Effects of the austenitizing temperature on the mechanical properties of cold-rolled medium-Mn steel system

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Abstract

Quenching and partitioning heat treatment (Q&P) was applied to a cold-rolled high strength steel system (0.28C – 1.42Si – 4.08Mn, wt%) . The quenching temperature was optimized through the constrained carbon equilibrium (CCE) criterion. The effects of austenitizing temperature on the mechanical properties of the experimental steels were systematically investigated. The microstructure was analyzed by means of scanning electron microscopy (SEM), transmission electron microscope (TEM), electron backscatter diffraction (EBSD) and X-ray diffraction (XRD). A significant change of mechanical properties was found when the austenitizing temperature varies from 760°C to 900°C with the different microstructure uniformity owing to the prior austenite. The yield strength changes from 1038MPa to 1276MPa and the tensile strength changes from 1399MPa to 1548MPa, while the elongation changes from 13.4% to 15.6%. By characterizing the retained austenite and analyzing the work hardening behavior, we found that the TRIP effect is heavily linked with the stability of retained austenite during the stretching process. The sustained TRIP effect and uniform microstructure contribute to the exceptional mechanical properties with both high strength and high ductility.

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1. Introduction

The quenching and partitioning (Q&P) process proposed by Speer et al. [1] is well-known as one of the most promising and innovative heat treatments for the production of the Third Generation advanced high strength steels with a good combination of strength and formability. It is featured by quenching the material from the austenitizing temperature to a temperature between M_s and M_f temperatures, where it is held to allow carbon to migrate from the oversaturated martensite to metastable austenite. As a result, the stability of austenite increases. The Q&P steel contains stabilized retained austenite between the laths of martensite, and therefore exhibits superior ductility. The constrained carbon equilibrium (CCE) model [1, 2] can be applied to predict an optimal quenching temperature, which will contribute to a maximum amount of retained austenite. However, because of the lack of the consideration of interface mobility [2-7], the predicted value through CCE model is a little higher than the measured one [8].

Most previous researches on the Q&P process are focused on the steels with a manganese content of less than 3(wt. %), while few studies [9-13] are focused on the effects of the austenitizing temperature rather than the quenching and partitioning parameters on the steels with higher manganese content. The present work aims to investigate the microstructure and mechanical property of a medium-Mn high strength C-Si-Mn steel system under Q&P process.

2. Experimental Procedures

The chemical composition of the investigated steel in this work was 0.28C – 1.42Si – 4.08Mn (wt.%). Carbon and manganese are important elements that can stabilize the retained austenite and improve the hardenability. Another main effect of Mn is to decrease the temperature of intercritical domain, thus enabling partitioning treatment without too strong grain coarsening. The suppression of cementite during the isothermal process was achieved by the addition of silicon. The steel was smelted in a vacuum induction furnace and cast into a mould. The 25kg ingot was forged into several rectangular ingots (30mm × 75mm × 100mm). The ingots were homogenized at 1200°C for 1 h, followed by hot rolling to sheets with a thickness of 4.5 mm. The coiling temperature was 650 °C. After pickling, the hot-rolled sheets were directly cold-rolled to about 1.5mm thickness with the cold-rolling reduction rate of about 66%. Rectangular

specimens of 70 mm width and 220 mm length were machined from the cold-rolled steel sheet parallel to the rolling direction prepared for the heat treatment.

The austenitizing temperature of the steel was investigated by specimens directly quenched from different temperatures. The specimens were heated with the heating rate of 2°C/s to the different temperatures (680°C, 720°C, 760°C, 780°C), and held for 5min, followed by subsequently gas quenching with the cooling rate of 50°C/s to room temperature by using the continuous annealing simulator (CCT-AY- II produced by Ulvac-Riko INC.). It is found that there is a decrease in the martensite start (M_s) temperature with a reduction of prior austenite grain size [14]. Considering the influence of prior austenite grain size on M_s Temperature, we investigated the different optimum quenching temperatures (calculated through CCE model). We machined specimens of $3 \times 1.5 \times 10 \text{ mm}^3$ for dilatometric tests. The specimens were also heated at 2°C/s to the specified temperature and held for 5min, and then cooled at 50°C/s to room temperature. The different austenitizing temperatures with the similar M_s Temperature were designed to avoid the influence of M_s Temperature. The corresponding M_s Temperatures were measured as 265°C, 256°C, and 264°C for the samples austenitizing at 760°C, 850°C and 900°C, respectively.

In order to determine an appropriate quenching temperature, CCE condition (**Eq. (1) - Eq. (4)**) [2] with the combination of an empirical relation (**Eq. (5)**) [15] and the Koistinen–Marburger relationship (**Eq. (6)**) [16] was utilized. Where T is absolute temperature calculated as 723.15K (partitioning at 450°C), X_C^{alloy} is the overall carbon content of the steel (mole fraction), f_i^γ is the mole fraction of retained austenite before partitioning begins, and $X_{C_i}^\gamma$ is the initial carbon concentration of the austenite. f_{CCE}^α and f_{CCE}^γ , and X_{CCE}^α and X_{CCE}^γ represent the ferrite and austenite amount and carbon concentration, respectively, at constrained carbon equilibrium when carbon partitioning is completed [2]. In the entire calculation process, there are a number of points need to be noted: (1) the α/γ interface is immobile, (2) competing reactions (such as carbide precipitation) are precluded [2], (3) the contents of elements Mn and Si do not change in both phases, and (4) the calculation of K-M equation is just a rough estimation for the end amount of retained austenite as it does not describe the kinetics in a reliable way for all steels and the parameter depends strongly on the carbon content in solution.

$$f_{CCE}^\gamma(1 - X_{CCE}^\gamma) = f_i^\gamma(1 - X_{C_i}^\gamma) \quad (1)$$

$$f_{CCE}^\alpha X_{CCE}^\alpha + f_{CCE}^\gamma X_{CCE}^\gamma = X_C^{alloy} \quad (2)$$

$$X_{CCE}^{\gamma} = X_{CCE}^{\alpha} \cdot \exp\left(\frac{76789 - 43.8T - (169105 - 120.4T)X_{CCE}^{\gamma}}{RT}\right) \quad (3)$$

$$f_{CCE}^{\alpha} + f_{CCE}^{\gamma} = 1 \quad (4)$$

$$M_s = 539 - 42300C - 3040Mn - 750Si \quad (5)$$

$$f_M^{T_q} = 1 - e^{-1.1 \times 10^{-2}(M_s - T_q)} \quad (6)$$

The final retained austenite fraction and the carbon content in solution in austenite under each quenching temperature is presented in **Fig. 1**. The highest volume fraction of retained austenite can reach 31.8% with the carbon concentration in austenite calculated as 0.89 (wt. %), when the quenching temperature is 159°C. Considering the accuracy of gas quenching under the continuous annealing simulator, the quenching and partitioning parameters was selected as partitioning at 450°C for 60s after quenching to 160°C. Sample A, B and C were heated to the different temperatures (760°C, 850°C and 900°C, respectively) and held for 5min, then gas quenched to 160°C for 6s and reheat to 450°C for 60s. The schematic diagram of Q&P process is presented in **Fig. 2**.

Fig. 1. Theoretical calculation of the carbon content and retained austenite fraction with the change of quenching temperature

Fig. 2. Schematic diagram of Q&P process

Tensile properties were estimated by the uniaxial tensile test with the standard specimen (according to the GB/T 228.1-2010 standard, Gauge length: 50 mm), of which the tensile direction is parallel to rolling direction. The tests were performed with a crosshead speed of 2.0 mm/ min at room temperature. The resulting microstructures before the tensile test were observed by Zeiss ULTRA 55-type field emission scanning electron microscopy (FE-SEM) after etching with 2% nital. Twin-jet polishing technique (5% perchlorate alcohol at -30°C, applied potential of 50 V) was implemented to prepare thin foil samples for the TEM observation (JEM 2100). Electron backscatter diffraction (EBSD) was performed on the samples both before and after the tensile test with 20 kV and a step size of 0.1μm in the electron microscope. EBSD data processed with Channel 5 software provided by Oxford HKL Technology was used to characterize the microstructures and provide orientation information of the austenite. X-ray diffraction (XRD) was used to measure the retained austenite fraction and estimate the carbon content in austenite of all samples. XRD was performed on the specimen 15 × 15 mm² before the tensile test and the specimen 15 × 12.5 mm² after the tensile test at room temperature using the

D/MAX-RB operating at 45 kV and 150 mA, the 2θ containing the three austenite peaks ($\{200\}$, $\{220\}$ and $\{311\}$) and two martensite peaks ($\{200\}$, $\{211\}$) scan in order to estimate the fraction of retained austenite with the step size of 0.02° using Cu-K α radiation. The austenite lattice parameter a_γ (10^{-1}nm) of samples was determined by the position of the maximum of each austenite reflection ignoring the influence of residual stresses, and the manganese content x_{Mn} (wt.%) was considered to be homogeneous and constant in the steel. Thus the average carbon concentration x_c (wt.%) of the austenite was approximately obtained by using the equation [17, 18]:

$$a_\gamma = 3.556 + 0.0453x_c + 0.00095x_{Mn} \quad (7)$$

The surface of the samples for EBSD and XRD tests were electrolytically polished with a negligible internal stress in a mixture reagent of 20% perchloric acid and 80% ethanol with the voltage of 15V for 20-30s.

3. Results and Discussion

3.1 Microstructural analysis

SEM micrographs of samples under different austenitizing conditions are shown in **Fig. 3**. The microstructure of the specimens after heat treatment consists of different proportions of ferrite and martensite. A lot of granular carbides exist in the ferrite matrix with the austenitizing temperature of 680°C . As the holding temperature is increased to 720°C , most carbides are dissolved. The relative proportion of ferrite continuously decreases as the holding temperature increases, and the corresponding values are 48%, 22% and less than 3% in turn. Only martensite is observed in the specimen with an austenitizing temperature of 780°C .

Fig. 3. SEM micrographs of samples under different austenitizing conditions, quenched from (a) 680°C ; (b) 720°C ; (c) 760°C and (d) 780°C

SEM microstructures of samples under Q&P process with different austenitizing temperatures are shown in **Fig. 4**. All steels consist mainly of tempered martensite, retained austenite between the martensite lath and the mixture of retained austenite and fresh martensite formed after partitioning. The carbides embedded in the martensite matrix are the unavoidable precipitations in this steel during the partitioning at 450°C for 60s, which also indicates that it is impossible to get the highest final retained austenite volume fraction as calculated under the CCE criterion.

The film-like retained austenite with a bright contrast in the SEM micrograph of [Fig. 4](#) is located between martensite laths.

Fig. 4. SEM micrographs of (a) Sample A, (b) Sample B and (c) Sample C before the tensile test, in which TM, RA, and M/A represent tempered martensite, retained austenite between the martensite lath and the M/A islands, respectively

However, there are also distinctive differences among the three samples. There is a ferrite grain with the diameter about $1\mu\text{m}$ in Sample A as shown in [Fig. 4\(a\)](#), which is also observed in the specimen directly quenched from 760°C ([Fig. 3c](#)). The dash line marks the part of the legible boundaries. These boundaries could be the prior austenite grain boundaries or the martensite packet boundaries, and the dash line also indicates that Sample C has the biggest prior austenite grain size, which can be further confirmed by the following EBSD observation. Furthermore, it can be found that the blocky structure with a bright contrast breaks the continuity of the boundary, and this phenomenon is clearly observed in [Fig. 4\(a and b\)](#) marked with the dash line circle), as there is a larger prior austenite grain size in Sample C that results in a lower grain boundary density. And this blocky structure can be further confirmed as the blocky retained austenite by the following EBSD observation.

As known from the literature, most of the boundaries created during the martensite transformation are either low angle boundaries ($<15^{\circ}$) or high angle boundaries in the range from 50 to 63° [[19](#)]. In this paper, the boundaries with angle varying from 22 to 48° are marked by the solid white line in order to extract the possible prior austenite grain boundaries. As part of the prior austenite grain boundaries may be covered by the martensite boundary newly formed or blurry due to interfacial migration during the partitioning, the outline of solid white line in [Fig. 5\(a, b and c\)](#) can be initially identified as the prior austenite grain boundaries. Further, since the individual austenite grains, presented in the microstructure, sharing common orientations are inherited from the parent austenite grains, the regions where retained austenite with the similar orientation gathered can also be considered as a prior austenite grain. Thus, the outline of solid white line can be determined as the prior austenite grain boundary, through the combination analysis of the degree of the retained austenite orientation dispersion. Obviously, higher prior austenite grain boundary density of sample can be obtained when quenched from lower austenitizing temperature.

Fig. 5. EBSD analysis results of the samples before the tensile test austenitizing at (a) 760°C, (b) 850°C, (c) 900°C, combined band-contrast map of martensite and the orientation map of retained austenite; The phase transformation degree of the samples after the tensile test austenitizing at (d) 760°C, (e) 850°C, (f) 900°C, combined band-contrast map of martensite and the blue-coded map of retained austenite; The inset shows the inverse pole-figure map for the austenite phase in (a), (b) and (c). (ND: normal direction, RD: rolling direction)

The morphology of the microstructure in the samples is further investigated. The martensite is mainly in fine laths, while the retained austenite is presented as both the film-like retained austenite (or the wider retained austenite lath) between martensite laths and the blocky phase along the prior austenite boundaries or martensite packet boundaries in [Fig. 5\(a, b and c\)](#). The phase map of the sample after the tensile test reveals that almost all of the retained austenite detected by EBSD with the step size of 0.1μm has been transformed into martensite, as shown in [Fig. 5\(d, e and f\)](#).

Related data, such as the grain size of martensite and retained austenite (defined by large-angle boundary 10°), volume fraction and average carbon content of retained austenite, provided by EBSD and XRD analyses is given in [Table 1](#). The corresponding XRD patterns of the investigated steels treated by Q-P process both before and after the tensile test are shown in [Fig. 6](#). The results of volume fractions of the retained austenite measured by XRD and EBSD exhibit similar regularities, while localized EBSD scans the significantly lower austenite. In addition, the spatial resolution of the EBSD technique is limited by the step size of 0.1μm. Smaller austenite cannot be resolved by EBSD but can be detected with XRD. From the XRD investigation, the retained austenite of Sample A has the highest volume fraction of 19.3% with an average carbon content of 0.83%. As the heating temperature increased to 850°C and 900°C, the volume fractions of retained austenite are decreased to 12.3% and 12.2%, while the average carbon contents is elevated to 0.97% and 0.89%, respectively.

Fig. 6. XRD patterns of the investigated steels treated by Q-P process both before and after the tensile test

Table 1. The information of martensite and retained austenite from EBSD and XRD

The retained austenite grain size is divided into four dimensions for analysis, the grain area of $0.04\text{-}0.1\mu\text{m}^2$, $0.1\text{-}0.5\mu\text{m}^2$, $0.5\text{-}1\mu\text{m}^2$ and above $1\mu\text{m}^2$, respectively. Here the grain area below $0.04\mu\text{m}^2$ is removed from the statistical data because of the possible inaccurate recognition under the EBSD system. The proportion and number of retained austenite with the corresponding grain area in EBSD data is given in Fig. 7. The multiple morphologies of retained austenite are also shown through transmission electron micrograph in Fig. 8. Accordingly the retained austenite of the four dimensions can represent the film-like retained austenite or (the retained austenite lath) between the martensite laths, small blocky retained austenite, big blocky retained austenite and the bigger M/A island, respectively, marked with 1, 2, 3, 4 in both Fig. 4 and Fig. 8.

Fig. 7. The EBSD statistics of the retained austenite divided into four dimensions with the corresponding grain areas, representing the multiple morphologies of retained austenite, in the samples before the tensile test

Fig. 8. The transmission electron micrographs of retained austenite in Sample B before the tensile test: (a) the bright-field image and (b) dark-field image with the inserted SAED pattern.

It is worth noted that there is a lot of blocky retained austenite along the prior austenite grain boundaries, marked with the dash line circle in Fig. 4 and Fig. 5. The sample with the highest prior austenite grain boundary density has the largest amount of blocky retained austenite. It could be due to the chemical heterogeneity of the austenite grains, which are heated at the low temperature for just a few minutes. The fact that thimbleful of ferrite observed in the specimen contributes to the enrichment of carbon and manganese in the ferrite/austenite interface of austenite [20, 21]. After quenching from the austenitizing temperature, the center of an austenite grain with the lower alloy components may firstly finish the martensite transformation, while a lot of blocky retained austenite along the prior austenite grain boundaries remains totally untransformed at the same time. This phenomenon becomes non-obvious with the increasing austenitizing temperature, resulting in an improved uniformity of austenite. Thus, compared with Sample A and B, less blocky retained austenite along the prior austenite grain boundaries can be found in Sample C, as shown by both SEM and EBSD in Fig. 4 and Fig. 5.

The presence of thimbleful of ferrite in Sample A suggests the chemical heterogeneity throughout the austenite grain and accounts for the largest amount of blocky retained austenite along the prior austenite. Mn heterogeneity should be inherited from the cold rolled microstructure (as shown in **Fig. 3**) in our medium-Mn steel, which is also accounts for the above phenomenon, and similar Mn heterogeneity has been confirmed by EPMA quantitative Mn mapping of different samples in the literature[22]. The influence of Mn heterogeneity on the final retained austenite fraction in Q&P steel may be complicated. The following mechanism could be proposed to analyze this situation. During the first quenching step of the Q&P treatment, the austenite in the Mn-lean regions partially transforms into martensite, while the austenite in the Mn-rich regions maintain the same state. Then in the following partitioning treatment, the partially untransformed austenite with the carbon sufficiently enriched can be retained, while the austenite in the Mn-rich regions could also be influenced by the surrounding microstructure. If both the enrichment of carbon from surrounding microstructure and manganese itself of the austenite in the Mn-rich regions is insufficient, the final retained austenite fraction of Sample A would be less than that in the sample under the higher austenitizing temperature with the improved uniformity of austenite. In fact, the EBSD data suggests the sufficient enrichment stability of austenite both in the Mn-lean regions and Mn-rich regions.

Moreover, smaller prior austenite grain contributes to higher grain boundary density, resulting in the martensite transformation limited in a small space, in which the formation of large martensite is prevented. As shown in the morphology observed through SEM and EBSD, Sample C has a significantly larger size of tempered martensite, especially in **Fig.4(c)**, while Sample A has the smallest average grain diameter of $0.78\mu\text{m}$ by the statistical data of martensite in EBSD.

3.2 Mechanical properties and work hardening behaviors

The steel austenitizing at 760°C exhibits good mechanical properties with the yield strength of 1038MPa, tensile strength of 1514MPa and elongation of 15.2%. When being austenitized at 850°C , it has similar mechanical properties, i.e., a yield strength of 1276MPa, a tensile strength of 1548MPa and an elongation of 15.6%, respectively. There is a slight difference when being austenitized at 900°C , i.e., a yield strength of 1216MPa, a lower tensile strength and elongation, 1399MPa and 13.4%, respectively. The stress-strain curve and the work hardening rate-strain curve are given in **Fig. 9**. There is an obvious serrated flow stress in the true stress-strain curve and a fluctuation in the work hardening rate-strain curve (**Fig. 9b**), indicating highly localized deformation and TRIP effect during tensile test in the steel. The three-stage hardening behavior is proposed to explain TRIP effect in ultrafine-grained austenite–ferrite Mn-TRIP steels [23]. In

stage I, the work-hardening rate is contributed mainly from the dislocation hardening in ferrite and austenite. As strain increases, the work-hardening contributions from the ferrite and the austenite decrease sharply after stage I and the hardening contributed from the martensite increases with increased strain. In stage II, the increasing of work hardening rate is due to a large amount of phase transformations. In stage III, while the phase transformation remains unchanged, the work hardening rate decreases significantly. In contrast to the ultrafine-grained austenite–ferrite Mn-TRIP steels, the microstructure of medium-Mn Q&P steel consists of the martensite with high dislocation density and the retained austenite with stabilities at different levels. As the stress or strain reaches a certain value during the stretching process, part of the retained austenite in the experimental steel transforms into martensite, resulting in the rising of work hardening rate. Then the stress or strain transfers to the other grains. And the work hardening rate won't rise again until the strain rising to a higher level for the transformation of retained austenite with relatively higher stability. This stage is similar to the stage III of TRIP effect in ultrafine-grained austenite–ferrite Mn-TRIP steels. The matrix microstructure of the martensite with high dislocation density contributes to the high strength of the steel. The difference of matrix microstructures will be also reflected in the different work hardening behaviors. The combination of the strong localized deformation and the TRIP effect finally results in obvious fluctuation feature in **Fig. 9**.

Fig. 9. (a) The engineering stress-strain curve, (b) the work hardening rate-strain combined with the true stress-strain curve (the data before necking)

The yield strength of 1038MPa in Sample A is notably lower than the other two (1276MPa and 1216MPa) as shown in **Fig. 9**. But the trend of the amount of retained austenite is opposite, combining with the results shown in **Table 1**. The thimbleful of ferrite observed in the specimen quenched from 760°C may contribute to the lower yield strength of Sample A. Besides, similar results have been mentioned in studies on the low-temperature-tempered martensite and Q&P steel [24-26], it seems that the larger amount of retained austenite, the lower the elastic limit and yield strength. It should be attributed to the particular face-centered cubic lattice of austenite, which contains four close-packed planes. Each plane contains three close packed directions, and dislocation motion is easy in a face-centered cubic structure. In fact, the legible TRIP effect at the initial stages with the strain about 0.01 shown in the work hardening rate-strain curve in Sample A also supports the motion of dislocations in the retained austenite resulting from the martensitic transformation.

Compared with Sample A shown in the [Fig. 9\(b\)](#), similar phenomena only occur at the higher strain level above 0.2 in the other two samples. We can summarize the following information, which is obtained from [Fig. 9](#).

(1) The work hardening rate of Sample A is the highest among the three samples, and has remained at a high level throughout the whole tensile process. This finally contributes to the high tensile strength, although the yield strength of Sample A is relatively low. The true stress of Sample A exceeds that of Sample C at the strain about 0.05, and remains almost consistent with the true stress of Sample B at the strain about 0.11.

(2) Although there are similar yield strengths of both Sample B and Sample C, the work-hardening rate of Sample B is always higher than that of Sample C throughout the whole tensile process. Necking occurred in Sample C with the true strain about 0.11 results in both lower tensile strength and uniform elongation. In contrast, the continuous work hardening rate in Sample B results in high tensile strength and uniform elongation.

The observed differences in work hardening behaviors can be understood in terms of different stabilities of retained austenite in the three samples. Previous studies [\[27, 28\]](#) showed that the stability of austenite depends on not only the chemical enrichment but also the size of austenite. Under the same condition, in general, both small grain size and high carbon content of the retained austenite contribute to the higher the stability of the retained austenite. Besides, the morphology and distribution of retained austenite also play an important role on the stability of retained austenite [\[21, 22, 29, 30\]](#). The stability of retained austenite is often influenced by the surrounding matrix or the strain distribution during deformation. For example, the mechanical stability of film-like retained austenite between the martensite laths must be higher than the stability of retained austenite islands surrounded by ferrite even if it has the higher carbon content.

As shown in the EBSD and XRD results before and after the tensile test, a large amount of retained austenite has almost entirely transformed into martensite during the whole tensile process in all three samples. The previously experimental data shown in [Fig. 5](#) and [Table 1](#) has proven the presence of multiple morphologies and dimensions of retained austenite in all three samples, which results in the continuous TRIP effect shown in [Fig. 9\(b\)](#). That is to say, the retained austenite with different stabilities in all three samples contributes to the TRIP effect at the different strain stages until the necking or damage occurs. It is also supported by the fluctuation presenting throughout the work hardening rate-strain curve in [Fig. 9\(b\)](#).

However, there is a difference concerning the degree of TRIP effect in all three samples. Compared with Sample C, the TRIP effect of Sample A and Sample B is significantly stronger. Undoubtedly, the difference of the carbon content between Sample B (0.97%) and Sample C (0.89%) could account for the different TRIP effects between Sample B and Sample C, as there is no significant difference in regard to the morphology and distribution of retained austenite. The M_s calculated by the empirical formula reduces by 33°C with the carbon content decreasing by 0.08%. The data in [Table 1](#) shows that partial retained austenite with a higher average carbon content of 0.89% still present in the Sample A after the tensile test. That is to say, although the average carbon content of Sample A is significantly lower than that of Sample C, there must be more retained austenite with higher carbon content in Sample A. This results in the stronger TRIP effect owing to the large volume fraction of retained austenite i.e., 19.3%. Interestingly, both high strength and high ductility are achieved in Sample A and Sample B. It could be concluded that the carbon content in the retained austenite plays a more important role on the stability of retained austenite than the retained austenite volume fraction.

In addition to the stability of retained austenite, the uniformity of the microstructure may also have an impact on the final mechanical property of the steel. The morphology and distribution of the microstructure present in these materials shown in [Fig. 4](#) are totally different. There is a microstructure with poor uniformity in Sample C, i.e. a large tempered martensite along the M/A island with the similar size, as shown in [Fig. 4\(c\)](#). In general, the homogeneous enrichment of carbon element in the big blocky austenite is more difficult than that in the small blocky austenite and film-like austenite, so that the fresh martensite forms after the partitioning. Then the presence of less ductile fresh martensite has a negative effect on austenite transformation stability owing to its constraining influence on the strain distribution [\[31\]](#), as well as the presence of large M/A islands. In contrast, smaller prior austenite contributes to more uniform microstructure, which can be found by comparing Sample A and Sample B.

Finally, it should be noted that an obvious necking occurs in tensile specimen of Sample C with lower tensile strength, rather than the instantaneous fracture in the Sample A and B. It can be proposed that there are a large amount of dislocations, which can be regarded as the micro voids existing in the microstructure of Sample A and B under the higher stress state after the sustained TRIP effect, compared to the Sample C. Once the micro crack occurs, the micro crack will coalesce with the large amounts of dislocations or micro voids due to the enormous stress concentration, followed by the rapid crack propagation and the occurrence of macroscopic

fracture. This phenomenon deserves the further investigation, especially in the context of future application of high strength steel.

Conclusions

The austenitizing temperature plays a critical role on the microstructural evolution of cold-rolled medium-Mn steel under quenching and partitioning treatment. The fine prior austenite grain under the low austenitizing temperature contributes to a large volume fraction of retained austenite, especially the blocky retained austenite along the prior grain boundaries, which is due to chemical heterogeneity in the austenite and the high grain boundary density. In addition, small prior austenite grain with high grain boundary density effectively improves the uniformity of microstructure. The formation of large size martensite or M/A islands will be avoided.

The mechanical properties are dictated by both the sustained TRIP effect and the uniformity of microstructure. The sustained TRIP effect reflected in the work hardening behavior is associated with the amount of retained austenite and the retained austenite stability. Compared with the retained austenite volume fraction, the stability of retained austenite could be more important in providing the sustained TRIP effect. A thimbleful of ferrite observed in the specimen quenched from 760°C and high volume fraction of retained austenite in steel is favorable for obtaining relatively low yield strength, i.e., 1038MPa. The small size prior austenite results in a uniform microstructure and the sustained TRIP effect throughout the whole tensile process, which contributes to the favorable mechanical properties with both high strength and high ductility.

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Table 1

Table 1. The information of martensite and retained austenite from EBSD and XRD

T/°C	Area of RA grain before tensile test/μm ²								Equivalent grain size of martensite /μm	The RA fraction and average carbon content				
	0.04~0.1		0.1~0.5		0.5~1		>1			Before tensile test			After tensile test	
	The number of RA and the corresponding proportion									f _γ -EBSD	f _γ -XRD	C _{wt} -XRD	f _γ -XRD	C _{wt} -XRD
760	153	27.8%	276	50.2%	66	12.0%	55	10.0%	0.78	17.2%	19.3%	0.83%	3.2%	0.89%
850	97	30.8%	163	51.7%	28	8.9%	27	8.6%	0.99	11.2%	12.3%	0.97%	3.5%	0.88%
900	94	30.6%	148	48.2%	27	8.8%	38	12.4%	0.89	11.6%	12.2%	0.89%	3.8%	0.88%

Fig. 1

Fig.2

Fig.3(a)

Fig.3(b)

Fig.3(c)

Fig.3(d)

Fig.4(a)

Fig.4(b)

Fig.4(c)

Fig. 5(a)

Fig. 5(b)

Fig. 5(c)

Fig. 5(d)

Fig. 5(e)

Fig. 5(f)

Fig. 5 inset

Fig. 6

Fig. 7

Fig. 8(a)

Fig. 8(b)

Fig. 9(a)

Fig. 9(b)