Formability of quenching-partitioning-tempering martensitic steel

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ABSTRACT

A Fe-0.20C-1.49Mn-1.52Si-0.58Cr-0.05Nb (wt%) steel was treated by a novel quenching-partitioning-tempering (Q-P-T) process and traditional quenching and tempering (Q&T) for comparison, respectively. The researches of the mechanical properties and forming limit diagrams reveal that Q-P-T martensitic steel has more retained austenite (10.8%) than the Q&T martensitic steel (less than 3%), which makes Q-P-T martensitic steel possess higher strain hardening exponent and true uniform elongation than the Q&T martensitic steel. The high value of hardening exponent (n) and true uniform elongation (\( \varepsilon_u' \)) stem from the low initial dislocation density in martensitic matrix in Q-P-T martensitic steel before deformation and the high strain hardening rate and dislocation absorption by retained austenite (DARA) effect of retained austenite during deformation. Moreover, the n or \( \varepsilon_u' \) is a most important parameter in all the apparent parameters affecting the formability of metal sheets, and this conclusion is of practical importance in the comparison of several steels’ formability.

Introduction

Advanced high strength steels (AHSSs) have rapidly developed in recent ten years due to the requirement of lightweight, environmental protection and safety in modern automotive industries [1]. To help cope with these often conflicting requirements, dual-phase (DP) steels [2], transformation induced plasticity (TRIP) steels [3], twinning induced plasticity (TWIP) steels [4], quenching and partitioning (Q&P) steels [5] and quenching-partitioning-tempering (Q-P-T) steels [6] were successively developed. As well-known, before these newly developed steels can be used, they often need to be formed into potentially complex parts through processes such as stamping thin strips at room temperature [7]. The complexity of forming during such processes requires careful and constant monitoring of both the mechanical behavior and formability of metal sheets [4,8–12]. The formability of a metal is defined by its ability to deform into a desired shape without any local necking or fracture [13]. As an aid to evaluating the success of sheet forming operations, forming limit diagrams (FLDs) have proven to be a very useful and popular technique for optimizing metal sheet forming processes. There are considerable researches into the correlation between the formability of metal sheets and apparent parameters such as their strain hardening exponent (n), plastic strain ratio (s), true uniform elongation (\( \varepsilon_u' \)), and yield ratio (\( \sigma_y/\sigma_u \)) [17–20]. Studies have revealed that the formability of AHSSs is dependent on the microstructure through different heat treatments [21,22].

Of the various AHSSs available, Q&P and Q-P-T steels possess the highest strength due to their martensitic matrix. However, as they are a relatively new development, much less is known about their formability than DP, TWIP and TRIP steels [23,24]. Therefore, this study [25] on the formability of high strength low-carbon Q-P-T martensitic steel is to reveal the effects of microstructure on apparent parameters and formability by compared to traditional quenching and tempering (Q&T) martensitic steel with the same composition.

Experimental procedure

A low carbon steel was melted in a medium frequency induction furnace, and then a hot-rolled to a 20 mm-thick plate by the Central Iron and Steel Research Institute (Beijing, China). The chemical composition of this steel was analyzed as Fe-0.20C-1.49Mn-1.52Si-0.58Cr-0.05Nb (wt%). Specimens with 4 mm in thickness were cut from this hot-rolled plate, and then subjected to Q-P-T process and Q&T process for comparison, respectively.

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Those samples treated by the Q-P-T process were austenitized at 950°C for 600 s, followed by quenching to 320°C and holding for 30 s. They were then partitioned/tempered at 450°C 60 s and finally quenched to room temperature in water. The Q&T process used was essentially the same as the Q-P-T process, except that the samples were only quenched to room temperature (20°C) in the vicinity of different cracks to investigate the effects of strain on the volume fraction of retained austenite (V\text{RA}) and average dislocation density in the martensite and retained austenite regions of the Q-P-T specimens. As shown in Table 4, these locations included a region of low strain far from the fracture region (No. 1), an area of high strain

Two different forming methods [31] have been developed for experimentally determining the FLD value: in-plane stretching (e.g. the Marciniak test) [32] and out-of-plane stretching (e.g. the Nakazima test) [33]. In the present study, Nakazima tests were carried out in accordance with GB/T 15825.8-2008 and ASTM E2218-02 (2008) using a Zwick/Roell BUP600 sheet metal testing machine to determine the FLDs of the Q-P-T and Q&T steel samples. Figure 1 shows the punched samples at various strain states until failure. All tests were performed with a punch speed of 0.5 mm/s until either local necking or fracture occurred, and the load–displacement data was collected throughout the test. Strain analysis of the specimens was conducted by using an optical measuring system (Auto-Grid) to measure the deformation of a 2 mm square grid pattern electrochemically etched into the surface of the samples (note that the typical FLDs obtained from the literature [7,17] presented in Figure 2 used a circular initial grid). The major and minor strains were subsequently calculated using the following equations [34]:

where $e_1$ and $e_2$ are the major and minor project strain; $d_0, d_1$ and $d_2$ are the initial length and lengths of the major and minor axis; and $e_1$ and $e_2$ are the major and minor true strain, respectively. Finally, a forming limit curve (FLC) was constructed with the minor true strain as the abscissa and the major true strain as the ordinate, as this can be used to distinguish the range of strain between safety and failure in sheet forming.

Microstructural characterization was carried out with a JEOL-2100F transmission electron microscope (TEM) operated at 200 kV using specimens 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. Each specimen was cut close to the vicinity of different cracks to investigate the effects of strain on the volume fraction of retained austenite ($V_{\text{RA}}$) and average dislocation density in the martensite and retained austenite regions of the Q-P-T specimens. As shown in Table 4, these locations included a region of low strain far from the fracture region (No. 1), an area of high strain

where $a = n/0.21$ when $n \leq 0.21$ and the material thickness $t \leq 3$.

In order to obtain the strain hardening exponent from the tensile data, the classic Hollomon equation was used [26,27]:

\[
\sigma = K \varepsilon^n \tag{2}
\]

where $\sigma$ is the true stress, $\varepsilon$ the true strain and $K$ is the strength coefficient. When $\frac{d\sigma}{d\varepsilon} = \sigma$ (necking criterion), the value of $n$ can be written as:

\[
n = \frac{\varepsilon}{\ln (1 + \varepsilon)} \tag{3}
\]

And then

\[
n = \varepsilon_u \tag{4}
\]

where $\varepsilon_u$ is the true uniform elongation.

The plastic strain ratio, $r$, was calculated using the following equation [28]:

\[
r = \frac{e_w}{e_l} = \frac{\ln(w_w/w_l)}{(l/l'_w)/l'_w} \tag{5}
\]

where $e_w$ is the true width strain, $e_l$ the true thickness strain, $w_0$ and $l_0$ are the initial width and gauge length, and $w_f$ and $l_f$ are the final width and gauge length, respectively. The normal anisotropy $\tilde{r}$ was further calculated from the $r$ values using:

\[
\tilde{r} = (r_0 + 2r_{45} + r_{90})/4 \tag{6}
\]

where the subscripts of $r$ represent the 0°, 45° and 90° rolling directions, respectively.

The North American Deep Drawing Research Group (NADDRG) has developed a modified equation for describing the relationship between the value of true major strain (FLD\text{Engineering}) in the plane strain condition and the strain hardening exponent ($n$) to predict FLD [29,30]:

\[
\begin{align*}
\text{FLD}_0^{\text{Engineering Strain}}(\%) &= 23.3 + 14.13 \cdot (t, \text{mm}) \\
\text{FLD}_0^{\text{True Strain}} &= \ln(1 + a \cdot [23.3 + 14.13 \cdot (t, \text{mm})])/100
\end{align*} \tag{7}
\]

where $a = n/0.21$ when $n \leq 0.21$ and the material thickness $t \leq 3$.

To determine the mechanical properties of the Q-P-T and Q&T samples, they were first cut along their tensile axis in directions parallel (0°), diagonal (45°) and perpendicular (90°) to the rolling direction of the sheet. The true stress and true strain were then calculated from their respective engineering values obtained using a Zwick/Roell Z100 tensile testing machine in accordance with GB/T 228–2002 and B/T 5027–2007/ISO 10,113-2006 using the following formula:

\[
\begin{align*}
\sigma_{\text{True}} &= \sigma_{\text{Engineering}}(1 + \varepsilon_{\text{Engineering}}) \\
\varepsilon_{\text{True}} &= \ln(1 + \varepsilon_{\text{Engineering}})
\end{align*} \tag{1}
\]

where $\sigma$ is the true stress, $\varepsilon$ is the true strain and $K$ is the strength coefficient. When $\frac{d\sigma}{d\varepsilon} = \sigma$ (necking criterion), the value of $n$ can be written as:

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\end{align*} \tag{7}
\]
near the fracture region (No. 3), and an area of medium strain between them (No. 2). At each of these, the $V_{RA}$ was measured by X-ray diffraction (XRD) [35] and the average dislocation density by X-ray diffraction linear profile analysis (XLPA) [36,37] with a D/max-2550VL/PC.

**Results and discussions**

**Uniaxial tensile properties and apparent parameters**

To understand the behavior of Q-P-T and Q&T steels during forming, uniaxial tensile tests were performed.

*Figure 1.* Deformed shapes of (a) Q-P-T and (b) Q&T samples after Nakazima testing.

*Figure 2.* Typical FLD showing the five main strain states of sheet forming.
on each sample at room temperature in three directions (0°, 45° and 90° to the rolling direction) along with FLD tests. Figure 3 shows the engineering stress vs. engineering strain curves under uniaxial tension, with the mechanical properties being listed in Table 1. These results indicate that the yield strength (YS) and ultimate tensile strength (UTS) of Q-P-T steel is lower in all three directions than in the Q&T steel. The yield ratio (YS/UTS) of the Q-P-T steel was similarly lower in all three directions, but its total elongation (TE) was higher in all three directions than in the Q&T samples. Using the uniaxial tensile test data, true stress-true strain curves were plotted based on Equation (1), as shown in Figure 4. The strain hardening exponent, $n$, and strength coefficient, $K$, were further calculated based on Equation (2) by fitting the true stress-true strain curves of the Q-P-T samples and Q&T ones along with three different directions. These fitting results are listed in Table 2. From the $n$ and $K$ values listed in Table 2, it is evident that both values are higher in Q-P-T samples than in the Q&T ones. Furthermore, the minimum and maximum values in Q-P-T samples occur at 45° and 0° to the rolling direction, respectively, whereas with the Q&T samples the minimum and maximum occur at 90° and 0°. True uniform elongation ($\epsilon_{um}$) in Q-P-T samples was higher in all three directions than in the Q&T samples. The plastic strain ratio, $r$, and normal anisotropy, $\bar{r}$, also listed in Table 2 were calculated using Equations (5) and (6), respectively. These demonstrate that the plastic strain ratio and normal anisotropy is higher in all three directions in Q-P-T steel than in the Q&T samples.

**Table 1.** Mechanical properties of Q-P-T and Q&T steels obtained from uniaxial tensile tests.

| Steel | Direction | YS (MPa) | UTS (MPa) | TE (%) | PSE (MPa-%) | YS/UTS (\(\epsilon_{um}/\epsilon_{b}\)) |
|-------|------------|----------|-----------|--------|-------------|----------------------------------|
| Q-P-T | 0°         | 1046     | 1288      | 16.9   | 21,767      | 0.8121                           |
|       | 45°        | 1037     | 1268      | 16.4   | 20,795      | 0.8178                           |
|       | 90°        | 1049     | 1281      | 16.1   | 20,624      | 0.8189                           |
| Q&T   | 0°         | 1278     | 1387      | 10.9   | 15,118      | 0.9214                           |
|       | 45°        | 1267     | 1368      | 10.5   | 14,364      | 0.9262                           |
|       | 90°        | 1216     | 1299      | 9.8    | 12,730      | 0.9361                           |

**Figure 3.** Engineering stress vs. engineering strain for (a) Q-P-T and (b) Q&T steel sheets at 0°, 90° and 45° to their rolling direction.

**Figure 4.** True stress vs. true strain for (a) Q-P-T and (b) Q&T steel sheets at 0°, 90° and 45° to their rolling direction.

The FLDs of the Q-P-T and Q&T samples were determined by measuring the strain of deformed grids near cracks at various strain states, as listed in Table 3. Figure 5(a,b) present the FLDs of Q-P-T and Q&T steels, respectively, in which it can be seen that experimental data in the tension-tension (T-T) region of the Q-P-T samples exhibits a smaller scatter than that of the Q&T steel. As this likely relates to the existence of considerable retained austenite in Q-P-T steel, fitted curves were plotted, as shown in Figure 5(c), with a safe working zone being established below this line, and a failure zone above it. As these fitted lines obviously cannot precisely indicate the onset of failure, they should ultimately be replaced by a banded region [18]. The strain limits for the Q-P-T and Q&T steel sheets under various strain states given in Table 4 show that in the tension-compression
(T-C) region, the maximum major true strain ($\varepsilon_1$) and minor true strain ($\varepsilon_2$) of the Q-P-T steel sheet were 0.26 and $-0.09$, respectively, while in the Q&T steel these values were 0.24 and $-0.08$. Under plane strain (P-S) conditions, the maximum $\varepsilon_1$ of the Q-P-T and Q&T steel sheets were 0.24 and 0.13, respectively. In the T-T region, the maximum $\varepsilon_1$ and $\varepsilon_2$ of the Q-P-T steel sheet were 0.51 and 0.25, respectively, while the Q&T sheet had values of 0.45 and 0.19. This means that the safe working zones of the T-T region and T-C region increase with strain limit and is therefore larger in the case of the Q-P-T sheet, especially in the tension-tension region (Figure 5(c)). This is mainly attributed to the larger FLD$_0$ and $\varepsilon_2$ of the Q-P-T sheet, with the fact that the safe working zone of both steels is much higher in the tension-tension region than in the tension-compression region, likely being a feature of martensitic steels.

### Microstructural characterization

The XRD spectra of the Q-P-T samples before and after Nakazima punch testing (Figure 6) at different strain regions indicate that it consisted of bcc-martensite and fcc-retained austenite, with $V_{\text{RA}}$ determined to be 10.8%. Subsequent TEM characterization revealed that the bcc-phase consists of dislocation-type martensite laths, as shown in the bright field (BF) image in Figure 7(a), while the fcc-phase is made up of flake-like retained austenite between martensite laths, as shown in the dark field (DF) image and inserted selected area electron diffraction (SAED) pattern in Figure 7(b). The SAED pattern in Figure 7(b) also reveals a K-S orientation relationship. In contrast, no peak for retained austenite could be found in the XRD spectra of the Q&T steel (Figure 6(a)), which would suggest that it consists of single-phase martensite. However, TEM characterization revealed that it consists of dislocation-type martensite with a small

| Table 2. Formability parameters of Q-P-T and Q&T steels. |
|----------------|----------------|-------------|-------------|-------------|-------------|-------------|
| Steel          | Direction     | $\varepsilon_u$ (%) | $\bar{\varepsilon}_u$ (%) | $n$          | $\bar{n}$     | K (MPa)     | $r$         | $\bar{r}$ |
| Q-P-T          | $0^\circ$     | 10.5         | 10.3        | 0.1058      | 0.0958       | 1812        | 0.2406      | 0.2717     |
|                | $45^\circ$    | 10.2         | 10.1        | 0.0923      | 0.0826       | 1725        | 0.2731      | 0.1731     |
|                | $90^\circ$    | 3.9          | 3.6         | 0.00463     | 0.0104       | 1607        | 0.1793      | 0.0897     |
| Q&T            | $0^\circ$     | 3.6          | 3.3         | 0.0401      | 0.0371       | 1511        | 0.1839      | 0.0973     |

| Table 3. Details of forming limit strains. |
|----------------|----------------|-------------|-------------|-------------|-------------|
| Sample         | Tension-compression strain Major strain | Minor strain | Plane strain Major strain | Minor strain | Tension-tension strain Major strain | Minor strain | FLD$_0$ Major strain |
| Q-P-T          | 0.26           | $-0.09$      | 0.24        | 0.51        | 0.25        | 0.22        |
| Q&T            | 0.24           | $-0.08$      | 0.13        | 0.45        | 0.19        | 0.10        |

Figure 5. Experimentally determined forming limit diagrams of: (a) Q-P-T steel, (b) Q&T steel sheets and (c) comparison of fitted FLDs for Q-P-T and Q&T steel.
amount of film-like retained austenite (Figure 7(c)), the latter being identified by the SAED pattern in Figure 7 (d). This indicates that the volume fraction of retained austenite in Q&T specimen is less than 3%, as this was the XRD detection limit.

**Measurement of average dislocation density and retained austenite fraction during forming**

From the XRD spectra obtained from different strain regions of Q-P-T steel with different strain paths (Figure 6), the variation with strain in average dislocation density in both the martensitic matrix and retained austenite was determined, as listed in Table 4. This shows that prior to forming, the average dislocation density in the martensitic matrix and retained austenite is $6.59 \times 10^{14}$ and $11.62 \times 10^{14}$ m$^{-2}$, respectively. With increasing strain (from No. 1 to No. 3 in Table 4), the average dislocation density in the martensitic matrix first reduces, and then increases, which shows obviously dislocation absorption by retained austenite (DARA) effect [38]. The DARA effect

![Figure 6](image1.png)  
**Figure 6.** XRD spectra of Q&T and Q-P-T steel at different strain states for: (a) undeformed, (b) T-C, (c) P-S and (d) T-T FLD samples.

![Figure 7](image2.png)  
**Figure 7.** (a) BF TEM micrograph and (b) DF image with SAED pattern of retained austenite in Q-P-T steel. (c) BF image and (d) DF image with SAED pattern of retained austenite in Q&T steel.
evidently enhances the deformation ability of hard-phase martensitic matrix, whereas the average dislocation density in the retained austenite increases continuously, as a result, the strain energy accumulated by dislocation multiplication inside the austenite grains is what provides the mechanical driving force needed for the strain-induced transformation of retained austenite that accompanies the formation of twinned martensite, which further hardens low-carbon Q-P-T martensitic steel. Previous work has indicated that the average dislocation density \((\rho_0)\) in the martensite of this Q-P-T steel is \(6.59 \times 10^{14} \text{ m}^{-2}\) before deformation, but is \(6.78 \times 10^{14} \text{ m}^{-2}\) in Q&T steel, and so it is the lower dislocation density of Q-P-T steel that leads to its lower yield strength [39]. It is therefore believed that high \(n\) value of Q-P-T steel stems from the low initial dislocation density in its martensitic matrix before deformation and the high strain hardening rate of retained austenite during deformation. Similarly, the low \(n\) value of Q&T steel stems from the high initial dislocation density in its martensitic matrix and low strain hardening rate during deformation. The softening of the martensitic matrix due to carbon depletion during the Q-P-T process and the DARA effect during deformation effectively enhance the formability of the martensitic matrix, moreover, the soft retained austenite phase possesses an intrinsic ductility that leads to a high uniform elongation \((\varepsilon_u)\) of low-carbon Q-P-T martensitic steel. Since there is almost no retained austenite and a single martensite with high dislocation density in Q&T steel, it has a low uniform elongation.

### Effect of apparent parameters on formability

In order to understand the origin of a material’s behavior during forming, one must analyze the relation of apparent parameters each other and their effects on formability. For example, the strain hardening exponent, \(n\), is proportional to strain, \(\varepsilon\), and reversely proportional to stress, \(\sigma\), as shown in Equation (3). The strain hardening exponent means that a larger \(n\) value produces a lower yield strength \(\sigma_y\) and higher tensile strength \(\sigma_b\) (i.e. a low yield ratio, \(\sigma_y/\sigma_b\)). It is also evident from Equation (3) that when necking occurs (i.e. \(d\sigma/d\varepsilon = 0\)), \(n\) is equal to the true uniform elongation \((\varepsilon_u)\), which means that \(n\) reflects both the strength and ductility. The value of \(\varepsilon_u\) on the other hand, represents the deformability of the material immediately prior to necking or crack initiation, and so is related to the plastic strain ratio, \(r\). We can see from Equation (5) that a large value of \(r\) coincides with a large true uniform elongation, and as \(r\) is related by Equation (6) to the normal anisotropy \((\bar{r})\), a large value of \(r\) reflects a large \(r\) despite describing the normal anisotropy. Based on this analysis, the strain hardening exponent, \(n\), or the true uniform elongation, \(\varepsilon_u\) is considered a most important parameter in all the apparent parameters affecting the formability of metal sheets. This conclusion is of importance in the comparison of several steels’ formability because simple tensile test for \(n\) or \(\varepsilon_u\) can replace complex forming test to qualitatively judge their relative formability.

### Conclusions

The FLDs of Fe-0.20C-1.49Mn-1.52Si-0.58Cr-0.05Nb (wt %) steel treated by a novel Q-P-T process and Q&T one for comparison were measured through both uniaxial tensile tests and Nakazima punching tests, respectively. The microstructures were characterized by XRD, and TEM, and main conclusions are described as follows.

1. The yield strength and ultimate tensile strength of Q-P-T steel is lower in all three directions (0°, 45° and 90°) than in the Q&T steel. The yield ratio of the Q-P-T steel was similarly lower in all three directions, but its total elongation was higher in all three directions than in the Q&T samples.
2. The strain hardening exponent, \(n\), and the true uniform elongation, \(\varepsilon_u\) are higher in Q-P-T samples than in the Q&T ones, which stems from the low initial dislocation density in martensitic matrix in Q-P-T martensitic steel before deformation and the high strain hardening rate and DARA effect of retained austenite during deformation.
3. The DARA effect in Q-P-T steel during the forming test was found, but the DARA effect does not exist in Q&T steel since the latter has hardly retained austenite.
4. The strain hardening exponent, \(n\), or the true uniform elongation, \(\varepsilon_u\) is a most important parameter in all the apparent parameters affecting the formability of metal sheets, and this conclusion is of practical importance in the comparison of several steels’ formability.

### Disclosure statement

No potential conflict of interest was reported by the authors.
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