenite is responsible for the TRIP effect by its transformation to martensite during deformation [6]. Furthermore, The complex metastable microstructure plays a critical role in their damage behavior with several different fracture modes [7,8]. The multi-phase microstructure is usually generated by a standard two stage heat-treatment [9]. The first stage is intercritical annealing (IA) where a controlled volume fraction of ferrite is formed. The IA is followed by an isothermal stage in the bainitic transformation temperature range. During the second stage, partial transformation of austenite to bainite and the concomitant carbon partitioning to the austenite phase leads to a stabilization of the remaining austenite at room temperature [10]. Thus, the isothermal bainitic transformation (IBT) time and temperature at the second stage of the heat treatment have a decisive influence on the microstructure and the mechanical properties of the TRIP steel [11].

There are several challenges in the continuous galvanizing process for TRIP steels. First, the heat treatments conventionally used to produce the desired TRIP steel multi-phase microstructure are not necessarily compatible with the hot-dip galvanizing thermal cycle, since the IBT time for hot-dip galvanizing lines is short and the range of IBT temperature is limited due to the requirements for the Zn bath [12]. Moreover, selective oxidation of alloying elements such as Si during the annealing prior to galvanizing can result in poor reactive wetting and unacceptable bare spots in the Zn coating.

Conventional TRIP steels contain roughly 0.15C, 1.5Si and 1.5Mn (mass%) but the high Si content is known to give rise to galvanizing problems [13]. Si is also known to increase the ductile-to-brittle transition temperature of ferritic steels which may result in low ductility levels in the as-cast condition [14]. These are the main reasons to keep the Si content of the TRIP steels as low as

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http://dx.doi.org/10.1016/j.msea.2016.07.119
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possible and, hence, the replacement of Si with other alloying elements is an important issue in alloy design of advanced TRIP steels [15]. Al has been actively investigated as a candidate to replace Si since it is known to suppress carbide formation in austenite. However, Al generates little solid-solution hardening of the ferrite phase, and thus the substitution of Si by Al significantly reduces the strength. Therefore, when using Al to replace Si the solid-solution strengthening must instead come from, for instance, micro-alloying elements such as Nb or V [11,16]. It should be noted, however, that Al cannot fully replace Si since the ability of Al to suppress cementite precipitation is lower than for Si [10,17].

Because of these disadvantages, the potential of P addition to TRIP steels where the Si is partially replaced by Al has been evaluated. P has been reported to cause grain boundary embrittlement, but it is an effective element to provide solid-solution hardening of ferrite, and has been reported to promote carbon diffusion into the austenite by increasing the carbon activity in the ferrite [18]. Chen et al. [18] investigated the effect of P additions in Si-free and Si-containing low carbon TRIP steel, their results showed a beneficial effect of P on the retention of metastable austenite, especially in combination with Si. Barbé et al. [19] studied the effect of P addition on the mechanical properties of TRIP steels and their results showed that P and Si have a synergistic effect to enhance the tensile strength. Furthermore, Jing et al. [20] have studied the microstructure and mechanical properties of C–Mn–Al–P TRIP steels, but in that case Cu was added to the steel. To the authors' knowledge there are no previous reports on C–Mn–Al–P TRIP steels without any other alloying elements. Hence, the purpose of this report is to study the effect of continuous galvanizing heat treatments on the microstructure and mechanical properties of a C–Mn–Al–P TRIP steel, and to compare with an alloy without P.

2. Experimental procedure

The chemical compositions of the hot dip galvanizing TRIP steels used in this study are given in Table 1. The mechanical properties of the C–Mn–Al–P steel were compared with a reference steel prepared in the same way but without any P addition. 50 kg ingots were prepared in an air induction furnace and forged into 60 mm thick blocks. The blocks were subsequently soaked at 1200 °C for 1.5 h, and hot rolled to the thickness of 4.2 mm in six passes. The hot-rolled plates were pickled in 25 vol% HCl solution and cold-rolled to 1.4 mm thickness.

To obtain the TRIP steel microstructure, a two-step heat treatment was used. The samples, 200 mm in length and 50 mm in width, were heat treated in a Gleeble3500 equipment. The heat treatment cycles imposed were: IA for 60 s, cooling at 20 °C s^{-1} to the IBT temperature, holding for 20, 30, or 60 s, and cooling to room temperature, a schematic illustration of the heat treatment cycle can be seen in Fig. 1.

Tensile specimens from the heat-treated samples were prepared with a dog-bone shape and a gauge length and width of 50 and 12.5 mm, respectively. Tensile tests were carried out in a MTS universal testing machine with a load range of 5000 kgf. The cross head speed was maintained at 2 mm/min.

Table 1
Chemical composition (mass%) of the steels.

Steel	C	Mn	Si	Al	P	Nb	V
C–Mn–Al–P	0.21	1.58	0.05	1.16	0.068	–	–
C–Mn–Si–Al	0.20	1.50	0.26	1.20	0.007	0.05	0.08

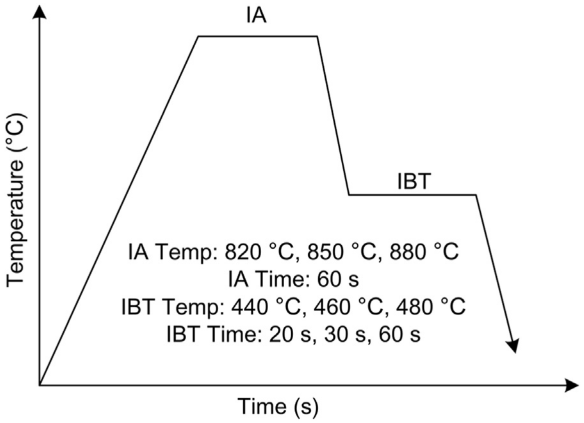

Fig. 1. Schematic illustration of the heat treatment cycle.

The microstructures were studied by scanning electron microscopy (SEM) after conventional metallographic sample preparation and Nital etching. Transmission electron microscopy (TEM) specimens were prepared by mechanical polishing and electropolishing in a twin-jet polisher using 5% perchloric acid solution at −20 °C. The TEM investigations were performed using a JEOL 2010 instrument at an acceleration voltage of 200 kV.

The volume fraction of retained austenite in the heat-treated samples was measured by X-ray diffraction (XRD) using Cu Kα radiation. Samples for XRD were prepared by first mechanical polishing, then, electropolishing at 15 V at room temperature and using an electrolyte consisting of 10 vol% glycerol, 20 vol% perchloric acid and 70 vol% ethanol. The integrated intensity (I) of the (200) and (211) peaks of bcc and the (200), (220) and (311) peaks of fcc were determined by using least-square fitting of a Pearson VII type function. The volume fraction of retained austenite was subsequently calculated using the direct comparison method according to Eq. (1) [21].

$$V_\gamma = \left[\frac{1}{q}\sum_{j=1}^{q}\frac{I_{\gamma j}}{R_{\gamma j}}\right] \Big/ \left[\frac{1}{q}\sum_{j=1}^{q}\frac{I_{\gamma j}}{R_{\gamma j}} + \frac{1}{p}\sum_{i=1}^{p}\frac{I_{\alpha i}}{R_{\alpha i}}\right] \times 100\% \tag{1}$$

V_γ is the volume fraction of austenite and R is the calculated theoretical intensity. The values of R are collected in Table 2.

For the calculation of the carbon content in the retained austenite (C_γ), the lattice parameter a_γ was determined using the fcc (220) diffraction peak and Eq. (2) [12]. The equation is considering the effect of Mn, Si [22] and Al [23] on the lattice parameter a_γ.

$$a_\gamma = 3.572 + 0.033 \times wt\%C + 0.0012 \times wt\%Mn - 0.00157 \times wt\%Si + 0.0056 \times wt\%Al \tag{2}$$

The dilatometry was conducted using a DIL 805 instrument. Cylindrical specimens with 4 mm diameter and 10 mm length were used.

3. Results and discussion

3.1. Microstructure

3.1.1. General microstructure

Fig. 2 shows an SEM micrograph of the C–Mn–Al–P steel. It can

Table 2
Values of $R_{\gamma(hkl)}$ and $R_{\alpha(hkl)}$.

$R_{\gamma\ (200)}$	$R_{\gamma\ (220)}$	$R_{\gamma\ (311)}$	$R_{\alpha\ (200)}$	$R_{\alpha\ (211)}$
63.3	32.3	37.4	15.3	46.4