Dynamic tensile behavior of novel quenching-partitioning-tempering martensitic steel

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A Fe–0.20C–1.49Mn1.52Si–0.58Cr–0.05Nb low-carbon steel was treated by a novel quenching-partitioning-tempering (Q-P-T) process and the traditional quenching and tempering (Q&T) process for comparison, respectively. X-ray diffraction results indicate that there is 10.8% volume fraction of retained austenite (VrA) in the Q-P-T martensitic steel, while no diffraction peak of retained austenite was detected in Q&T martensitic steel. The Q-P-T steel and the Q&T steel were subjected to a dynamic tensile test at a strain rate of 500 s⁻¹ and a quasi-static tensile test at a strain rate of 5.6 × 10⁻⁴ s⁻¹, respectively. The results indicate that the high strain rate in the dynamic tension raises the strength of the Q-P-T and Q&T steels compared with their quasi-static tensions owing to strain rate hardening effect, however, the elongation of Q-P-T steel decreases slightly, while the elongation of Q&T steel evidently increases. This difference is attributed that the existence of the dislocation absorption by the retained austenite (DARA) effect in quasi-static tension is suppressed under dynamic tensile loading so that the enhancement influence of DARA effect on ductility cannot be effectively compensated by the adiabatic softening of the martensitic matrix, which leads to a slight decrease in the elongation of the Q-P-T steel in dynamic tension compared with that in the quasi-static tension. While Q&T steel has no DARA effect in quasi-static tension due to no retained austenite, and thus the adiabatic softening of the martensitic matrix increases the ductility of Q&T steel in dynamic tension.

Introduction

Among the advanced high strength steels (AHSSs), quenching and partitioning (Q&P) steels [1,2] are the first AHSSs with martensitic matrix, exhibiting higher strength than dual phase steels, transformation induced plasticity (TRIP) steels, and twinning induced plasticity (TWIP) steels, moreover, Q&P steels keep enough ductility. The Q&P process proposed by Speer et al. [1] is that a Si-containing steel is quenched from the austenitizing temperature to a temperature (Tq) below the finish temperature of martensitic transformation, then followed by a ‘partitioning’ treatment either at or above Tq. During this ‘partitioning’, the carbon partitions (diffuses) from the supersaturated martensite phase to the untransformed austenite phase, which leads to the stability of the carbon-enriched retained austenite during subsequent cooling to room temperature. In the case of low or medium carbon Q&P steels, the Tq is usually much higher than room temperature [3]; thus, considerable amounts of the retained austenite can be obtained. Based on the ‘constrained carbon paraequilibrium’ (CCE) theory as the basis of the Q&P process proposed by Speer et al. [1], the formation of carbides is not permitted during the Q&P process, thus precluding precipitation strengthening. For this reason, Hsu [4] proposed the quenching-partitioning-tempering (Q-P-T) process in 2007; thus, additional carbide forming elements such as Nb and Mo are added to Q-P-T steels. The addition of these elements leads to the formation of stable carbides and grain refinement for precipitation strengthening and refined-grain strengthening [5]. Because the Q-P-T process encompasses the core idea of the Q&P process, the quenching temperature (Tq) is determined by a combination of the CCE theory and the K-M equation [6]. Q-P-T martensitic steels also have considerable amounts of retained austenite compared to traditional quenching and tempering (Q&T) steels. The effects of the retained austenite on enhancing the ductility in high-strength steels can be summarized in terms of three aspects: the TRIP effect proposed by Webster in 1968 [8], and the dislocation absorption by retained austenite (DARA)
effect that we proposed [9] in 2011 based on the measurement of the average dislocation densities in martensite and retained austenite by X-ray diffraction line profile analysis (XLPA) [10,11]; in the study [9], we demonstrated that the dislocations in the martensitic matrix can move into the nearby retained austenite through the martensite-retained austenite interfaces, and these dislocations are absorbed by the retained austenite.

The tensile properties of structural components are usually measured under conditions employing quasi-static (10^{-3}–10^{-1} s^{-1}) loading rates. However, the structural components of automobiles are also subjected to dynamic loading rates (10^{3}–10^{5} s^{-1}), such as those encountered in a car collision or in sheet metal formation. Hence, it is imperative to understand the behavior of these materials under these loading rates. Dynamic loading differs from quasi-static loading because the features of dynamic loading include both the strain rate hardening effect and the adiabatic softening effect, as revealed in studies of DP steels and TRIP steels [12–16]. Wang and Speer [17,18] reported the dynamic mechanical properties of Q&P 980 steel (consisting of ferrite, martensite, and retained austenite) in low-alloyed Fe-Mn-Si based AHSSs, but they did not investigate the influence of the strain rates or retained austenite on the dynamic mechanical behavior.

Our interest is in the effect of the retained austenite on the dynamic mechanical behavior of AHSSs, and thus we investigate the following two issues [19]: (1) the effect of a high strain rate on the DARA effect, and (2) the effect of the retained austenite on the ductility of Q-P-T martensitic steel under dynamic tensile loading, compared with that under quasi-static tensile loading.

**Experimental procedure**

The composition of the steel determined by chemical analysis was Fe–0.20C–1.49Mn–1.52Si–0.58Cr–0.05Nb (wt%). The steel was melted in a medium-frequency induction furnace and was then hot-rolled into a 20 mm thick plate. The samples (4 mm thick) were cut from the hot-rolled plate. After cutting, the respective samples were subjected to the Q-P-T process and the Q&T process for comparison. The samples treated by the Q-P-T process were austenitized at 950°C for 600 s, followed by quenching in a salt bath at 320°C (T_q) for 30 s. The samples were subsequently partitioned/tempered at 450°C for 60 s in molten salt, and finally rapidly quenched in water. The only difference between the Q-P-T process and the Q&T process is the quenching temperature (T_q). For the Q-P-T process, the T_q is 320°C, whereas for the Q&T process, the T_q is room temperature.

The quasi-static mechanical properties of the samples were measured in triplicate with a Zwick/Roell Z100 tensile testing machine at a strain rate of 5.6×10^{-4} s^{-1} at room temperature. Rectangular tensile samples were prepared with a gauge length of 15 mm, width of 5 mm, and thickness of 1.5 mm.

Split Hopkinson bar or Kolsky bar experiments are well established for characterizing the dynamic mechanical behavior of materials [20]. The dynamic mechanical experiments were conducted using a modified split Hopkinson tensile bar (SHTB) apparatus [21,22]. This consists of a loading device, bar components, and a data acquisition and recording system. A relatively short specimen is sandwiched between the incident bar and the transmitter bar, as schematically shown in Figure 1. Based on one-dimensional elastic wave propagation theory and on the records of the incident, the reflected, and the transmitted pulses (which are equivalent to the strains ε_t, ε_r, and ε_i, respectively), the stress σ(t), the strain ε(t), and the strain rate, ˙ε(t) within the sample in the dynamic tensile tests can be derived as follows [23]:

\[
\begin{align*}
\sigma(t) &= E \cdot \left( \frac{A}{A_0} \right) \cdot \epsilon_i(t), \\
\epsilon(t) &= \frac{2C}{L_0} \int \left[ \epsilon_i(\tau) - \epsilon_r(\tau) \right] d\tau, \\
\dot{\epsilon}(t) &= \frac{2C}{L_0} \left[ \epsilon_i(\tau) - \epsilon_r(\tau) \right],
\end{align*}
\]

where \( E \) is Young’s modulus of elasticity for the SHTB materials, \( C \) is the modulus of elasticity for the SHTB material, \( \epsilon_i(t) \) and \( \dot{\epsilon}(t) \) are the transmitted and incident strains, respectively, and \( \tau \) is the time delay between the incident and transmitted wavefronts.
the bars and the specimen, respectively. $L_0$ is the gauge length of the specimen, $\varepsilon_i$ is the incident pulse, and $\varepsilon_t$ is the transmitted pulse. Rectangular tensile samples were prepared with a gauge length of 10 mm, width of 4 mm, and thickness of 1 mm. Dynamic tensile tests were conducted at room temperature at a strain rate of 500 s$^{-1}$. At this strain rate, the dynamic tensile test data were highly reproducible for triplicate experiments.

A D/max-2550VL/PC X-ray diffractometer (with CuKα radiation) was used to analyze the austenite (200)γ, (220)γ, and (311)γ peaks and the martensite (200)α and (211)α peaks at room temperature. Based on direct comparison of the integrated intensity of the austenite (200)γ, (220)γ, and (311)γ peaks and the martensite (200)α and (211)α peaks at room temperature, the volume fraction ($V_{RA}$) of the retained austenite in the Q-P-T specimens before and after the tensile tests were determined. The average dislocation densities in both the martensite and retained austenite were measured by XLPA. The average dislocation density ($\bar{\rho}_M$ for the martensite, $\bar{\rho}_A$ for the austenite) was evaluated using XLPA, as described in detail in the literature [10,11]. Two samples unloaded at different strain stages were used for the XLPA measurements of the dislocation density, and the average values were obtained for both the Q-P-T and the Q&T steels. Microstructural characterization was carried out with a JEOL-2100F transmission electron microscope (TEM) operated at 200 kV. The TEM specimens were prepared by mechanical polishing, followed by electro-polishing in a twin-jet polisher using 5 vol.% perchloric acid and 95 vol.% ethanol at 20°C with an applied potential of 70 V.

Results and discussion

Dynamic and quasi-static tensile behavior

Figure 2 shows the engineering stress vs. engineering strain curves acquired under quasi-static tension (QST) with a strain rate of $5.6 \times 10^{-4}$ s$^{-1}$ and under dynamic tension (DT) with a strain rate of 500 s$^{-1}$, respectively. The results indicate that the yield strength (YS) and ultimate tensile strength (UTS) of the Q-P-T steel were 1043 and 1275 MPa, respectively, under quasi-static tension, where these values are lower than the yield strength (YS = 1251 MPa) and ultimate tensile strength (UTS = 1414 MPa) of QT steel. However, the total elongation (TE) and the product of the strength and elongation (PSE) of the Q-P-T steel were 16.8% and 21,420 MPa%, respectively, which are much higher than the TE (10.4%) and PSE (14,706 MPa%) of QT steel. The YS and UTS of the Q-P-T steel from the dynamic tensile test were larger than those of the Q-P-T steel from the quasi-static tensile test and increased by 289 and 146 MPa, respectively, whereas the elongation decreased slightly; thus, the PSE of Q-P-T steel increased by 2169 MPa%. Unlike the case for Q-P-T steel, both the strength and elongation of the QT steel under dynamic tensile testing conditions were larger than those from the quasi-static tensile tests. The mechanical properties of the Q-P-T steel and the QT steel under quasi-static and dynamic tensions are summarized in Table 1.

Variation of retained austenite fraction with strain

Figure 3 shows the XRD spectra of the Q-P-T steels with the application of various tensile strains. The patterns indicate that the steel consists of bcc-martensite and fcc-retained austenite. The $V_{RA}$ of the Q-P-T steel before the tensile test was determined to be 10.8%. Figure 3 also shows the XRD spectra of the Q-P-T tensile samples unloaded at different strain stages from 1% to 15%. The values of $V_{RA}$ at the different strain stages were calculated from these XRD spectra and are listed in Table 2. Table 2 shows that the $V_{RA}$ decreased with increasing strain and became undetectable at 15% strain. This finding indicates that the martensitic transformation occurred in the Q-P-T steel, accompanied by the TRIP effect. No diffraction peak of the retained austenite was observed in the XRD pattern of the QT steel, as shown in Figure 3.

Variation of average dislocation density with strain

To understand the mechanical behavior of the martensite and the retained austenite during the deformation, XLPA was used to measure the average dislocation density before and after the quasi-static tensile tests at different strain stages. As shown in Table 2, duplicate measurements were performed at each strain and the average dislocation density ($\bar{\rho}_M$ or $\bar{\rho}_A$) was obtained. It is clear from Table 2 that the $\bar{\rho}_A$ in the retained austenite increased rapidly with increasing strain for the Q-P-T samples. However, for the martensite, the $\bar{\rho}_M$ did not follow a similar trend. For example, the $\bar{\rho}_M$ was...
Table 1. The mechanical properties of the quasi-static and dynamic tensile tests samples treated by the Q-P-T and the Q&T processes, respectively (YS is at 0.2% offset strain).

| Condition | Process | YS (MPa) | UTS (MPa) | TE (%) | PSE (MPa%) |
|-----------|---------|----------|-----------|--------|------------|
| QST       | Q-P-T   | 1043 ± 6 | 1275 ± 11 | 16.8 ± 0.09 | 21420 ± 71 |
|           | Q&T     | 1251 ± 9 | 1414 ± 13 | 10.4 ± 0.08 | 14706 ± 23 |
| DT        | Q-P-T   | 1332 ± 10| 1421 ± 14 | 16.6 ± 0.10 | 23589 ± 92 |
|           | Q&T     | 1549 ± 15| 1666 ± 19 | 12.3 ± 0.12 | 20492 ± 34 |

Figure 3. The XRD spectra of Q-P-T quasi-static tensile (QST) samples at different strain stages (from 0% to 15% strains) and dynamic tensile (DT) sample as well as Q&T sample before tension.

The microstructure parameters of the martensite and the retained austenite in the quasi-static tensile (QST) test samples at different strain stages (from 0% to 15% strains) were characterized by TEM. The TEM characterization further indicates that the bcc phase apparent from the XRD spectra consists of dislocation-type martensite laths, as shown in the bright-field (BF) image in Figure 5(a). Moreover, the fcc phase indicated in the XRD spectra comprises flake-like retained austenite between martensite laths, as shown in the dark field (DF) image in Figure 5(b). The selected area electron diffraction (SAED) patterns presented as insets in Figure 5(b) and (d) show the $\{110\}_{//}\{111\}_{//}$ orientation relationship between the martensite and the retained austenite. The microstructure of the Q&T specimen consists of dislocation-type martensite and film-like retained austenite (Figure 5(c)), where the retained austenite is identified by the SAED pattern presented as an inset in Figure 5(d), while the retained austenite could not be detected in the XRD spectrum (Figure 3). This indicates that the volume fraction of the retained austenite in the Q&T sample was less than 3%, which is the lowest limit of the XRD measurement in this work. The microstructures of the Q-P-T specimen and the Q&T specimen after applying dynamic tension were markedly different from those after applying quasi-static tension. The features of the sample subjected to dynamic tension are as follows:

| Process (Condition) | Strain (%) | $\epsilon_a(1/2)$ ($\times10^{-3}$) | $\rho_{AI}$ ($\times10^{14}$ m$^{-2}$) | $\rho_M$ ($\times10^{14}$ m$^{-2}$) | $\epsilon_a(1/2)$ ($\times10^{-3}$) | $\rho_{AI}$ ($\times10^{14}$ m$^{-2}$) | $\rho_M$ ($\times10^{14}$ m$^{-2}$) | $\rho_A$ (MPa) | $\rho_{VRA}$ (%) |
|---------------------|------------|---------------------------------|---------------------------------|-----------------|---------------------------------|---------------------------------|-----------------|---------------|---------------|
| Q-P-T (QST)         | 0          | 2.36 ± 0.05                     | 6.61 ± 0.25                     | 6.57 ± 0.21     | 2.26 ± 0.31                     | 6.92 ± 0.73                     | 11.62 ± 0.83     | 10.8          |
|                     | 1          | 2.27 ± 0.04                     | 6.18 ± 0.24                     | 6.12 ± 0.20     | 2.15 ± 0.32                     | 6.15 ± 0.20                     | 14.32 ± 0.93     | 14706 ± 23    |
|                     | 2          | 2.09 ± 0.03                     | 5.05 ± 0.21                     | 4.93 ± 0.17     | 2.15 ± 0.39                     | 5.85 ± 0.27                     | 16.83 ± 1.32     | 19.13 ± 1.39  |
|                     | 3          | 2.15 ± 0.05                     | 5.60 ± 0.22                     | 5.46 ± 0.18     | 3.43 ± 0.45                     | 7.89 ± 0.43                     | 33.51 ± 3.33     | 35.62 ± 3.36  |
|                     | 7          | 2.19 ± 0.06                     | 5.85 ± 0.23                     | 5.59 ± 0.21     | 3.75 ± 0.46                     | 42.96 ± 4.51                     | 41.36 ± 4.37     | 42.16 ± 4.39  |
|                     | 11         | 2.28 ± 0.07                     | 6.46 ± 0.27                     | 5.96 ± 0.23     | 6.21 ± 0.25                     | 41.36 ± 4.37                     | 41.36 ± 4.37     | 41.36 ± 4.37  |
|                     | 15         | 2.31 ± 0.08                     | 6.53 ± 0.28                     | 6.19 ± 0.26     | 6.36 ± 0.27                     | 41.36 ± 4.37                     | 41.36 ± 4.37     | 41.36 ± 4.37  |
|                     | 16.8       | 2.33 ± 0.09                     | 6.65 ± 0.33                     | 6.39 ± 0.29     | 6.52 ± 0.31                     | 41.36 ± 4.37                     | 41.36 ± 4.37     | 41.36 ± 4.37  |
| (DT)                | 16.6       | 2.11 ± 0.03                     | 3.40 ± 0.25                     | 2.96 ± 0.21     | 3.18 ± 0.23                     | 41.36 ± 4.37                     | 41.36 ± 4.37     | 41.36 ± 4.37  |
| Q&T (QST)           | 0          | 2.39 ± 0.03                     | 6.80 ± 0.33                     | 6.76 ± 0.29     | 6.78 ± 0.31                     | 6.80 ± 0.33                     | 6.80 ± 0.33     | 6.80 ± 0.33   |
|                     | 1          | 2.41 ± 0.04                     | 6.99 ± 0.33                     | 6.79 ± 0.31     | 6.89 ± 0.32                     | 6.80 ± 0.33                     | 6.80 ± 0.33     | 6.80 ± 0.33   |
|                     | 3          | 2.43 ± 0.06                     | 7.21 ± 0.29                     | 6.85 ± 0.25     | 7.03 ± 0.27                     | 6.80 ± 0.33                     | 6.80 ± 0.33     | 6.80 ± 0.33   |
|                     | 5          | 2.46 ± 0.08                     | 7.34 ± 0.33                     | 6.98 ± 0.27     | 7.16 ± 0.30                     | 6.80 ± 0.33                     | 6.80 ± 0.33     | 6.80 ± 0.33   |
|                     | 7          | 2.52 ± 0.09                     | 7.45 ± 0.35                     | 7.19 ± 0.31     | 7.32 ± 0.33                     | 6.80 ± 0.33                     | 6.80 ± 0.33     | 6.80 ± 0.33   |
|                     | 9          | 2.53 ± 0.11                     | 7.82 ± 0.42                     | 7.56 ± 0.33     | 7.69 ± 0.36                     | 6.80 ± 0.33                     | 6.80 ± 0.33     | 6.80 ± 0.33   |
|                     | 10.4       | 2.56 ± 0.12                     | 7.89 ± 0.43                     | 7.81 ± 0.39     | 7.85 ± 0.41                     | 6.80 ± 0.33                     | 6.80 ± 0.33     | 6.80 ± 0.33   |
|                     | 12.3       | 2.19 ± 0.04                     | 3.17 ± 0.23                     | 2.89 ± 0.19     | 3.03 ± 0.21                     | 6.80 ± 0.33                     | 6.80 ± 0.33     | 6.80 ± 0.33   |
(1) most of the long and straight martensite laths were bent, as shown Figure 6; (2) there were several almost parallel microshear bands in the martensite lath (Figure 6), and each of these microshear bands was formed by pile-up dislocations on a certain slip plane [24,25]; therefore, the martensite areas on the two sides of each microshear band occurred as sub-grains rather than twin-type martensite, as they belong to one zone of the diffraction pattern [26]; and (3) compared with the martensite laths formed under quasi-static tensile conditions, under dynamic tensile conditions, there were low-density dislocations in some of the martensite laths, as shown Figure 7. Many individual dislocations without tangles could be clearly seen. This indicates the occurrence of dynamic recovery in these local areas due to the adiabatic heating in the dynamic tensile test [27–29].

As mentioned above, under quasi-static tensile conditions, the $\rho_M$ in the martensite decreased with increasing strain during the initial deformation, whereas after 3% strain, the $\rho_M$ gradually increased with increasing strain. This phenomenon can be explained as follows. Before 3% strain, the $\rho_M$ of the martensite gradually decreased with increasing strain, indicating that the number of dislocations transported to the nearby retained austenite exceeds the multiplication of the dislocations in the martensite, whereas after 3% strain, the $\rho_M$ of the martensite increased with increasing strain, indicating that the number of dislocations transported to the nearby retained austenite is less than that by dislocation multiplication in the martensite. This is termed the DARA effect. However, for the Q&T sample, the average dislocation density in the martensite increased monotonously with increasing strain, indicating that there was no DARA effect in the Q&T sample due to the very small amount of retained austenite. The TEM data reveal the occurrence of dynamic recovery caused by adiabatic heating in a local area with the concomitant formation of low-density dislocations in some
martensite laths (Figure 7) during the dynamic tensile test. The XLPA results show that the average dislocation densities in the martensite in the Q-P-T and Q&T samples fractured after the dynamic tensile tests were $3.18 \times 10^{14}$ m$^{-2}$ and $3.03 \times 10^{14}$ m$^{-2}$, respectively. The average dislocation density in the martensite before the tensile test was $6.59 \times 10^{14}$ m$^{-2}$ for the Q-P-T sample and $6.78 \times 10^{14}$ m$^{-2}$ for the Q&T sample. This indicates that the adiabatic heating during the dynamic tensile test leads to almost the same drop in the value of the average dislocation density in the martensite for the Q-P-T sample as for the Q&T sample, i.e. almost the same adiabatic softening of the martensite occurs. If the DARA effect occurs in the dynamic tensile test, the drop in the average dislocation density in the martensite for the Q-P-T sample should be much lower than that in the Q&T sample. This conflicts with the experimental results. Therefore, it is reasonable to believe that the DARA effect is negligible in the Q-P-T sample during the dynamic tensile test. That is, the high strain rate in the dynamic tensile test almost suppresses the DARA effect. This may be explained by the molecular mechanics simulation as follows. Under the applied load, the bcc-fcc interface can absorb $a/2 < 111>$ dislocations from the bcc phase, with concomitant generation of interface stress. Hence, the interface acts as a major barrier to dislocation transmission and allows the $a/6 < 112>$ partial dislocations to nucleate in the fcc-phase under the interface stress [30]. The complexity of this process requires sufficient time for the dislocations to be transmitted through the bcc-fcc interface. Because the dynamic tensile test at a high strain rate is completed in a very short period, there is not sufficient time for transmission of the dislocations in the bcc-martensite through the bcc-fcc interface, resulting in suppression of the DARA effect. Notably, the movement of the dislocations through the bcc-fcc interface is different from that of the dislocations within the martensitic matrix because the latter does not require the complex process of absorption of the $a/2 < 111>$ dislocations by the interface and nucleation of the $a/6 < 112>$ partial dislocations in the fcc-phase, which allows the movement and multiplication of dislocations in the martensitic matrix under dynamic tensile loading to occur. Moreover, adiabatic softening of the martensitic matrix also occurs. Therefore, the competition of adiabatic softening (decrease in the number of dislocations) of the martensitic matrix versus strain rate hardening (multiplication of the dislocations) leads to a drop in the dislocation density in the martensitic matrix, as demonstrated in this work.

The Q&T steel consists mainly of a single martensite. During the dynamic tensile test, strain rate hardening of the martensite occurs in competition with adiabatic softening of the martensite. In the initial deformation...
stage, the strain rate hardening effect of the martensite is predominant; thus, the tensile strength is greater in the dynamic tensile test than in the quasi-static tensile test. During the subsequent deformation prior to fracture, adiabatic softening of the martensite is predominant; hence, the increase in elongation is greater in the dynamic tensile test than that in the quasi-static tensile test. This gives rise to an enhancement in both the strength and ductility of the Q&T steel. For the Q-P-T steel, the factors influencing the dynamic tensile behavior are relatively complex due to the retained austenite. In the initial deformation stage, the strain rate hardening effect of the martensite is predominant. Thus, the tensile strength in the dynamic tensile test is greater than in the quasi-static tensile test. In the subsequent deformation prior to fracture, the adiabatic softening of the martensite matrix in the dynamic tensile test cannot effectively compensate the loss of ductility due to the suppression of the DARA effect, but adiabatic heating of the retained austenite has a significant effect on enhancing the ductility, which gives rise to a slight decrease in the elongation compared with that under quasi-static tension.

Conclusions

For comparative analysis, novel Q-P-T martensitic steel and traditional Q&T martensitic steel were subjected to a dynamic tensile test at a strain rate of 500 s⁻¹ and quasi-static tensile test at a strain rate of 5.6 × 10⁻⁴ s⁻¹. From characterization of the microstructures using XRD and TEM, the main conclusions are as follows. Under dynamic tensile loading, the Q-P-T martensitic steel with a considerable amount of retained austenite shows greater strength, and the elongation is slightly lower than that under quasi-static tensile loading. This differs from the enhancements in both the strength and the elongation of the Q&T martensitic steel that contains little retained austenite under dynamic tensile loading. The experiment verifies that the DARA effect is largely suppressed under dynamic tensile loading for the Q-P-T sample. The suppression of the DARA effect favorably affects the ductility, and cannot be effectively compensated by the adiabatic softening of the martensite matrix; this leads to a slight decrease in the elongation compared with that of the Q-P-T sample under quasi-static tensile loading.

Disclosure statement

No potential conflict of interest was reported by the authors.

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References

[1] Speer J, Matlock DK, De Cooman BC, et al. Carbon partitioning into austenite after martensite transformation. Acta Mater. 2003;51:2611–2622.
[2] Qin SW, Liu Y, Hao QG, et al. Ultrahigh ductility, high-carbon martensitic steel. Metall Mater Trans A. 2016;47:4853–4861.
[3] Rong YH. High strength ductility steels treated by novel quenching–partitioning–tempering process. Int Heat Treat Surf Eng. 2011;5(4):145–154.
[4] Hsu TY, Xu ZY. Design of structure, composition and heat treatment process for high strength steel. Mater Sci Forum. 2007;561–565:2283–2286.
[5] Zhang K, Liu P, Li W, et al. High strength-ductility Nb-microalloyed low martensitic carbon steel: novel process and mechanism. Acta Metall Sin (Engl Lett). 2015;28(10):1264–1271.
[6] Koistinen DP, Marburger RE. A general equation prescribing the extent of the austenite-martensite transformation in pure iron-carbon alloys and plain carbon steels. Acta Metall. 1959;7:59–60.
[7] Zackay VF, Parker ER, Fehr D, Busch D. The enhancement of ductility in high-strength steels. Transcations of the ASM. 1967;60(2):252–259.
[8] Webster D. Increasing the toughness of the martensitic stainless steel AFC77 by control of retained austenite content, ausforming and strain aging. Transactions of the ASM. 1968;61(4):816–828.
[9] Zhang K, Zhang MH, Guo ZH, et al. A new effect of retained austenite on ductility enhancement in high-strength quenching–partitioning–tempering martensitic steel. Mater Sci Eng A. 2011;528:8486–8491.
[10] Woo W, Balogh L, Ungár T, et al. Grain structure and dislocation density measurements in a friction-stir welded aluminum alloy using X-ray peak profile analysis. Mater Sci Eng A. 2008;498:308–313.
[11] Li W, Xu WZ, Wang XD, et al. Measurement of microstructural parameters of nanocrystalline Fe–30wt.%Ni alloy produced by surface mechanical attrition treatment. J Alloy Compd. 2009;474:546–550.
[12] Slycken JV, Verleysen P, Degrieck J, et al. High-strain-rate behavior of low-alloy multiphase aluminum- and silicon-based transformation-induced plasticity steels. Metall Mater Trans A. 2006;37:1527–1539.
[13] Choi ID, Kim DM, Kim SJ, et al. The effect of retained austenite stability on high speed deformation behavior of TRIP steels. Met Mater Int. 2006;12(1):13–19.
[14] Kim S, Lee S. Effects of martensite morphology and volume fraction on quasi-static and dynamic deformation behavior of dual-phase steels. Metall Mater Trans A. 2000;31:1753–1760.
[15] Oliver S, Jones TB, Fournaris G. Dual phase versus TRIP strip steels: microstructural changes as a consequence of quasi-static and dynamic tensile testing. Mater Charact. 2007;58:390–400.
[16] Choi ID, Bruce DM, Kim SJ, et al. Deformation behavior of low carbon TRIP sheet steels at high strain rates. ISIJ Int. 2002;42(12):1483–1489.
[17] Wang L, Speer JG. Quenching and partitioning steel heat treatment. Metallogr Microstruct Anal. 2013;2:268–281.
[18] Yang X, Xiong X, Yin Z, et al. Interrupted test of advanced high strength steel with tensile Split Hopkinson bar method. Exp Mech. 2014;54:641–652.
[19] Hao QG, Qin SW, Liu Y, et al. Effect of retained austenite on the dynamic tensile behavior of a novel quenching-partitioning-tempering martensitic steel. Mater Sci Eng A. 2016;662:16–25.

[20] Kolsky H. An investigation of the mechanical properties of materials at very high rates of loading. Proc Phys, Soc Lond B. 1949;62:676–700.

[21] Huh H, Kang WJ, Han SS. A tension split Hopkinson bar for investigating the dynamic behavior of sheet metals. Exp Mech. 2002;42(1):8–17.

[22] Gerlach R, Sathianathan SK, Siviour C, et al. A novel method for pulse shaping of Split Hopkinson tensile bar signals. Int J Impact Eng. 2011;38:976–980.

[23] Xia KW, Yao W. Dynamic rock tests using split Hopkinson (Kolsky) bar system – a review. J Rock Mech Geotech Eng. 2015;7:27–59.

[24] Xue Q, Cerreta EK, Gray III GT. Microstructural characteristics of post-shear localization in cold-rolled 316L stainless steel. Acta Mater. 2007;55:691–704.

[25] Dougherty LM, Cerreta EK, Pfeif EA, et al. The impact of peak shock stress on the microstructure and shear behavior of 1018 steel. Acta Mater. 2007;55:6356–6364.

[26] Wen CS, Chen Z, Huang BX, et al. Nanocrystallization and magnetic properties of Fe-30 weight percent Ni alloy by surface mechanical attrition treatment. Metall Mater Trans A. 2006;37:1413–1421.

[27] Talonen J, Nenonen P, Pape G, et al. Effect of strain rate on the strain-induced γ → α′-martensite transformation and mechanical properties of austenitic stainless steels. Metall Mater Trans A. 2005;36:421–432.

[28] Lichtenfeld JA, Mataya MC, Tyne CJV. Effect of strain rate on stress-strain behavior of alloy 309 and 304L austenitic stainless steel. Metall Mater Trans A. 2006;37:147–161.

[29] Ha Y, Kim H, Kwon KH, et al. Microstructural evolution in Fe-22Mn-0.4C twinning-induced plasticity steel during high strain rate deformation. Metall Mater Trans A. 2015;46:545–548.

[30] Shao S, Medyanik SN. Interaction of dislocations with incoherent interfaces in nanoscale FCC–BCC metallic bi-layers. Model Simul Mater Sci Eng. 2010;18:055010.