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Abstract

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Keywords

steel, produced, strip, casting, simulated, mechanical, microstructures, properties, laboratory, trip

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Microstructures and Mechanical Properties of TRIP Steel Produced by Strip Casting Simulated in the Laboratory

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Abstract: Conventional transformation induced plasticity (TRIP) steel (0.17C-1.52Si-1.61Mn-0.03Al, wt. %) was produced via strip casting technology simulated in the laboratory. Effects of holding temperature, holding time and cooling rate on ferrite formation were studied via analysis of the continuous cooling transformation diagram obtained here. A typical microstructure for conventional TRIP steels consisting of ~ 0.55 fraction of polygonal ferrite with bainite, retained austenite and martensite was obtained. However, coarse prior austenite grain size of ~ 80 µm led to large polygonal ferrite grain size of ~ 17 µm, coarse second phase regions of ~ 21 µm size, small amount of retained austenite (0.02 – 0.045) and the presence of Widmanstätten ferrite. Optimisation of the microstructure-property relationship was reached via a variation in the isothermal bainite transformation temperature. The highest retained austenite fraction of 0.045±0.003 with medium carbon content of 1.23±0.01 wt. % was obtained after holding at 400 °C, resulting in the highest ultimate tensile strength of 590±35 MPa and largest total elongation of 0.27±0.05. The presence of TRIP effect in the studied steel was revealed through the analysis of strain hardening exponent and modified Crussard – Jaoul model. Effect of processing parameters on retained austenite retention and stress-strain behaviour was discussed.

Keywords: Strip casting; TRIP steel; Mechanical properties; Retained austenite; Isothermal bainite transformation; Prior austenite grain size.

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1. Introduction

Transformation-induced plasticity (TRIP) steel is one of the advanced high strength steels, which is used in automotive industry [1]. Compared to dual phase (DP) steel, TRIP steel has a better combination of strength and formability in terms of both elongation and strain hardening exponent, which attracted many researchers to study it in recent decades [1, 2]. In order to produce such multi-phase steel, a strict control of processing window parameters is needed, which somewhat restricts a wide production of TRIP steels. The microstructure of TRIP steel consists of polygonal ferrite, bainite, retained austenite (RA) and martensite [3-11]. RA is a key phase which ensures a good combination of high strength and high ductility through the transformation of RA to martensite during deformation [12-16]. The production of TRIP steel includes a two-stage heat treatment following hot rolling or cold rolling [15, 17]. The first stage of heat treatment results in austenite-to-ferrite transformation. At this stage, a TRIP steel is cooled through or held in the austenite-ferrite two phase temperature region. The ferrite fraction is controlled by the adjustment of cooling rate or holding time and temperature [3, 4]. At the second stage, a TRIP steel is fast cooled to bainite temperature region, in order to avoid pearlite formation, and held at a selected temperature for a certain time period to assure bainite formation and austenite enrichment with carbon. The isothermal bainite holding (IBT) temperature and time are used to adjust the retained austenite fraction and its stability [7, 12, 18]. Following cooling to the ambient temperature from the bainite transformation temperature, the microstructure will contain some retained austenite, amount of which will determine the scale of TRIP effect.

At present, most of researchers are focusing on hot rolling and cold rolling followed by intercritically annealing, which are technologies that demonstrated their feasibilities to produce TRIP steel in industry [1]. However, the strip casting is another potential way to produce TRIP steel. It is defined as production of strip steels through direct casting from molten metal in a single process, which may combine subsequent inline hot-rolling by a micro-mill [19]. At present the strip casting is utilised for industrial manufacture of carbon steels, silicon steels and stainless steels, and also lead, aluminium and brass strips [20, 21]. Besides, a conventional DP steel was produced using strip casting technology in the laboratory [22, 23]. Due to the strips being directly manufactured from liquid metal, the strip casting technology eliminates many subsequent processing stages inherent for conventional processing routes, such as multi-pass rough rolling of continuously cast thick slab and finishing rolling in continuous rolling mills. Therefore, the strip casting process has many advantages, such as energy saving and emission reduction, lower capital and operating costs, a more flexible operating regime, a shorter production line, smaller amount of residual scrap and easy adjustment of equipment for different steel grades [21, 24].

In this study, a laboratory simulation of strip casting to produce 0.17C-1.52Si-1.61Mn-0.03Al wt. % TRIP steel was carried out for the first time. The average grain size and grain size distribution of prior austenite were simulated being equal to those of cast products. In order to obtain typical microstructures of TRIP steels, a two-stage heat treatment was carried out: first stage – for ferrite formation; the second stage – for bainite formation. The microstructures

were studied using optical, scanning and transmission electron microscopy, X-ray diffraction, electron backscattered diffraction, and atom probe tomography. Mechanical properties measured via tensile testing were correlated to the microstructural parameters and process variations.

2. Materials and experimental techniques

2.1 Processing route

The chemical composition of studied steel is listed in Table 1. As-cast specimens of $36 \times 36 \text{ mm}^2$ and 1.2 mm thickness were produced using dip tester at Deakin University [25], which can simulate fast cooling during solidification between twin rolls in strip casting process [24, 26]. The dip tester is used to immerse a copper substrate into molten steel for a short and controlled period of time and then immediately lift it out to simulate rapid solidification [27]. Microstructure after casting contained martensite and bainite (Fig. 1).

Table 1 The chemical composition of the studied steel (wt. %).

C	Si	Mn	Al	Cu	Cr	P	S	B
0.172	1.520	1.610	0.0266	0.0153	0.195	0.0044	<0.00050	0.00057

Figure 1 As-cast microstructure of studied steel.

A Theta Dilatronic III dilatometer was used to carry out heat treatments. To prevent oxidation or decarburization of the samples, it was operated under vacuum of $\sim 6.7 \times 10^{-2} \text{ Pa}$. The temperature was monitored using an S-type (Pt/Pt-10%Rh) thermocouple spot-welded to the surface centre of cylindrical and flat samples. The cylindrical samples had an outer diameter of 5 mm and an inner diameter of 3.5 mm with the length of 10 mm, and were used for determination of the continuous cooling transformation (CCT) diagram. The flat samples of $14 \times 6 \text{ mm}^2$ and $\sim 1 \text{ mm}$ thickness with small holders at each port were used for the other experiments.

To simulate the prior austenite grain size (PAGS) observed in as-cast specimen, the flat samples were heated at a rate of 30 Ks^{-1} to 1250 or 1300 °C, held for 180 or 300 s, and then helium quenched to room temperature at a cooling rate of approximately 140 Ks^{-1} . Noticeably, the sample melted when it was heated to 1300 °C, thus 1250 °C austenitising temperature was selected for further experiments.

To obtain the CCT diagram, cylindrical samples were heated to 1250 °C at a rate of 30 Ks^{-1} , held for 300 s, and then cooled to ambient temperature at cooling rates of 1, 10, 40 or 110 Ks^{-1} .

The schedule used to simulate the strip casting process is illustrated in Fig. 2. The flat samples were heated at 30 Ks^{-1} to austenitisation temperature $T_A=1250 \text{ °C}$, held for time $t_A=300 \text{ s}$, cooled at a rate of 90 Ks^{-1} to austenite-to-ferrite transformation temperature or cooled to 1000 °C at a rate of 30 Ks^{-1} followed by cooling at a rate of 10 Ks^{-1} to ferrite formation temperature T_F (630, 650 and 670 °C) (hereafter referred to as 30-10 Ks^{-1} schedule), held for time t_F (0 – 300 s), quickly cooled at a rate of 50 Ks^{-1} to IBT temperatures T_{IBT} of 350, 400, 450 and 500 °C, held for time $t_{IBT} = 900 \text{ s}$ at these temperatures, and helium quenched to room temperature at a rate of $\sim 140 \text{ Ks}^{-1}$.

Figure 2 A schematic diagram of heat treatments to simulate strip casting.

2.2 Microstructure characterisation

2.2.1 Optical microscopy

The heat treated samples were cross-cut in the centre perpendicular to the long axis. The

central area of the cross-section was used for observations. A Leica DMR research optical microscope (OM) was employed. To reveal the prior austenite grain boundaries, the specimens were etched for 15 – 20 s at a temperature of 65 °C in the solution of saturated picric acid in ethanol plus few drops of benzene sulfonate. Equivalent circle diameter was utilised to describe the grain size distribution of prior austenite. Approximately 150 – 250 grains were measured for each condition.

To observe the room temperature microstructures, the heat treated samples were mechanically polished and then electropolished using an electrolyte of 330 ml methanol, 330 ml butoxyethanol and 40 ml perchloric acid at 50 V, ~ 1.0 mA and 17 °C for ~ 90 s. Microstructures were revealed using etching in 2 vol.% nital. Fractions of phases were calculated based on pixel quantities of different grey scales using Photoshop and Image Pro-Plus softwares. The equivalent circle diameter was calculated using at least 260 polygonal ferrite grains or second phase regions.

In addition, colour etching was employed to distinguish polygonal ferrite, bainite, martensite and retained austenite according to Ref. [28]. The samples were firstly dipped for 18 – 20 s in a solution of 4 % picral (4 gm dry picric acid in 100 ml ethanol) mixed with few drops of concentrated hydrochloric acid (1 ml per 100 ml of picral solution) and then etched for 6 – 8 s with freshly prepared 10 % aqueous sodium metabisulfite solution.

2.2.2 Scanning electron microscopy

Microstructures were studied further using a JEOL JSM-7001F field emission gun – scanning electron microscope (FEG – SEM) operating at 15 kV accelerating voltage and ~ 5.1 nA probe current and fitted with an 80 mm² X-Max energy dispersive X-ray spectroscopy (EDS) detector and a Nordlys-II(S) EBSD detector interfacing with the Oxford Instruments Aztec software suite. Electron backscattering diffraction (EBSD) and EDS information were obtained simultaneously at the centre of the electropolished cross section using a step size of 0.04 µm and a working distance of 12 mm.

2.2.3 Transmission electron microscopy

A JEOL 2011 transmission electron microscope (TEM) operating at 200 kV was used to characterise the microstructure in detail. The 3 mm diameter discs were mechanically polished to ~ 80 µm thickness, and then electropolished using twin-jet electropolishing method with an electrolyte containing 10 vol. % perchloric acid in methanol.

2.2.4 Atom probe tomography

Atom probe tomography (APT) experiment was carried out at Deakin University using a LEAP 3000X HR instrument (CAMECA Instruments Inc.) operating in voltage mode with a 20% pulse fraction, while the needle-like sample was at a temperature of 53 Kelvin and a pressure of 9.6×10^{-9} Pa. The needle-like sample was prepared applying a standard two-stage

electropolishing technique [29]. The data was analysed using IVAS 3.6.8 software.

2.3 X-ray diffraction technique

The RA fraction was measured using a PANalytical X'pert-Pro MRD goniometer equipped with Ni-filtered Cu K_α radiation. It was operated at 40 kV accelerating voltage and 45 mA beam current. The X-ray diffraction (XRD) scans were conducted over range of $2\theta = 38 - 150^\circ$ at a step size of 0.01° and a step time of 750 s on the surface of electropolished samples. The RA fraction was calculated using the integrated intensities of four austenite peaks {(111), (200), (220) and (113)} and four ferrite peaks {(110), (200), (211) and (310)} [13]. The carbon content in RA was determined using the following empirical equation [30]:

$$a = 3.572 + 0.033 \times [C\%] + 0.0012 \times [Mn\%] - 0.00157 \times [Si\%] + 0.0056 \times [Al\%] \quad (1)$$

where a is austenite lattice parameter ($\text{\AA}$) [31], and $[C\%]$, $[Mn\%]$, $[Si\%]$ and $[Al\%]$ represent the concentrations in weight percentage of carbon, manganese, silicon and aluminum, respectively. For convenience, the contents of alloying elements in RA were assumed to be equivalent to the nominal steel's composition for all processing conditions.

2.4 Tensile tests

An in-house modified Kammrath and Weiss GmbH tensile stage was used to carry out tensile tests at a constant crosshead speed of $2 \mu\text{ms}^{-1}$, which assured an initial strain rate of $4 \times 10^{-4} \text{ s}^{-1}$. The dog-bone tensile test samples had 4.9 mm gauge length, 2.1 mm width and ~ 1 mm thickness. At least, two samples were tested for each condition. The strain hardening rate (θ) and strain hardening exponent (n) were calculated as follows,

$$\theta = d\sigma / d\varepsilon \quad (2)$$

$$n = d(\ln \sigma) / d(\ln \varepsilon) \quad (3)$$

where σ is true stress and ε is true strain. The modified Crussard – Jaoul (C – J) model was used to analyse the multistage hardening behaviour in the studied steel [32]:

$$\varepsilon = \varepsilon_0 + c\sigma^m \quad (4)$$

where ε_0 is the initial true strain, c is a constant and m is the strain hardening index. m is calculated using the following equation:

$$\ln(d\sigma / d\varepsilon) = (1 - m) \ln \sigma - \ln(cm) \quad (5)$$

3. Results

3.1 Prior austenite microstructure simulation

Fig. 3 shows typical prior austenite microstructures and PAGS distributions in as-cast and heat treated samples. PAGS increased with an increase in holding time. After holding at 1250°C for 180 and 300 s, the average grain size was 71 ± 28 and $80 \pm 27 \mu\text{m}$, respectively. The average grain size of $80 \pm 27 \mu\text{m}$ after holding at 1250°C for 300 s was similar to as-cast

sample's average grain size of $83 \pm 31 \mu\text{m}$. In addition, the sample held at 1250°C for 300 s had comparable PAGS distribution to that of the as-cast sample (c.f. Figs. 3(c, d)). Thus, this condition was selected for all future experiments (Fig. 2).

Figure 3 Comparison of (a, b) prior austenite microstructures and (c, d) prior austenite grain size distribution (a, c) in as-cast condition and (b, d) as-quenched after holding at 1250°C for 300 s.

3.2 Continuous cooling transformation diagram

To assess the effect of cooling rate on phase balance, the CCT diagram was obtained (Fig. 4) using microstructures shown in Fig. 5. A small amount of ferrite and pearlite, and a large amount of bainite and martensite were observed at the cooling rate of 1 Ks^{-1} . At cooling rates between 10 and 110 Ks^{-1} , the microstructures consisted of bainite and martensite. Bainite was still observed at the highest cooling rate of 110 Ks^{-1} , due to large prior austenite grains [33, 34]. Ferrite transformation start temperature (A_{r1}) decreased from 662 to 625°C when the cooling rates increased from 1 to 10 Ks^{-1} . Bainite transformation start temperature (B_s) first increased from 559 to 625°C , when the cooling rate increased from 1 to 10 Ks^{-1} and then decreased from 625 to 530°C , with an increase in cooling rate to 40 Ks^{-1} . On further increasing cooling rate to 110 Ks^{-1} , the B_s increased again up to 631°C . It corresponds to observations by other researchers [33, 35]. B_s and martensite transformation start temperature (M_s) were calculated as 625 and 415°C , respectively, based on Steven & Haynes's [36] and

Andrews's [37] equations. The calculated B_s and M_s are consistent with the experimental values as shown in Fig. 4.

Figure 4 Continuous cooling transformation diagram of as-cast TRIP steel cooled from 1250 °C after holding for 300 s. F is ferrite, P is pearlite, B is bainite and M is martensite.

Figure 5 Microstructures after austenitising for 300 s at 1250 °C and cooling at (a) 1, (b) 10, (c) 40 and (d) 110 Ks^{-1} . F is ferrite, P is pearlite, B is bainite and M is martensite.

3.3 Effect of holding temperature, holding time and cooling rate on ferrite and pearlite formation

According to the CCT diagram (Fig. 4), three holding temperatures of 630, 650 and 670 °C were selected to investigate ferrite formation. For the cooling rate of 90 Ks⁻¹ from austenitisation temperature to ferrite formation temperature, after holding for 300 s at ferrite formation temperature followed by helium quenching to room temperature, the corresponding microstructures are shown in Fig. 6. The ferrite fractions were 0.46±0.04, 0.56±0.02 and 0.50±0.03 after holding at 630, 650 and 670 °C, respectively. With an increase in holding temperature, the ferrite fraction increased and then decreased, showing a maximum point after holding at 650 °C. A maximum of ferrite fraction corresponds to the nose temperature of ferrite transformation field (~ 650 °C) of the CCT diagram. However, some pearlite was observed after holding at 630 and 650 °C (Figs. 6(a, b, d)), which is adverse to TRIP steel [4, 38]. In addition, the ferrite grain obviously exhibited a more polygonal shape after holding at 670 °C than that after holding at 630 °C. Therefore, the temperature of 670 °C was set as holding temperature at austenite-to-ferrite transformation region.

Figure 6 Microstructures after austenitising at 1250 °C for 300 s, cooling at 90 Ks⁻¹ to ferrite formation region and then holding at (a) 630 °C for 300 s, (b, d) 650 °C for 300 s and (c) 670 °C for 300 s. *F* is ferrite, *P* is pearlite and *M* is martensite.

The influence of holding time for two cooling rates on ferrite fraction is plotted in Fig. 7. For

the cooling schedule of $30\text{-}10\text{ Ks}^{-1}$, the ferrite formation was fast within the first 100 s of holding. After holding for more than 100 s, the formation of ferrite continued at a slow rate. A similar phenomenon was observed in Ref. [3]. After 0 s holding, 0.22 ± 0.01 ferrite was observed, which supported the ferrite formation during cooling from 1250 to 670 °C. Thus, the cooling schedule of $30\text{-}10\text{ Ks}^{-1}$ resulted in a higher ferrite fraction for the same holding time, compared to the cooling rate of 90 Ks^{-1} . In addition, cooling with $30\text{-}10\text{ Ks}^{-1}$ schedule better depicted the industrial processing [24]. Thus, the cooling rate of $30\text{-}10\text{ Ks}^{-1}$ was chosen as the cooling rate from austenitisation temperature to ferrite formation temperature. Based on published data [4, 18], the presence of ~ 0.5 fraction of ferrite resulted in a better combination of mechanical properties in TRIP steels. Therefore, holding at 670 °C for 90 s, which led to 0.53 ± 0.04 ferrite formation, was chosen for austenite-to-ferrite transformation.

Figure 7 The ferrite fraction variation with holding time when holding at 670 °C.

3.4 Microstructure characterisation

After holding at 670 °C for 90 s, samples were quickly cooled at a rate of 50 Ks^{-1} to IBT temperature in order to avoid pearlite formation during cooling [38, 39] and then held for 900 s (in order to simulate the coiling process in industry). Four IBT temperatures of 350, 400, 450 and 500 °C were chosen. Hereafter, TQ refers to the sample directly quenched from 670 °C to room temperature, and T 350, T 400, T 450 and T 500 refer to the samples held at 350, 400, 450 and 500 °C, respectively.

3.4.1 Optical microstructures

Fig. 8 shows the nital- and colour-etched microstructures. Polygonal ferrite, martensite and a small amount of Widmanstätten ferrite were formed in TQ sample (Figs. 8(a, b)) after direct quenching to room temperature from ferrite formation temperature. On the other hand, after the subsequent holding at bainite formation temperature, the samples consisted of polygonal

ferrite, bainite, martensite, retained austenite, and Widmanstätten ferrite (Figs. 8(c-h)).

After holding at 350 °C, very small austenite islands remained after ferrite formation (referred to as remaining austenite hereafter) transformed to bainite, as demonstrated by dark bluish colour (indicated by the arrows) in Fig. 8(d). Very little amount of retained austenite at room temperature (hereafter referred to as retained austenite, RA) located in the vicinity of the interfaces between polygonal ferrite and second phase region. When the IBT temperature increased to 400 °C, the colour etched images showed less of small dark bluish islands and more RA adjacent to the interfaces between polygonal ferrite and second phase region (Fig. 8(f)). In addition, some RA was surrounded by polygonal ferrite, and some blocky martensite was revealed adjacent to the interfaces between polygonal ferrite and second phase region. With an increase in IBT temperature to 450 °C, the colour etching image showed more martensite (brown islands in Fig. 8(h)). Noticeably, more RA adjacent to the interfaces between polygonal ferrite and second phase region was observed in T 400 sample, whereas more martensite islands in T 450 sample were at these locations (c.f. Figs. 8 (f, h)). In addition, the small second phase region in T 450 sample was composed of more martensite and less RA compared to the T 400 sample (c.f. Figs. 8 (f, h)).

Figure 8 Optical and SEM images of microstructures after directly quenching following the holding at 670 °C for 90 s; (c, e, g) nital and (d, f, h) colour etched microstructures after holding at (c, d) 350 °C, (e, f) 400 °C and (g, h) 450 °C for 900 s (15 min). Ferrite is shown with grey/white, bainite with dark grey/bluish, retained austenite with grey/white and martensite with brown colours in (d, f, h). PF is polygonal ferrite, B is bainite, M is martensite, RA is retained austenite and WF is Widmanstätten ferrite.

Figure 8 Optical and SEM images of microstructures after directly quenching following the holding at 670 °C for 90 s; (c, e, g) nital and (d, f, h) colour etched microstructures after holding at (c, d) 350 °C, (e, f) 400 °C and (g, h) 450 °C for 900 s (15 min). Ferrite is shown with grey/white, bainite with dark grey/bruish, retained austenite with grey/white and martensite with brown colours in (d, f, h). *PF* is polygonal ferrite, *B* is bainite, *M* is martensite, *RA* is retained austenite and *WF* is Widmanstätten ferrite.

In the T 400 sample, the average polygonal ferrite grain size was $17\pm 10\ \mu\text{m}$ while the average size of second phase region was $21\pm 24\ \mu\text{m}$. The other samples had a very similar polygonal ferrite average grain size and the size of second phase region. The distributions of the sizes in T 400 sample are shown as a representative of all the samples in Fig. 9. Polygonal ferrite exhibited a normal distribution and the second phase region showed an exponential distribution. The inhomogeneity of the size distribution of second phase region was related to large area percentage of the second phase islands larger than $100\ \mu\text{m}$.

Figure 9 (a) Ferrite grain size distribution and (b) the size distribution of second phase region for T 400 sample.

3.4.2 X-ray diffraction analysis

Fig. 10(a) shows the XRD patterns for different heat treated samples. After IBT holding, the face centred cubic phase was detected, which revealed the existence of RA. On the other hand, RA was not detected in TQ sample. The RA fraction increased from 0.025 ± 0.002 to 0.045 ± 0.003 and then decreased to 0.023 ± 0.002 when IBT increased from 350 to 400 and 450 $^{\circ}\text{C}$ (Fig. 10(b)). The carbon content in RA decreased from 1.36 ± 0.01 to $0.92\pm 0.01\ \text{wt.}\%$.

Figure 10 (a) X-ray diffraction patterns after different heat treatments; (b) retained austenite fraction and carbon content as a function of bainite holding temperature.

3.4.3 Scanning and transmission electron microscopy observations

After holding at 350 °C, second phase regions consisted of predominantly bainitic ferrite, some granular bainite, and martensite and/or retained austenite islands (Fig. 11(a)). Bainitic ferrite is defined as ferrite laths separated by film RA and/or martensite; whereas, granular bainite is the arrangement of carbide-free bainite with irregular-shaped ferrite and dispersed blocky RA/martensite constituent [40]. Small second phase regions (3 – 5 μm in equivalent circle diameter) transformed to bainitic ferrite or granular bainite during IBT holding (Fig. 11(b)), which support the dark bluish bainite areas observed in Fig. 8(d). When the islands were smaller than 1 μm , they were martensite and/or RA islands (Fig. 11(b)). Some of the martensite and RA islands were located at the polygonal ferrite/second phase region interfaces.

Figure 11 Typical SEM images of (a) overview and (b) small second phase regions after holding at 350 °C for 900 s (15 min). *PF* is polygonal ferrite, *GB* is granular bainite, *BF* is bainitic ferrite, and *M/RA* is martensite and/or retained austenite.

When the holding temperature was 400 °C, the microstructure of large second phase regions ($> 20 \mu\text{m}$) consisted of predominately bainitic ferrite, some martensite and/or RA islands embedded in them (Fig. 12(b)). Small second phase regions ($< 20 \mu\text{m}$) included bainitic ferrite and/or granular bainite (Figs. 12(c, d)). Bainitic ferrite laths in large second phase region showed multi-orientations (Fig. 12(b)); whereas, in small second phase region bainitic ferrite exhibited only one orientation (Fig. 12(d)). Very small martensite and/or RA islands, indicated by an arrow in Fig. 12(c), were observed within polygonal ferrite grains. Some martensite or RA islands were located between polygonal ferrite and second phase region, which corresponded to the observation in Fig. 8(f). Film martensite or RA could also be seen between bainitic ferrite laths.

Figure 12 Typical SEM images of (a) overview, (b) large second phase region and (c, d) small second phase regions after holding at 400 °C for 900 s (15 min). *PF* is polygonal ferrite, *GB* is granular bainite, *BF* is bainitic ferrite, and *M/RA* is martensite and/or retained austenite.

Film RA of smaller than 200 nm size (Figs. 13 (a, b)) was observed between bainitic ferrite laths in the large second phase regions (such as the one in Fig. 12(a)). The RA near the polygonal ferrite/second phase region interfaces and the RA in granular bainite were observed in Fig. 13(c). Martensite was also observed adjacent to the polygonal ferrite/second phase region interface (Fig. 13(d)). The TEM observations (Fig. 13) corresponded to those presented in the colour etching images in Fig. 8(f) and SEM images in Fig. 12. An increased dislocation density near the martensite-ferrite grain boundary (Fig. 13(d)) is clearly visible.

Figure 13 TEM images after holding at 400 °C for 900 s (15 min): (a) bright field and (b) dark field images showing retained austenite between bainitic ferrite laths (the zone axis of inset in (a) is $[\bar{1}33]_{\gamma}$, and dark field image was taken from $(02\bar{2})_{\gamma}$; (c) retained austenite and (d) martensite adjacent to the interfaces between polygonal ferrite and second phase region (the zone axis of insets in (c, d) are $[\bar{1}13]_{\gamma}$ and $[011]_{\alpha}$, respectively). *PF* is polygonal ferrite, *GB* is granular bainite, *BF* is bainitic ferrite, *RA* is retained austenite and *M* is martensite.

For the 450 °C IBT temperature, the microstructure was similar to that of T 400 sample (c.f. Figs. 12 and 14). The largest difference observed was the presence of serrated interfaces between polygonal ferrite and second phase regions (indicated by a circle in Fig. 14(c)). They probably contained martensite. Adjacent to the polygonal ferrite/second phase region interfaces, smaller blocky RA and larger blocky martensite were observed in granular bainite (Fig. 15(a)). Fig. 15(b) demonstrated a film martensite (instead of RA) present between bainitic ferrite laths, which may indicate transformation of unstable RA to martensite during cooling from IBT to ambient temperature. Twinned martensite was also observed, and a blocky martensite with small RA was present near the twinned martensite (Fig. 15(c)), which may be the microstructure similar to the circled area in Fig. 14(c). The twinned martensite forms from higher carbon content austenite and at lower temperature compared to martensite with only dislocation substructure [41]. In addition, a blocky martensite was observed, along with small blocky RA (Fig. 15(d)) corresponding to the martensite/RA constituent in Figs.

8(h) and 14(b).

Figure 14 Typical SEM images of (a) large second phase region and (b, c, d) small second phase regions after holding at 450 °C for 900 s (15 min). *PF* is polygonal ferrite, *GB* is granular bainite, *BF* is bainitic ferrite and *M/RA* is martensite and/or retained austenite.

Figure 15 TEM images of microstructures after holding at 450 °C for 900 s (15 min): (a) retained austenite and martensite near the interface between polygonal ferrite and second phase region; (b) martensite between bainitic ferrite laths (the zone axis of BF and M is $[\bar{1}33]_{\alpha}$ and $[012]_{\alpha}$, respectively); (c) twinned martensite; (d) martensite island with some retained austenite among polygonal ferrite. *PF* is polygonal ferrite, *GB* is granular bainite, *BF* is bainitic ferrite, *RA* is retained austenite, *M* is martensite and *TM* is twinned martensite.

With increasing IBT temperature up to 500 °C, some pearlite was observed by optical microscopy (Fig. 16(a)), which was revealed as a lamellar structure using SEM (Fig. 16(b)). Some large martensite islands were also observed (Fig. 16(b)).

Figure 16 (a) Optical and (b) SEM images after holding at 500 °C for 900 s (15 min). *PF* is polygonal ferrite, *P* is pearlite and *M* is martensite.

3.4.4 EBSD and APT analysis of the T 400 sample

The microstructure of T 400 sample was further studied using EBSD and APT because this sample contained the largest RA fraction. Fig. 17 shows the EBSD maps and partly corresponding EDS maps of carbon and manganese. In the larger second phase region, RA was mainly located near the interfaces between second phase region and polygonal ferrite; although, some very tiny RA between bainitic ferrite laths or between bainitic packets were also observed (Fig. 17(b)). In the smaller second phase region, more RA was observed in bainitic ferrite and granular bainite (Fig. 17(b)). Based on this, the smaller second phase region contained more RA, however, it should be noted that only RA larger than 120 nm could be detected by EBSD due to the step size of 40 nm used. The sensitivity of EDS/EBSD simultaneous approach to carbon was proved by correlative electron microscopy study on TRIP steel [40]. As could be clearly seen in Fig. 17(c), the polygonal ferrite is depleted in carbon compared to bainite regions. The grain boundary between polygonal ferrite and second phase region had a larger carbon content (Fig. 17 (c)), which was also observed in Ref. [42]. The RA regions were carbon enriched (Fig. 17(c)). Some carbon enriched areas which did not correspond to RA in phase map were martensite (c.f. Figs. 17(c, d)). On the other hand, manganese was nearly homogeneously distributed (Fig. 17(d)). This can be a result of a low tendency for partitioning of substitutional alloying elements [43] and also of the less variation in Mn content between different phases compared to the order of magnitude difference for carbon [44]. The rectangle area in Fig. 17(c) was contaminated by carbon during the focusing process.

Figure 17 T 400 sample's EBSD maps of (a) the band contrast with grain boundaries (black = high angle grain boundaries, blue = low angle grain boundaries) and (b) phase distribution (blue = fcc austenite, red = bcc ferrite), and corresponding EDS maps of (c) carbon and (d) manganese for the area below the dash line in (a). The area marked by a rectangle in (c) was contaminated by carbon during focusing.

Although SEM and TEM data did not show the presence of any carbides, atom probe revealed several carbon segregations (Fig. 18(a)). The carbon depleted area having 0.6 at.% carbon was bainitic ferrite lath according to published researches [39, 45]. The carbon content in this enriched area was approximately 25 at.% (Figs. 18(b, c)) and corresponds to that of the cementite [46]. As can be seen in Fig. 18(c), the carbon content rapidly increased across the interface between bainitic ferrite lath and cementite while the silicon content obviously decreased due to its rejection from cementite [13, 18]. A slight enrichment of Mn in cementite is also present.

Figure 18 Atom probe tomography characterisation of T 400 sample: (a) carbon atom map, (b) carbon iso concentration surface at 12.0 at. % and (c) a proximity histogram across the bainitic ferrite lath/cementite interface indicated by the arrow in (b). *BF* is bainitic ferrite.

3.5 Mechanical properties

Fig. 19 shows mechanical behaviours of differently heat treated samples, while corresponding mechanical properties extracted from the engineering strain – engineering stress curves are listed in Table 2. Thus, the T 500 sample was not tensile tested due to the harmful effect of pearlite on mechanical properties via a decrease in RA fraction [4, 38]. Continuous yielding was observed for all conditions (Fig. 19(a)), which is characteristic of multiphase microstructures [47]. Thus, the yield stress (YS) was determined using 0.2 % offset. The TQ sample exhibited a typical behaviour of ferrite-martensite DP steel [6, 22]: larger YS and ultimate tensile strength (UTS), and lower uniform elongation (UE) and total elongation (TE) compared to the samples held at bainite transformation temperature. Amongst samples held for bainite formation, the T 400 sample showed the largest UTS and TE, leading to the best combination of mechanical properties (Table 2). The strain hardening rate for all conditions showed an exponential decrease with an increase in true strain (Fig. 19(b)). With respect to IBT held samples, the strain hardening rate of TQ sample decreased more slowly at lower strains but much faster at higher strains. Amongst IBT held samples, the strain hardening rate of T 400 sample decreased faster at lower strains but more slowly at higher strains than for other conditions.

Figure 19 (a) Engineering stress – strain curves, (b) strain hardening rate and (c) instantaneous strain hardening exponent n as a function of true strain, and (d) strain hardening behaviours based on the modified Crussard – Jaoul model analysis, where three distinct stages of strain hardening were denoted by the vertical lines and numbers.

Table 2 Mechanical properties.

Specimen	Ferrite fraction	Yield strength, MPa	Ultimate tensile strength, MPa	Uniform elongation	Total elongation	Product of ultimate tensile strength and total elongation, MPa%
TQ	0.56±0.03	398±22	608±39	0.13±0.008	0.20±0.018	12160
T 350	0.60±0.04	344±20	552±31	0.17±0.007	0.25±0.012	13800
T 400	0.55±0.06	356±12	590±35	0.17±0.029	0.27±0.049	15930
T 450	0.54±0.05	335±23	581±3	0.15±0.004	0.24±0.014	13944

The TQ sample showed a strain hardening exponent behaviour similar to ferrite-martensite DP steel [6, 22], which is a rapid increase to a maximum point of 0.193 and then a fast decrease (Fig. 19(c)). The T 350 sample also had a quick increase in strain hardening exponent to a maximum point of 0.190 but then showed a slow decrease. The strain hardening exponents of the T 400 and T 450 samples slowly increased to maximum points of

0.187 and 0.172, respectively. Then they decreased at an intermediate rate. Similar observations were reported in Refs. [31, 48]. The rate of strain hardening exponent variation in the T 400 sample was lower than that in the T 450 sample.

In addition, the strain hardening behaviour was analysed using a modified C – J model. The strain hardening behaviour was divided into three stages (Fig. 19(d)). The strain hardening index for every stage was included in Table 3. A smaller strain hardening index m means a higher strain hardening ability. m increased from stage 1 to stage 3 for the TQ and T 350 samples, while m decreased first from stage 1 to stage 2 and then increased from stage 2 to stage 3 for the T 400 and T 450 samples. The T 400 sample had the lowest experimental m , while TQ had the largest one.

Table 3 Comparison of the modified Crussard – Jaoul (C – J) analysis with the experimental results.

Specimen	Modified C-J model			Experimental m^b
	m_1^a	m_2	m_3	
TQ	4.9	7.5	25.6	9.1
T 350	5.4	5.6	11.4	6.6
T 400	7.3	5.0	34.6	6.4
T 450	6.4	5.6	32.9	7.4

^a1 – m_{1-3} = slopes at stage 1 to 3

^b $m = 1/(\epsilon_u - \epsilon_y)$, where ϵ_u is maximum uniform true strain and ϵ_y is true strain at 0.2% offset proof stress

4 Discussion

4.1 The dependence of microstructure on isothermal bainite transformation temperature

After holding at 350, 400 and 450 °C, the samples consisted of polygonal ferrite, bainitic ferrite, granular bainite, martensite, retained austenite and very little amount of Widmanstätten ferrite, characterised using XRD (Fig. 10), OM (Fig. 8), SEM (Figs. 11, 12 and 14), EBSD (Fig. 17), TEM (Figs. 13 and 15) and APT (Fig. 18). Except for the Widmanstätten ferrite, the microstructure was similar to a typical microstructure of hot rolled TRIP steels [3-5, 13, 14].

The fraction of RA in T 400 sample was the largest among all conditions, which was 0.045 ± 0.003 with the average carbon content of 1.23 ± 0.01 wt.% (Fig. 10(b)), although, nano-scale cementite was also observed in T 400 sample using APT (Fig. 18). It was reported that the nano-scale carbides precipitation could take place despite a high silicon concentration (1.5 wt. %) in TRIP steel [39, 49]. When the IBT temperature increased to 450 °C, the amount of RA decreased to 0.025 ± 0.002 . This could be explained by the increase in IBT temperature reducing the driving force for bainite formation, leading to a lower bainite fraction [13, 17, 50]. Therefore, the remaining austenite was insufficiently stabilised and transformed to martensite during quenching to room temperature. As can be seen from Fig. 10(b), the carbon content in RA decreased from 1.23 ± 0.01 to 0.92 ± 0.01 wt. % when the IBT temperature increased from 400 to 450 °C. This corresponds to a larger martensite fraction observed using colour etching in the T 450 sample, compared to that in the T 400 sample (c.f.

Figs. 8(f, h)). On further increasing in IBT temperature to 500 °C, pearlite formation took place (Fig. 16), which was detrimental for RA formation as this consumed carbon [38]. The consumption of carbon by carbide and pearlite reduced the carbon content in remaining austenite, which affected the chemical stability of RA leading to its transformation to martensite during quenching [7, 39]. On the other hand, when the IBT temperature was decreased from 400 to 350 °C, the RA fraction also decreased. As shown in Figs. 8 and 11, a small amount of bainitic ferrite or granular bainite (3 – 5 µm diameter regions) was formed in the T 350 sample, while in contrast such regions in the T 400 sample consisted of RA and/or martensite (Fig. 12(c)). Such observations in the phase variation could be explained by an increase in the driving force for bainite transformation with a decrease in IBT temperature [7, 13, 50], leading to a larger fraction of remaining austenite transforming to bainitic ferrite and granular bainite (Fig. 11). This resulted in a decrease in RA fraction with decreasing the IBT temperature from 400 to 350 °C. An increase in the amount of polygonal ferrite and bainite led to more carbon being rejected into RA. In addition, RA fraction in T 350 was smaller than that in T 400 sample. These two reasons resulted in increased carbon content in RA measured for the T 350 sample compared to that for the T 400 sample (Fig. 10).

In summary, the IBT temperature of 400 °C was the best choice for obtaining the largest RA fraction with medium carbon content. It consents with many researches [4, 6, 17, 51]. A higher IBT temperature would increase the tendency for carbide precipitation or pearlite formation. This would reduce the carbon content in the remaining austenite, leading to its transformation to martensite during subsequent cooling. A lower IBT temperature may result in more bainite formation from remaining austenite during holding leading to a reduced fraction of RA. Higher carbon content in the RA after lower temperature holding would increase the RA stability at room temperature. However, the TRIP effect may be less pronounced due to a reduced RA fraction and the overstabilisation of RA [8, 10, 15, 16].

4.2 The influence of coarse prior austenite structure on microstructure evolution and retained austenite retention

The challenge to produce TRIP steel using strip casting is the coarse prior austenite microstructure. The average PAGS of 80 ± 27 µm in this study before ferrite transformation was very large. As the deformation was not exerted during heat treatment such defects induced by deformation as deformation bands, dislocation tangles and sub-grain boundaries were not present [38, 52]; therefore, ferrite nucleated preferentially at prior austenite grain boundaries. Due to few nucleation sites, the polygonal ferrite grain size was large (17 ± 10 µm) as the impingement between them was postponed [42]. Such a size was very coarse compared to the typical polygonal ferrite grain size of ~ 10 µm obtained via conventional thermo-mechanical processing [14]. The consumption of prior austenite by polygonal ferrite growth from the boundary to the centre of prior austenite grains determined the size of second phase region. The PAGS ranged from 23 to 146 µm (Fig. 3(d)). When the prior austenite grain was large, the corresponding second phase region was also coarse. For example, the largest second phase region observed was 137 µm. Conversely, the size of second phase region was small when the prior austenite grain was small. In the T 400 sample,

the average size of second phase region was $21\pm 24\ \mu\text{m}$ (Fig. 9(b)), which held a large standard deviation due to the inhomogeneous prior austenite structure (Fig. 3(d)). Besides, a small amount of Widmanstätten ferrite observed (Fig. 8) in all conditions was also probably due to the coarse PAGS.

Carbon and manganese are the most important elements for the stability of austenite [17, 53]. Fig. 17(c) showed that carbon concentration near the polygonal ferrite/second phase region interface was higher than in the centre in the second phase region (namely remaining austenite at higher temperature), which lead to higher retained austenite presence. It can be explained by the carbon diffusing from polygonal ferrite to remaining austenite during ferrite growth [52]. A larger second phase region had a longer distance for carbon diffusion to the centre, leading to a larger difference in carbon content between the interface and the centre. A smaller second phase region had a larger area of polygonal ferrite/second phase region interfaces per volume than a larger second phase region, resulting in an increase in the rate of carbon and Mn diffusion during ferrite growth [4, 54]. As a result, the smaller second phase region was more chemically homogeneous with higher carbon content. Besides, very small areas inside the second phase region also showed high carbon content due to the carbon rejection into remaining austenite from bainite (Fig. 17(c)). Fig.17(d) showed a nearly homogeneous manganese distribution with some dark red dots, which were MnS inclusions according to EDS analysis. However, the manganese content should be higher in RA and bainite than that in polygonal ferrite, due to manganese being one of austenite forming elements [53]. This phenomenon was not observed here probably because of EDS accuracy and small difference in Mn content between phases.

Most of carbon enriched area corresponded to the RA, whereas, some carbon enriched areas were martensite (c.f. Figs. 17(b, c)). Besides, when the remaining austenite in polygonal ferrite was small, it also did not transform during quenching (Figs. 8(f) and 12(c)) despite having only benefit of carbon enrichment from polygonal ferrite, as other factors such as grain size, stress state and morphology also affect the stability of austenite [6]. In interior of large second phase regions, which may have a lower carbon content, the bainitic ferrite lath tended to coalesce with neighbours [47], leading to enlargement of bainitic ferrite lath (Fig. 17(a)). This would reduce the amount of RA between bainitic ferrite lath, as demonstrated by very tiny RA films observed using TEM (Fig. 13(b)) and EBSD (Fig. 17(b)). The same phenomenon was also observed in Ref. [55].

4.3 Correlation of mechanical properties with microstructure

As expected, the highest UTS and lowest TE were obtained after directly quenching from austenite-to-ferrite region (TQ sample, Table 2). Martensite formation resulted in a higher dislocation density in the TQ sample [17], compared to the IBT held samples containing predominantly bainite as the second phase. Therefore, in the TQ sample, the strain hardening exponent increased and then decreased faster, compared to the strain hardening behaviour of the IBT samples. As a result, the TE was low in the TQ sample.

For IBT held samples, RA played an important role in mechanical behaviours. Besides, other factors such as polygonal ferrite grain size and its fraction, and the interaction between polygonal ferrite and second phases also affected the mechanical behaviour [6, 9, 12, 16]. In this discussion, because of similar microstructures, only the effect of RA on mechanical behaviours is addressed. The strain hardening exponent of the T 350 sample increased very quickly at first and then decreased at a very slow rate (Fig. 19c); probably some of RA transformed to martensite during deformation leading to a larger TE than that of the TQ sample. As the carbon content in RA in the T 350 sample was very high (1.36 ± 0.01 wt. %) and most of RA were of film shape leading to high stability of RA, the austenite-to-martensite transformation during tensile test was very slow and some RA may not have transformed. The strain hardening exponent of the T 400 sample increased at a moderate rate and then slowly decreased with an increase in true strain. Probably, the RA continuously transformed to martensite during straining. The blocky RA could transform to martensite at lower strains, while the film RA between bainitic ferrite would transform at larger strains due to high stability [7, 18]. The RA transformation resulted in the dislocation generation around the martensite, due to the accommodation of volume change (Fig. 13(d)) [17, 56]. This process contributed to a continuous increase in the strain hardening exponent by inhibiting the glide of dislocations and then a slow decrease after the maximum point [6], leading to the highest observed TE of 0.27 (Table 2). The highest TE in T 400 sample resulted in the largest product of UTS and TE (Table 2) among all the samples. The T 450 sample showed the highest initial strain hardening exponent among all IBT held samples, which could be explained by the following. The carbon content in RA decreased from 1.23 ± 0.01 to 0.92 ± 0.01 wt.% with the IBT temperature increasing from 400 to 450 °C. This could lead to a decrease in RA stability, earlier transformation of RA and a higher initial strain hardening exponent [57, 58]. Because of an earlier RA transformation, the TRIP effect became less pronounced at higher strains, leading to a quicker decrease in the strain hardening exponent. As a result, a lower TE was obtained in the T 450 sample than that in the T 400 sample.

Based on the modified C – J model, the strain hardening behaviour could be divided into three stages (Fig. 19(d)). The TQ sample showed a typical strain hardening behaviour of ferrite-martensite DP steel [59]: stage 1 was related to work hardening of polygonal ferrite during deformation; stage 2 corresponded to co-deformation of polygonal ferrite and martensite, when the ferrite was work hardened enough; during stage 3 the deformation was restricted by the dislocation pile-up leading to dislocation immobilisation [22, 59]. Thus, the strain hardening index m increased from stage 1 to stage 3. The product of UTS and TE in TQ sample was lower than those in the T 350, T 400 and T 450 samples, indicating the mechanical properties enhancement via the IBT holding stage [56]. The strain hardening behaviour of the T 350, T 400 and T 450 samples also had three stages (Fig. 19(d)). At stage 1 the deformation of polygonal ferrite dominated, as the RA with about 1 wt. % carbon had a higher yield stress than the ferrite matrix [14]. At stage 2 the co-deformation of different phases took place. The T 350 sample showed the m increased from stage 1 to 3. However, in the T 400 or T 450 samples, the m_2 were smaller than m_1 , meaning a higher work hardening ability in stage 2. This could be explained by an earlier transformation of RA to martensite in the T 400 and T 450 samples than that in the T 350 sample. Deformation during stage 3 was

restricted by the hardened phases. In addition, the experimental m in the T 400 and T 350 samples was lower than that in the T 450 sample. A higher RA fraction and higher stability due to a larger carbon content in the T 400 sample resulted in a lower experimental m compared to the T 450 sample. On the other hand, a higher stability of RA in the T 350 sample was responsible for a lower experimental m compared to the T 450 sample, although they had similar RA fraction. The experimental m in the T 400 sample was lower than that in the T 350 sample because the T 400 sample had more RA.

4.4 The approaches to improve mechanical properties

The average PAGS of $80 \pm 27 \mu\text{m}$ observed in this work resulted in the polygonal ferrite grain size of $17 \pm 10 \mu\text{m}$, which was larger than $10 \mu\text{m}$ in hot rolled steels [14]. The coarser polygonal ferrite had a lower strength based on the Hall-Petch relationship [60]. In addition, the large second phase region here ($21 \pm 24 \mu\text{m}$) was detrimental for tensile properties [57, 61].

The other more important factors are the RA fraction and its stability; as was shown previously, to maximise the TRIP effect and enhance mechanical properties an increased RA fraction with medium stability would be required [13, 16, 18, 31]. The amount of RA observed here ranged from 0.023 to 0.045, which was much lower than the $\sim 0.05 - 0.2$ fraction usually present in this kind of conventional low alloyed TRIP steels [12, 13, 15]. The maximum achieved RA fraction of 0.045 here was ascribed to the coarse prior austenite structure as discussed above. For comparison, a TRIP steel with similar composition to the one studied here, which had around 0.1 RA, exhibited a higher UTS of $\sim 720 \text{ MPa}$ with a corresponding higher TE of ~ 0.26 [57].

In addition, a small amount of Widmanstätten ferrite observed here was also harmful to the mechanical properties. The formation of Widmanstätten ferrite was probably due to large PAGS.

Thus, the first step to improve mechanical properties in strip cast TRIP steel is to refine the microstructures. Addition of alloying elements, such as V, Ti and Nb, could refine PAGS and the final microstructure by grain growth retardation [17, 62, 63]. Deformation above the non-recrystallisation temperature can help to reduce the austenite grain size by dynamic/static recrystallisation; and deformation below the non-recrystallisation temperature may increase the amount of crystal defects (dislocations, deformation bands) acting as nucleation sites for ferrite formation, which would refine the low temperature microstructure [38, 64]. These effects of deformation will be studied in the future taking into account the limitations for strip casting technology with respect to the amount of deformation imparted.

5. Conclusions

In this study a possibility to obtain conventional TRIP steel via strip casting technology was investigated via the process simulation in the laboratory. The analysis of effect of isothermal

bainite transformation temperature on microstructure evolution and mechanical properties resulted in the following conclusions:

- (1) A processing route based on strip casting was developed to produce conventional TRIP steel with the microstructure consisting of ~ 0.55 polygonal ferrite, bainite, 0.02 – 0.045 retained austenite, and small amounts of martensite and Widmanstätten ferrite.
- (2) For TRIP steel with nominal composition of 0.17C-1.52Si-1.61Mn-0.03Al (wt. %), an experimental CCT diagram was obtained. The nose temperature of ferrite formation field was determined being around 650 °C. As expected, decreasing the cooling rate from austenitisation temperature to 670 °C resulted in an increase in ferrite fraction.
- (3) With an increase in isothermal bainite transformation temperature from 350 to 400 and to 450 °C, the retained austenite fraction increased from 0.025 to 0.045 and then decreased to 0.023. Meantime, the carbon content decreased from 1.36 to 0.92 wt. %. A large amount of retained austenite was located along the polygonal ferrite/second phase region interfaces. The retained austenite fraction decreased towards the centre of second phase region due to the carbon inhomogeneity arising from its rejection from the neighbouring polygonal ferrite.
- (4) The best combination of ultimate tensile strength (590 MPa) and total elongation (0.27) was achieved after holding at 400 °C. At the same time, the maximum retained austenite fraction of 0.045 was obtained with an intermediate carbon content of 1.23 wt. %.
- (5) The strain hardening behaviour was divided into three stages based on Crussard – Jaoul model. The TRIP effect predominately affected the second stage, resulting in a decrease in strain hardening index. The stronger TRIP effect led to a larger decrease in the strain hardening index.

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