1. Introduction

In the field of low alloy steels, the 1990s have seen the development and characterisation of new formable high-strength steel grades presenting combinations of strength and ductility never reached before.1–6) These improvements are requested by the automotive industry facing the need for safer and lighter cars. The so-called ‘TRIP-assisted multiphase steels’ exhibit an improved balance of strength and ductility thanks to complex multiphase microstructures, consisting of an intercritical ferrite matrix and a dispersion of bainite, martensite and metastable retained austenite that brings about the TRIP effect, i.e. a mechanically-induced martensitic transformation.7)

As represented in Fig. 1, such cold-rolled and annealed multiphase microstructures are generated by a 2-stage heat-treatment. The first stage consists in an intercritical annealing during which controlled volume fractions of austenite and ferrite are formed. This intercritical annealing is then followed by an isothermal stage in the bainite transformation temperature range. During this second stage, partial transformation of the austenite to bainite leads to the stabilisation of the remaining austenite at room temperature.

The stabilisation of austenite at room temperature is due to its carbon enrichment all along the thermal scheme. Firstly, carbon concentrates within austenite during the intercritical annealing. Secondly, the bainite transformation is accompanied by carbon redistribution from bainitic ferrite to the surrounding residual austenite.8) In order to promote the formation of carbon-supersaturated austenite instead of cementite precipitation during the bainitic holding, TRIP-aided steels are conventionally alloyed with fairly large amounts of silicon (from 1 to 2.5 wt%)1–6) that inhibit cementite precipitation during bainite transformation.8,9)

Silicon is the conventional alloying element used for inhibiting cementite precipitation. However, the high silicon levels that are needed are not well suited to the industrial practice of galvanised flat products.10) As a consequence, some studies dealing with other elements can be found in the literature. Krauss et al. presented some results in the literature.11–13)
case of Ni-alloyed steels\textsuperscript{23}; Yue \textit{et al.} studied the effect of molybdenum on hot-rolled TRIP-aided steels\textsuperscript{11,12}; phosphorus is also sometimes mentioned as a suitable element.\textsuperscript{13} Recently, aluminium has also been proposed as a promising alloying element for the TRIP-aided steels.\textsuperscript{13,14} However, hardly anything can be found in the literature about the mechanical properties of Al-alloyed TRIP-aided steels. De Meyer \textit{et al.}\textsuperscript{19} presented recently some results about Si- and Al-alloyed TRIP-aided steels. However, their comparison of the mechanical properties of the two steels is misleading since the bulk carbon content (and thus the volume fraction and stability of retained austenite) was different in their two alloys.

This study aims at enlightening the influence of the heat-treating conditions on the retention of austenite in alloys containing either silicon, or aluminium, or both of these elements. The ensuing TRIP effect and mechanical properties of TRIP-assisted multiphase steels are also characterised. We show that the heat-treatment scheme can lead to various mixtures of retained austenite, bainite and martensite dispersed in a ferritic matrix. Mechanical testing shows that each steel grade brings about unique combinations of strength and ductility. Comparison of Si- and Al-alloyed TRIP-aided steels shows that aluminium (i) allows the retention of austenite; (ii) presents a weaker solid-solution strengthening effect than Si. The mixed Al-Si TRIP-aided steel with a chemical composition that could be more compatible with the industrial practice, presents intermediate properties with respect to the former grades.

2. Experimental Procedure

The chemical compositions of the investigated steels are given in Table 1. The two main alloys studied here, i.e. steels Al1 and Si1, differ only by the substitution of silicon by aluminium. Steel Al–Si is a mixed aluminium–silicon grade, however with a (Al+Si) content lower than the Si or Al contents of steels Si1 and Al1. Finally, steel Al2 presents a reduced aluminium content (0.5 wt%). It is to be noticed that the carbon and manganese contents were kept constant in these different grades.

Cast ingots of each alloy were hot-rolled following classical practice before being 75\% cold-rolled to 1 mm in thickness. Specimens of each steel were intercritically annealed in a fluidised bed furnace and then transferred to a lead bath furnace for the bainitic transformation.\textsuperscript{15} The heat-treatment schemes were also simulated in a dilatometer in order to follow the formation and transformation of the austenite.\textsuperscript{16,17}

Microstructures were studied by light microscopy and scanning electron microscopy (SEM). The sample preparation methods are described elsewhere.\textsuperscript{16–18} Retained austenite contents were measured by Mössbauer spectroscopy and x-ray diffraction (XRD). XRD was also used to determine the lattice parameter (\(a_0\)) of the retained austenite and therefore its carbon content by making use of the following relationship

\[
a_0 (\text{Å}) = 3.578 + 0.033 \text{ (wt\% C)} \quad \text{(1)}
\]

modified as proposed in Ref. 19) to take into account the effect of the substitutional alloying elements on the lattice parameter of austenite.

Loads and elongations measured during uniaxial tensile tests were converted to true stress–true strain curves. Yield strength (\(\sigma_y\)), true stress at maximum load (\(\sigma_{\text{max}}\)) and true uniform strain (\(\varepsilon_u\)) were determined from these tensile curves. Strain hardening was characterised by the incremental strain-hardening exponent (\(n_{\text{incr}}\)) defined as

\[
n_{\text{incr}} = \frac{d \ln \sigma}{d \ln \varepsilon} \quad \text{(2)}
\]

3. Results and Discussion

3.1. Heat-treatment and Microstructure

Figure 2 presents the transformation map of steel Si1 intercritically annealed at 750°C (i.e. 25°C above Ac1) for 4 minutes and bainitically transformed at 375°C. This type of map gives the evolution of the phases constituting the room temperature microstructure of steel Si1 as a function of the bainitic holding time at 75°C. Steel Si1 has been previously intercritically annealed at 750°C for 4 min.

Table 1. Chemical compositions (10\textsuperscript{-3} wt\%) of the investigated steels.

<table>
<thead>
<tr>
<th>Steel</th>
<th>C (wt%)</th>
<th>Mn (wt%)</th>
<th>Si (wt%)</th>
<th>Al (wt%)</th>
<th>P (wt%)</th>
<th>S (wt%)</th>
<th>N (wt%)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Al1</td>
<td>110</td>
<td>1550</td>
<td>59</td>
<td>1500</td>
<td>12</td>
<td>7</td>
<td>17</td>
</tr>
<tr>
<td>Si1</td>
<td>110</td>
<td>1530</td>
<td>1500</td>
<td>43</td>
<td>8</td>
<td>6</td>
<td>3.5</td>
</tr>
<tr>
<td>Al-Si</td>
<td>115</td>
<td>1510</td>
<td>490</td>
<td>380</td>
<td>3</td>
<td>9</td>
<td>30</td>
</tr>
<tr>
<td>Al2</td>
<td>128</td>
<td>1560</td>
<td>5</td>
<td>515</td>
<td>1</td>
<td>27</td>
<td>9</td>
</tr>
</tbody>
</table>

Fig. 2. Transformation map of steel Si1 giving the evolution of the room temperature microstructure with the bainitic holding time at 375°C. Steel Si1 has been previously intercritically annealed at 750°C for 4 min.

Figure 3 illustrates these different microstructures with micrographs corresponding to bainitic holdings of 3

1069 © 2001 ISIJ
min and 15 min, respectively. These micrographs show that the microstructure consists in a matrix of ferritic grains with a dispersion of martensite (Fig. 3a), bainite and retained austenite grains located at the grain boundaries.

Figure 2 also shows that up to 10% of austenite can be retained in steel Si1. Finally, it is worth noting that the bainite transformation stops before the complete consumption of the austenite. This incomplete reaction phenomenon, related to the carbon enrichment of residual austenite and to the diffusionless nature of the bainite transformation, will be discussed later.

For comparison, the same transformation map has been established for steel Al1 intercritically annealed at 750°C and bainitically transformed at 375°C (i.e. the same conditions as steel Si1). The results are summarised in Fig. 4 whereas Fig. 5 presents typical micrographs of the room temperature microstructure after bainitic holding of 1 min (Fig. 5(a)) and 5 min (Fig. 5(b)). The behaviour of the Al-alloyed TRIP-aided steel is similar to the behaviour of the Si-alloyed steel. Intercritical annealing gives rise to a ferrite–austenite microstructure containing 30% of austenite. During the bainitic holding, this austenite partially transforms to bainite, leading to the austenite stabilisation at room temperature in the same way as for steel Si1. Martensite can be found in the microstructure when the bainitic holding is short (holding times shorter than 2 min) (Fig. 5(a)). However, comparison of Figs. 2 and 4 shows that the transformation kinetics are faster for steel Al1. For longer holding times, up to 8% of austenite can be retained at room temperature, which is a slightly smaller amount than for steel Si1. Figure 4 also shows that after the very
fast bainite transformation in the early beginning of the holding, the austenite amount keeps progressively decreasing due to a slight carbide precipitation.\(^{13}\) However, aluminium seems to be an effective element for the retention of austenite in TRIP-assisted multiphase steels.

Finally, the transformation map of steel Al–Si, intercritically annealed at 750°C and bainitically transformed at 375°C, is given in Fig. 6. The initial volume fraction of intercritical austenite is 35%. Figure 6 shows that the kinetics of the bainite transformation and of the austenite stabilisation are quite comparable to that of steel Al1 (and thus faster than for steel Si1). The maximum amount of austenite retained at room temperature is 9\%, i.e., comparable to steels Al1 and Si1. It should be noticed that such a large amount of austenite can be retained with a reduced content of stabilising alloying elements (Al and Si). However, Fig. 6 shows that the continuation of the bainitic holding of steel Al–Si brings about a monotonous decrease of the retained austenite content at the expense of bainite. No incomplete reaction phenomenon is thus observed for steel Al–Si.

Figure 7 presents the evolution of the volume fraction of retained austenite as a function of bainitic holding time at 375°C for steels Al1, Si1 and Al–Si. As already shown in Figs. 2 and 4, steels Al1 and Si1, which differ only by the substitution of silicon by aluminium, present similar behaviours during the bainitic holding. In both cases, the volume fraction of retained austenite increases in the first steps of the bainite transformation, to reach a maximum that is maintained for longer holding times. Aluminium seems to be only slightly less effective for the retention of austenite. The behaviour is quite different for steel Al–Si. Even if short bainitic holding times lead to the increase of the retained austenite content up to a maximum of 9\%, this maximum is not maintained for longer holding times.

Since a large part of the effectiveness of the TRIP effect relies on the mechanical stability of the retained austenite and thus on its carbon enrichment,\(^{23}\) the evolution of the carbon content of austenite as a function of the bainitic holding time was measured for each steel. The results are summarised in Fig. 8. Once again, we can see that conventional Si-alloyed and newly developed Al-alloyed TRIP-aided steels present the same behaviour with respect to carbon enrichment. The retained austenite of steels Si1 and Al1 presents an increasing carbon content when the bainite transformation further proceeds up to a maximum level quite similar in the two steels. Steel Al–Si also presents a similar behaviour. The carbon content increases up to a level (0.97 wt\%) only slightly smaller than for steels Si1 and Al1 (~1.03 wt\%). This level is maintained for quite a long time before it decreases.

The present results show that 1.5 wt\% Al TRIP-aided steel behaves quite similarly to conventional 1.5 wt\% Si steels from the point of view of the heat-treatment (bainite transformation) and of the resulting microstructures. A mixed Al–Si TRIP-aided steel with a reduced amount of stabilising elements also presents a comparable behaviour even if large amounts of austenite cannot be retained during long bainitic holding times. A more precise control of the bainite transformation conditions is thus required for this steel in order to generate the optimised microstructures and ensuing properties.

The characterisation of bainite transformation and stabil-
isation of intercritical austenite (i.e., fine austenite grains dispersed in a ferritic matrix) in Si-alloyed TRIP-aided steels has been reported recently.\textsuperscript{16,17} As already mentioned and shown here, steels Si1 and Al1 exhibit an incomplete reaction phenomenon during bainite transformation (even though a slight carbide precipitation occurs in steel Al1 for long holding times). This phenomenon corresponds to a maximum carbon enrichment of austenite.\textsuperscript{16,17,20} Furthermore, this maximum carbon content is dictated by the theoretical limit given by the $T_0$-curve (i.e., the upper bound of the austenite carbon content allowing a diffusionless transformation of austenite to ferrite).\textsuperscript{13,16,17} The behaviour of these two steels can thus be explained by considering (i) the displacive mechanism of bainite formation, (ii) the carbon partitioning between bainitic ferrite and residual austenite and (iii) the inhibition of cementite precipitation from austenite.\textsuperscript{16,17,21} In the present case, Fig. 2 to 8 show that 1.5 wt\% aluminium is almost as effective as 1.5 wt\% silicon for allowing the retention of large amounts of carbon-enriched retained austenite at room temperature with only minor cementite precipitation. Comparable volume fraction of C-enriched austenite can also be retained in the mixed Al–Si steel grade. However, the lower amount of alloying elements in this steel does not allow a complete hindering of the cementite precipitation so that a too long bainitic holding leads to the complete transformation of austenite into bainite.

### 3.2. Mechanical Properties

Figure 9 presents the values of true uniform strain vs. true stress at maximum load for all the tested combinations of heat-treatment parameters of the four steels. We can see that each steel grade leads to particular and well-defined levels of strength and ductility.

For the conventional Si-alloyed TRIP-aided steel Si1, true stress at maximum load varies from 800 to 880 MPa, while true uniform strain ranges between 0.18 and 0.23. In comparison, steel Al1 presents globally smaller true stresses at maximum load that range only from 650 to 750 MPa, while true uniform strain can be slightly higher (from 0.19 to 0.25). It is quite surprising that steels Al1 and Si1, which present very similar behaviours during heat-treatment and quite identical microstructures, exhibit such different stress-strain levels at the onset of necking. Steel Al–Si presents intermediate mechanical properties, i.e., a true stress at maximum load ranging from 750 to 850 MPa and a true uniform strain ranging from 0.16 to 0.2. Steel Al2 exhibits the same level of true stress at maximum load as steel Al1 and the same true uniform strain level as steel Al–Si.

In order to better understand the different strength-ductility levels exhibited by each steel grade, one particular sample of each steel grade corresponding to the maximum value of the performance index $\varepsilon_u * \sigma_{TS}$, was considered more carefully. Table 2 summarises the microstructural parameters of these specimens. Specimens of steels Si1, Al1 and Al–Si present quite similar microstructures even though steel Si1 contains a slightly higher initial level of retained austenite. It is also noteworthy that the austenite retention and thus the TRIP effect is very limited for steel Al2.

Figure 10 presents the true stress–true strain curves of these four specimens. This figure illustrates that the specimens of steels Si1 and Al1 exhibit identical true uniform strains. However, as also shown in Fig. 9, they mainly differ by the strengthening of the microstructure. The specimen of steel Al–Si also presents a higher strength than the specimen of steel Al2 while its true uniform strain is slightly larger. The substitution of Al by Si between Al1 and Si1 or the addition of Si between steels Al2 and Al–Si thus lead to higher stresses without compromising the ductility. Furthermore, the higher aluminium content in steel Al1 in comparison with steel Al2 leads to a higher ductility while
the strength levels are identical.

Figure 11 presents, for the same specimens, the strain-hardening behaviour expressed by the incremental work-hardening exponent ($n_{\text{inc}}$—Eq. (2)). Steels Al1 and Si1 on the one hand and steels Al–Si and Al2 on the other hand present almost identical strain-hardening behaviours, except in the early part of plastic straining. However, despite identical work-hardening capabilities, the specimens containing silicon exhibit higher strengths. Indeed, not only the TRIP effect but also the mechanical properties of the other phases influence the strength–ductility balance of the TRIP-aided steels. Indeed, silicon is known as a potent solid-solution strengthening addition for ferrite. Some results for TRIP-aided steels\(^{22}\) show that the strengthening of the ferrite matrix by silicon plays an important role in the improvement of the mechanical properties of Si-alloyed TRIP-aided steels. On the other hand, aluminium is not as effective for solid-solution strengthening. Leslie\(^{23}\) and Pickering et al.\(^{24}\) proposed a strengthening effect by silicon of the order of 100 MPa for 1 wt% of solute while they reported that aluminium exhibits no solid-solution strengthening. This difference perfectly fits with the change of hardening observed when silicon is substituted by aluminium in TRIP-assisted multiphase steels. This influence of the solid-solution strengthening of the ferritic matrix can also be noticed from the fact that the strength exhibited by steel Al1 is not larger than that of steel Al2 whereas the content of silicon plays a large role on the strength level of the TRIP-aided steels.\(^{23}\) The mixed addition of Si and Al in steel Al–Si thus allows a good compromise between the austenite stabilising effect of both of these elements during the bainite transformation and the solid-solution strengthening effect of the ferritic matrix.

4. Conclusion

This study investigates the influence of the heat-treating conditions on the retention of austenite and on the ensuing TRIP effect and mechanical properties of TRIP-aided steels alloyed with either silicon, or aluminium, or both of these elements. In each steel grade, a detailed characterisation of the phase transformations and of the resulting microstructures allows to draw transformation maps giving the volume fractions of the different phases and the carbon enrichment of the retained austenite. Uniaxial tensile testing shows that each steel grade brings about unique and well-defined combinations of strength and ductility.

Comparison of Si- and Al-alloyed TRIP-aided steels shows that aluminium is an effective element for (i) the inhibition of cementite precipitation during bainite transformation, (ii) the retention of austenite at room temperature, and (iii) the optimisation of the TRIP effect during straining. However, we show that the absence of solid-solution strengthening by aluminium, in contrast to silicon which is a potent strengthening element, leads to TRIP-aided steels presenting low strength levels. A mixed Al–Si TRIP-assisted multiphase steel appeared to offer an efficient compromise between the processing practice of these steels, the exhibited mechanical properties and the industrial requirements.

Acknowledgements

This work was supported by the Belgian State, Prime Minister’s Office, Federal Office for Scientific, Technical and Cultural Affairs, under contract P4/33 Inter-University Poles of Attraction Programme. A. Mertens acknowledges the financial support of the FRIA. The work of P. J. Jacques was supported by the Fonds National de la Recherche Scientifique (Belgium) and by a fellowship of the Government of Quebec for a post-doctoral period at McGill University. PJ and FD are also indebted to R&D Cockerill-Sambre (Usinor Group) for continuous support.

REFERENCES


Fig. 11. Curves of the incremental work-hardening exponent ($n_{\text{inc}}$) of the specimen of each steel grade optimising the performance index $e_0^* \sigma_{\text{TS}}$. 

![Graph](Image)