Retained Austenite Characteristics and Stretch-flangeability of High-strength Low-alloy TRIP Type Bainitic Sheet Steels

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Retained austenite characteristics and stretch-flangeability in low alloy TRIP type bainitic sheet steels with different silicon and manganese contents were investigated for automotive applications. As increasing silicon and manganese contents, an initial volume fraction of retained austenite film along bainitic ferrite lath boundary was increased in accompany with a decrease in the carbon concentration. An excellent stretch-flangeability was completed in the steels containing a small amount of stable retained austenites (i.e., volume fraction of 2–4 vol% and carbon concentration of more than 1.0 mass%). This was caused by small surface damage on hole-punching and effective strain-induced transformation plasticity of untransformed retained austenite on hole-expanding. When austempered at temperatures less than MS of the steel after intercritical annealing, further superior stretch-flangeability was achieved due to absence of initial blocky martensite, resulting from developments of long shear section and severe plastic flow and difficult void-initiation on hole-punching.

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

1. Introduction

The transformation-induced plasticity (TRIP)\(^\text{1)}\) of retained austenite is very useful in improving the formability of high-strength sheet steels. Low carbon multi-phase sheet steel associated with the TRIP or “TRIP-aided multi-phase (TMP) sheet steel”\(^\text{2–9)}\) which was developed for the 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.\(^\text{3)}\) However, any application to the suspension parts such as suspension lower arms which can lead to a significant weight reduction has not been progressed due to its poor stretch-flangeability,\(^\text{4,5)}\) in spite of excellent stretch-formability,\(^\text{6)}\) good deep drawability\(^\text{7)}\) and high fatigue strength.\(^\text{8,9)}\)

The poor stretch-flangeability of the TMP steel may be essentially overcome by replacing the ferrite matrix with bainitic ferrite matrix because the bainitic steel generally possesses excellent stretch-flangeability due to uniform fine lath structure. On the basis of this idea, we have recently developed a new type of high-strength low-alloy cold-rolled bainitic sheet steel or “TRIP type bainitic (TB) sheet steel”\(^\text{10,11)}\) composing of bainitic ferrite matrix and interlath retained austenite films. However, relationships between stretch-flangeability and retained austenite parameters in the TB steels has not been investigated.

In the present paper, the retained austenite characteristics of the TB steels with different silicon and manganese contents were examined. And, the stretch-flangeability was related to them to clarify optimum microstructure of the TB steels.

2. Experimental Procedure

In the present study, steels A–G with different silicon and manganese contents as listed in Table 1 were prepared as vacuum-melted 100 kg ingots followed by hot forging to produce 30 mm thick slabs. Martensite-start temperatures (\(M_s\)) of the steels were estimated to be between 379 and 428°C from the following equation.\(^\text{12)}\)

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

where \(C, Mn, Ni, Cr\) and \(Mo\) are alloy contents in steels.

The slabs were reheated to 1200°C and then hot-rolled to 3.2 mm in thickness with finishing at 850°C followed by air-cooling to room temperature, as illustrated in Fig. 1. After cold-rolling up to 1.2 mm in thickness, they were annealed at 950°C for 1200 s and subsequently austempered at temperatures ranging from \(T_A = 350\) to 500°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 both a large amount of stable retained austenite and large elongations.\(^\text{61)}\)

Modified LePera etching,\(^\text{13)}\) as well as nital etching, was used to distinguish each constituent phase. 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)\textit{a}, (211)\textit{a}, (200)\textit{g}, (220)\textit{g}, and (311)\textit{g} diffraction peaks.\textsuperscript{14} The retained austenite lattice constant (\textit{a}_g) was measured from (220)\textit{g} diffraction peak using Cr-Kα radiation on the electrochemically polished surface with a negligible internal stress. Substituting the measured \textit{a}_g value (3.10210^{-10} m) into the following equation,\textsuperscript{15} carbon concentration of the retained austenite (C_g, mass\%) was calculated.

\[ C_g = (a_g - 3.578) / 0.033 \] ..................................(2)

Hole-punching and hole-expanding tests were conducted using the apparatus illustrated in Fig. 2 and disc specimens of 50 mm in diameter by 1.2 mm in thickness with a graphite type lubricant. First, a hole of 4.76 mm in diameter was punched out at 20°C and at a punching rate of 10 mm/min, with a clearance of 10% between die and punch. The successive hole-expanding test was performed at 20°C and at a punching rate of 1 mm/min. In this test, a punch was contacted with the roll-over section of the hole-punched specimens. The hole-expanding ratio (\textit{λ}) was determined in the following equation.

\[ \lambda = \{d_f - d_i\}/d_i \times 100\% \] ..................................(3)

\( d_i \): initial hole diameter
\( d_f \): hole diameter on cracking

3. Results

3.1. Microstructure and Ductility

Figure 3 shows variations in initial volume fraction (\textit{f}_0) and initial carbon concentration (C_{0g}) of retained austenite as a function of austempering temperature (\textit{T}_A) in TB steels (steels A–G). It is found that the initial retained austenite content becomes maximum when austempered at temperatures near \( M_S \), although the carbon concentration of retained austenite linearly decreases with increasing austempering temperature. In addition, the retained austenite content increases with increasing silicon and manganese contents, although the carbon concentration is lowered on the contrary, in the same way as the TMP steels.\textsuperscript{16}

Figures 4 and 5 show typical micrographs of the TB steels. The microstructure is principally characterized by bainitic ferrite lath matrix and interlath retained austenite films (Fig. 5). In all the TB steels austempered at temperatures above \( M_S \), however, a large amount of blocky martensite and quasi-ferrite\textsuperscript{17} coexist with coarsened bainitic ferrite lath and retained austenite films (Fig. 4). The retained austenite films tend to thicken somewhat in high silicon and high manganese steels.

Figure 6 shows tensile properties of the TB steels. The total elongation (TEI) is considerably increased when austempered at temperatures higher than \( M_S \) of the steel, with a decrease in tensile strength (TS). Note that reduction

<table>
<thead>
<tr>
<th>Steel</th>
<th>C</th>
<th>Si</th>
<th>Mn</th>
<th>P</th>
<th>S</th>
<th>Al</th>
<th>N</th>
<th>Mo</th>
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</thead>
<tbody>
<tr>
<td>A</td>
<td>0.21</td>
<td>1.51</td>
<td>1.00</td>
<td>0.015</td>
<td>0.013</td>
<td>0.041</td>
<td>0.010</td>
<td>0.0017</td>
</tr>
<tr>
<td>B</td>
<td>0.20</td>
<td>1.51</td>
<td>1.51</td>
<td>0.015</td>
<td>0.011</td>
<td>0.040</td>
<td>0.013</td>
<td>0.0021</td>
</tr>
<tr>
<td>C</td>
<td>0.20</td>
<td>1.49</td>
<td>1.99</td>
<td>0.015</td>
<td>0.015</td>
<td>0.039</td>
<td>0.0020</td>
<td>0.0017</td>
</tr>
<tr>
<td>D</td>
<td>0.21</td>
<td>1.50</td>
<td>2.51</td>
<td>0.014</td>
<td>0.017</td>
<td>0.038</td>
<td>0.0009</td>
<td>0.0027</td>
</tr>
<tr>
<td>E</td>
<td>0.20</td>
<td>1.00</td>
<td>1.50</td>
<td>0.014</td>
<td>0.013</td>
<td>0.038</td>
<td>0.0010</td>
<td>0.0026</td>
</tr>
<tr>
<td>F</td>
<td>0.18</td>
<td>2.00</td>
<td>1.50</td>
<td>0.015</td>
<td>0.013</td>
<td>0.037</td>
<td>0.0012</td>
<td>0.0021</td>
</tr>
<tr>
<td>G</td>
<td>0.19</td>
<td>2.48</td>
<td>1.49</td>
<td>0.014</td>
<td>0.013</td>
<td>0.036</td>
<td>0.0014</td>
<td>0.0027</td>
</tr>
</tbody>
</table>

Fig. 1. Hot and cold rolling process and heat treatment diagram of steels, in which “AC” and “OQ” denote air cooling and quenching in oil, respectively.

Fig. 2. Experimental apparatus for (a) hole-punching and (b) hole-expanding.

Fig. 3. Variations in initial volume fraction (\textit{f}_0) and initial carbon concentration (C_{0g}) of retained austenite as a function of austempering temperature (\textit{T}_A) in TB steels.
of area (\(RA\)) of the TB steels is extremely large and austempering temperature dependence of the reduction of area is relatively small, except for steels A and D.

### 3.2. Stretch-flangeability

Figure 7 shows the variations in hole-expanding ratio...
(λ) and strength–stretch-flangeability balance (TS×λ: a product of tensile strength and hole-expanding ratio) with austempering temperature (TA) in TB steels. Large λ and TS×λ values are achieved in low silicon and low manganese steels (steels A and E) austempered at temperatures below Ms. It is noteworthy that austempering temperature dependence of the λ value is analogous to that of RA (Fig. 6).

Figure 8 compares hole-expanding ratio of four types of high-strength steels, as well as the stretch-formability (maximum stretch height, Hmax). The TB steels exhibit superior stretch-flangeability to the other steels in a range of TS=800–1,500 MPa, particularly when austempered at temperatures below Mc. Also, the TB steels complete good stretch-formability. Therefore, the TB steel is concluded to possess the best combination of stretch-flangeability and stretch-formability among four types of high-strength steels.

Figure 9 shows typical scanning electron micrographs of cross sectional area in punched hole-surface layer of steel B austempered at (a) 400°C or (b) 425°C, in which arrows denote void initiation sites.

In general, the hole-expanding ratio is also controlled by a length of shear section, retained austenite stability and so on. The length of shear section tends to increase in low silicon and low manganese steels, as shown in Fig. 10.
4. Discussion

4.1. Relation between Retained Austenite Characteristics and Chemical Composition

Takahashi and Bhadeshia\(^\text{18}\) have proposed for carbide-free bainitic steels that the carbon concentration in retained austenite is equal to one in austenite at \(T_0\) temperature where austenite and ferrite of the same chemical composition have identical free energies. Carbon concentration in austenite at \(T_0\) computed by THERMO-CALC\(^{19}\) is shown in Fig. 11, in which the measured carbon concentration is plotted as a function of silicon and manganese contents. In the figure, manganese content dependence of the measured carbon concentration shows the same tendency as that of calculated one, although their values differ from each other. Thus, it is considered that manganese addition lowers the carbon concentration above 1.0 mass%.

4.2. Optimum Microstructure for Stretch-flangeability

Figure 12 shows relationships between TS×λ values and retained austenite parameters. From this figure, it can be seen that the best stretch-flangeability is completed in the steels containing 2–4 vol% retained austenite with carbon concentration above 1.0 mass%.

According to the previous work about the TMP steels,\(^7\) straining on hole-punching forces to transform the retained austenite to martensite and harden the ferrite matrix in punched surface layer. Simultaneously many voids are formed at the matrix/second phase interface near surface of the break section. Such voids also suppress the development of severe plastic flow and make the growth or propagation easy on hole-expanding. If the steels include an excessive amount of unstable retained austenite, many strain-induced martensites behave themselves like void initiation sites and consequently lower the stretch-flangeability although untransformed retained austenite improves the localized ductility through TRIP effect on hole-expanding. Hence, good stretch-flangeability is considered to be established in steels with 2–4 vol% stable retained austenites.

The stretch-flangeability of the TB steels was controlled by not only retained austenite characteristics but also other metallurgical factors (the amount and size of initial martensite content and matrix structure) and localized ductility.\(^{11}\) As seen in Fig. 13, the TS×λ values of the TB steels austempered at \(T_\alpha\leq M_\delta\) are higher than those of the steels austempered at \(T_\alpha>M_\delta\). The formers hardly possessed no initial blocky martensite particles (Fig. 4). In addition, the TS×λ values of the TB steels are simply related to the length of shear section (ss) on punching, differing from the \(k\) value, as shown in Fig. 14. These facts indicate that the absence of initial blocky martensites, as well as uniform fine bainitic ferrite structure, brings on the long shear section referring to developments of severe plastic flow and large void formation strain on punching and resultantly enhances the stretch-flangeability in the TB steels austempered at temperatures below \(M_\delta\).
5. Conclusions

(1) In the present TB steel an initial volume of retained austenite films along bainitic ferrite lath boundary was increased with increasing silicon and manganese contents, in accompany with the decreased carbon concentration in retained austenite.

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

(3) The excellent stretch-flangeability of the TB steels was considered to be principally owing to uniform fine lath structure matrix and a small amount (2–4 vol%) of 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 initiation and/or propagation, as well as TRIP effect of untransformed retained austenite.

(4) Initial blocky martensites in the TB steels austempered at temperatures higher than $M_S$ resulted in serious surface damage on punching and easy crack propagation on the successive hole-expanding, because they behaved themselves like stress concentration sites or void initiation sites.

REFERENCES


Fig. 13. Correlations between strength–stretch-flangeability balance ($TS\times \lambda$) and (a) reduction of area ($RA$) and (b) elongation for 5 mm gauge length ($EL[5]$) in TB steels.

Fig. 14. Correlations between strength–stretch-flangeability balance ($TS\times \lambda$) and (a) $k$ value and (b) length of shear section ($ss$) for TB steels austempered at 375°C (●) or 450°C (○).