Stretch-flangeability of a High-strength TRIP Type Bainitic Sheet Steel

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The stretch-flangeability of a newly developed high-strength bainitic sheet steel which is associated with the transformation-induced plasticity (TRIP) of retained austenite, or “TRIP type bainitic steel” was investigated for the automotive applications. An excellent stretch-flangeability was completed in the steel composing of bainitic ferrite matrix and interlath retained austenite films without initial blocky martensite. In this case, the stable or carbon-enriched retained austenite films enhanced the stretch-flangeability due to the reduction of the surface damage on hole-punching and the promotion of the TRIP effect on hole-expanding. Also, uniform fine bainitic ferrite lath structure contributed to improving the stretch-flangeability due to the increased localized ductility and development of severe plastic flow.

KEY WORDS: retained austenite; TRIP; strain-induced transformation; high-strength steel; austempering; heat-treatment; formability; stretch-flangeability.

1. Introduction

The transformation-induced plasticity (TRIP) of retained austenite is very useful in improving the formability of high-strength sheet steel. Low carbon dual phase sheet steel associated with the TRIP effect or “TRIP type multi-phase (TMP) sheet steel” which was developed for automotive applications before the last decade is one of the successful examples. Recently, the TMP steel has been applied to some automotive impact members as it improves the impact energy absorption characteristics. However, any application to the suspension parts such as lower arms which can lead to a significant weight reduction has not been progressed due to its poor stretch-flangeability, in spite of excellent stretch-formability, good deep drawability and high fatigue strength.

The poor stretch-flangeability of the TDP steel may be essentially overcome by changing the ferrite matrix for bainite or bainitic ferrite matrix because the bainitic steel possesses an excellent stretch-flangeability by a uniform fine lath structure. On the basis of this idea, we have recently developed a new type of low alloy high-strength cold-rolled bainitic sheet steel containing the retained austenite in the bainitic ferrite matrix or “TRIP type Bainitic (TB) sheet steel”. In the present paper, the stretch-flangeability of the TB steel subjected to different austempering treatment was investigated in accompany with the examination of punched hole-surface damage. In addition, the stretch-flangeability was related to some metallurgical factors such as matrix structure and retained austenite characteristics.

2. Experimental Procedure

In the present study, a TB steel with chemical composition as listed in Table 1 was prepared as a vacuum-melted 100 kg ingot followed by hot forging to produce 30 mm thick slabs. Martensite-start temperature ($M_s$) of the steel was estimated to be 417°C from the following equation:

$$
M_s(°C) = 561 - 474 \times C(\text{mass}%) - 33 \times Mn(\text{mass}%) - 17 \times Ni(\text{mass}%) - 17 \times Cr(\text{mass}%) - 21 \times Mo(\text{mass})
$$

Table 1. Chemical composition (mass%) and heat-treatment conditions of steels used.

| Steel | C  | Si  | Mn  | P  | S  | Al  | N  | annealing | austempering |
|-------|----|-----|-----|----|----|-----|----|-----------|--------------|
| TB    | 0.20 | 1.51 | 1.51 | 0.015 | 0.011 | 0.040 | 0.0021 | 950°C, 1200s | 350~475°C, 200s OQ |
| TMP   | 0.18/0.21 | 0.014/0.015 | 0.0017 | 0.036/0.041 | 0.0017/0.0027 | 780~860°C, 1200s | 400°C, 1000s OQ |
| DP    | 0.14 | 0.21 | 1.74 | 0.013 | 0.003 | 0.037 | 0.0028 | 780°C, 1200s OQ | 200~480°C, 3600s OQ |
| B     | 0.14 | 0.21 | 1.74 | 0.013 | 0.003 | 0.037 | 0.0028 | 950°C, 1200s | 300~450°C, 1000s |

*: tempering, OQ: quenching in oil

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The slabs were reheated to 1,200°C and were hot-rolled to 3.2 mm in thickness with finishing at 850°C and then air-cooled to room temperature. After cold-rolling to 1.2 mm in thickness, they were annealed at 950°C for 1,200 s and then were austempered at temperatures between 350 and 475°C for 200 s in a salt bath, followed by cooling in oil to 20°C. In this case, the austempering time was decided to obtain a large amount of stable retained austenite and large elongations.10)

For comparison, several TMP steels with different silicon or manganese content, conventional dual phase (DP) steels tempered at 200–600°C and bainitic (B) steels austempered at 300–450°C, with chemical composition as listed in Table 1, were also prepared. Heat-treatment conditions, second phase characteristics and tensile properties of these steels are given in Table 2. Modified LePera’s reagent,12) as well as nital regent, was used to distinguish martensite and/or bainite from retained austenite. The amount of retained austenite was quantified by X-ray diffractometry using Mo-Kα radiation. To minimize the effect of texture, the volume fraction of retained austenite was quantified on the basis of the integrated intensity of (200)α, (211)α, (200)γ, (220)γ, and (311)γ diffraction peaks, termed the five-peak method.13) The retained austenite lattice constant (aγ) was measured from (220)γ diffraction peak using Cr-Kα radiation on the electrochemically polished surface, with a negligible internal stress. Substituting the measured aγ value (×10−10 m) into the following equation,14) carbon concentration of the retained austenite (Cγ, mass%) was calculated.

\[
a_{\gamma} = 3.5467 + 0.0467C_{\gamma} \quad \text{(2)}
\]

Hole-punching and hole-expanding tests were conducted using disc specimens of 50 mm in diameter by 1.2 mm in thickness with a graphite type lubricant. Before hole-expanding test, a hole of 4.76 mm in diameter was punched out at 20°C and at a punch rate of 10 mm/min, with a clearance of 10% between die and punch. Then, hole-expanding tests were performed at 20°C and at a punch rate of 1 mm/min, using the apparatus illustrated in Fig. 1 in which expanding punch was contacted with the roll-over portion of the hole-punched specimens. The hole-expanding ratio (λ) was determined in the following equation.

\[
\lambda = \left( \frac{d_{f} - d_{i}}{d_{i}} \right) \times 100\% \quad \text{(3)}
\]

\(d_{i}\): initial hole diameter, 
\(d_{f}\): hole diameter on cracking

Surface damage characteristics on hole-punching such as a critical hardening depth (l*), a hardness increment (ΔHV)
and the aspects of void formation were examined using the micro-Vickers hardness tester (test weight: 0.25 N) and scanning electron microscope. The hardness increment was measured at the depth of 50 μm beneath the punched surface.15)

3. Results

3.1. Microstructure

Figure 2 shows the variations in retained austenite characteristics as a function of austempering temperature ($T_A$) in the TB steel. The initial carbon concentration of retained austenite ($C_g^0$) is between 1.25 and 1.52 mass% and linearly decreases with increasing austempering temperature. As listed in Table 2, the carbon concentration is higher than that of TMP-1.5 Mn steel with the same chemistry. On the other hand, the initial volume fraction of retained austenite ($f_g^0$) ranging from 8 to 12 vol% is nearly constant in the steels austempered at temperatures between 400 and 450°C which are close to the martensite start temperature of the present steel ($M_s=417°C$).

Figures 3 and 4 show typical micrographs of the TB steel. The microstructure is principally characterized by bainitic ferrite lath matrix and interlath retained austenite films (Fig. 4). In the steels austempered at temperatures above $M_s$, however, a large amount of blocky martensite and quasi-ferrite16) coexist with coarsened bainitic ferrite lath and retained austenite films (Figs. 3(d)–3(f)). If austempered at 475°C, a small amount of bainite island also coexists in the TB steel.

Retained austenite stability against the strain-induced transformation or “k-value” which is defined as the following equation2) is shown in Fig. 2(c).

$$\log f_g = \log f_g^0 - k \cdot \varepsilon_P \quad \text{.........}(4)$$

Fig. 2. Variations in (a) initial volume fractions of second phase ($f$), retained austenite ($f_g$), martensite ($f_m$) and bainite ($f_b$), (b) initial carbon concentration of retained austenite ($C_g^0$) and (c) k-value as a function of austempering temperature in TB steel.

Fig. 3. Scanning electron micrographs of TB steels austempered at temperatures ranging from $T_A=350°C$ to 475°C, in which $\alpha_b$, $\alpha_b$, $\alpha_q$, $\alpha_m$ and $f_g$ represent bainitic ferrite, bainite, quasi-ferrite, martensite and retained austenite, respectively.
the previous report.5)
(Hmax), as shown in Fig. 6(b). Therefore, the TB steel is concluded to possess the best combination of stretch-flangeability and stretch-formability among several types of high-strength steels.

**Figure 7** shows a relation between the TS×λ value and localized ductility such as (a) reduction of area (RA) and (b) elongation for 5 mm gauge length (El[5]) in the TB steel. From this figure, it is found that the TS×λ value of the TB steel exhibits a positive correlation with the localized ductility, in the same way as those of the TMP, DP and B steels. In addition, the TB steels austempered at temperatures below MS possess larger TS×λ value under constant RA and El[5] values.

### 3.3. Surface Damage Characteristics of Punched Hole

**Figure 8** shows typical scanning electron micrographs of cross sectional area in punched hole-surface layer of the TB steel. Many voids are formed at the matrix/coarse second phase interface in the break section, particularly in the TB steels austempered at temperatures above Ms. The coarse second phases seem to be initial blocky martensite particles.

**Figure 9** shows a shear stress (τmax) and surface damage characteristics on hole-punching in the TB and the TMP steels. The surface damage of the TB steels austempered at 425–475°C is characterized by a large critical hardening depth (l*) and a relatively short shearing section length (ss), although the ss values are larger than that (0.2 mm) of the TMP-1.5Mn steel with the same chemistry. On the other hand, the TB steels austempered at temperatures between 350 and 400°C possess smaller l* values and larger ss values. According to the previous study,15) the above mentioned l* value agrees well with a critical depth inside which the strain-induced martensite transformation occurs. So, the small l* value is corresponding to a little strain-induced martensite in the punched hole-surface layer.

A hardness increment (ΔHV) near the punched hole-surface is ranging from 108 to 155 in the TB steels. However, austempering temperature dependence of the hardness increment is indistinct.

### 4. Discussion

#### 4.1. Relation between Stretch-Flangeability and Retained Austenite Characteristics

According to the previous study,5) the stretch-flangeability of the TMP steel was mainly controlled by initial carbon concentration of the retained austenite, i.e., retained austenite stability. So, the TS×λ value in the present TB steels was plotted for the carbon concentration of retained austenite, as shown in **Fig. 10**(b). From this figure, it is found that large TS×λ value of the TB steel is apparently achieved by the increased retained austenite stability, similar to that of the TMP steel.5) The optimum volume fraction of retained austenite is not obvious although it is very important for a practical use. If analogous data of 0.2C–(1.0–2.5)Si–(1.0–2.0)Mn, mass%, TB steels²⁰ were added in the figure, the optimum volume fraction is determined to be about 2–3 vol%.

It is noteworthy that the TB steels possessed higher TS×λ values than the TMP steels under conditions of constant carbon concentration and volume fraction of retained austenite, as shown in Fig. 10. This is supposed to be
caused by uniform fine lath structure matrix in the TB steels.

4.2. Effects of Metallurgical Factors on Stretch-Flangeability

The present TB steels austempered at temperatures below $M_S$ exhibited the excellent stretch-flangeability in the tensile strength range over 900 MPa, accompanied with good stretch-formability (Fig. 6), although they possessed a relatively poor localized ductility. This reason must be discussed on the basis of microstructure and retained austenite stability.

First, let us discuss about the different in stretch-flangeability between the TB and TMP steels. In general, the stretch-flangeability of the steel can be significantly controlled by hole-surface damage characteristics on punching, as well as the localized ductility. When the hole-surface damage characteristics of the TB steels were compared to that of the TMP-1.5 Mn steel with the same chemical composition (Fig. 9), they were characterized by severer plastic flow, longer shearing section, a smaller number of voids and smaller critical hardening depth (or more volume fraction of untransformed retained austenite) than those of the TMP steel although the maximum hardness and hardness increment near the punched hole-surface are somewhat higher than those of the TMP steel. This indicates that the excellent stretch-flangeability of the TB steels was resulted from nonserious surface damage on punching due to “uniform fine lath structure matrix” and “more stable retained austenite films”, because the localized ductilities of the TB steels are smaller than those of the TMP steel. This indicated that the excellent stretch-flangeability of the TB steels is considered to reduce the fracture strain or cause easy crack propagation on hole-expanding, despite the TRIP effect of untransformed retained austenite.

Next, let us discuss about the effects of austempering temperature on the stretch-flangeability of the TB steel. When austempered at temperatures above $M_S$, the stretch-flangeability of the TB steels was deteriorated up to that of the TMP steels (Figs. 6(a) and 7), although they possessed larger localized ductility than the other TB steels. And, these TB steels included many initial blocky martensites and retained austenite films with low carbon concentration which resulted in large hole-surface damage such as many voids and a small amount of untransformed retained austenite, respectively. Thus, the serious surface damage is considered to reduce the fracture strain or cause easy crack propagation on hole-expanding, despite the TRIP effect of untransformed retained austenite.

5. Conclusions

The stretch-flangeability of the TB steel with stable interlath retained austenite films of 5–11 vol% in the bainitic ferrite lath matrix was investigated. The results are summarized as follows.

(1) The TB steel completed the best stretch-flangeability of several high-strength steels, accompanied with good stretch-formability, particularly when austempered at temperatures lower than $M_S$.

(2) The stretch-flangeability exhibited a good positive correlation to the carbon concentration of retained austenite and a negative correlation to the retained austenite content, in the same way as the TMP steel.

(3) The excellent stretch-flangeability of the TB steel was considered to be principally resulted from uniform fine lath structure matrix and more stable retained austenite films. They contributed to the small surface damage and development of severe plastic flow on hole-punching, and consequently enhanced the hole-expanding ratio by suppressing crack propagation and TRIP effect of untransformed retained austenite.

(4) Blocky initial martensites in the TB steel may result in serious surface damage on punching and easy crack propagation on the successive hole-expanding, because they behaved as stress concentration sites.

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