# ADVANCED FORMABLE HIGH- AND ULTRA HIGH-STRENGTH TRIP-AIDED SHEET STEELS

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### Abstract

Many formable high or ultra-high strength TRIP-aided steels have been developed to realize a drastic weight reduction and excellent crash worthiness of vehicles since 1987. In general, formability of the low alloy TRIP-aided steels is principally influenced by chemical composition, heat treatment conditions and forming conditions, which control the retained austenite characteristics and matrix microstructure of the steels. In this study, microstructure, retained austenite characteristics, tensile properties and formability of three types of TRIP-aided steels with different matrix structure, polygonal ferrite, annealed martensite or bainitic ferrite are introduced. In addition, the effects of alloying elements, heat treatment conditions and forming conditions on these properties are discussed.

## Introduction

In order to attain a drastic weight reduction and crash worthiness performance of automobiles, formable high- and ultra high-strength steels with tensile strength of 780-1470 MPa are required for automotive applications such as center pillar reinforcement, seat frame and so on. Transformation-induced plasticity (TRIP) [1] of retained austenite is very useful for improving the formability of these steels. On this basis, high-strength steel containing metastable retained austenite, "TRIP-aided steel", was developed by means of Si addition to low carbon steel and austempering heat treatment after annealing [2-8], as shown in Figure 1. This TRIP-aided steel with polygonal ferrite matrix (TPF steel) attained extremely large total elongation of 30-40% due to strain-induced transformation over a large strain range. However, the stretch-flangeability was maintained to the same low value as dual-phase steel [9,10]. Also, the tensile strength could not exceed 980 MPa, because the matrix structure is soft polygonal ferrite as a result of the intercritical anneal. Furthermore, galvanizing (GI) and/or galvanealing (GA) were difficult because an insoluble Si oxide layer formed due to the Si addition and only a small amount of retained austenite is left at the austempering temperature corresponding to hot-dip galvanizing and galvannealing temperatures (460 °C and 520 °C, respectively).

These weak points corresponding to formability may be overcome by the following means or techniques, which change the morphology and stability of retained austenite, as well as matrix structure (Table I).

- (1) Replacement of matrix structure [4,9-12]: improvement of formability (particularly stretch-flangeability) and increase in strength.
- (2) Warm forming and/or strain rate controlling [4,6,7,9,10]: enhancement of formability.
- (3) Addition of alloying elements [3,9,13-18]: raising the optimum austempering temperature (which is beneficial to GI and/or GA) and improved formability.

In this paper, the effects of the above means on the weak points of TPF steel are introduced, as well as a basic theory of the carbon enrichment mechanism and strain-induced transformation behavior of TRIP-aided steel.



Figure 1. Development history of three kinds of low alloy TRIP-aided steels with different matrix structure, polygonal ferrite (TPF), bainitic ferrite (TBF) or annealed martensite (TAM). DP and B steels are dual-phase and bainitic steels, respectively.

Process		Structural change			
		matrix	retained austenite		
		morphology	morphology	volume fraction	stability
Heat treatment condition	prior structure and annealing conditions	change	change	change	change
	austempering conditions	small change	small change	change	change
Forming condition	forming temperature	no change	no change	no change	change
	forming strain rate	no change	no change	no change	change
Addition of alloying element		small change*	small change	change	change

Table I. Structural change in TRIP-aided steels by some processes

\* carbides except for cementites precipitate in Nb, Ti, Mo, V etc. bearing steels

## Carbon Enrichment Mechanism and Strain-Induced Transformation Mechanism

In general, low alloy TRIP-aided steel is produced by annealing or intercritical annealing after hot rolling or cold rolling, followed by austempering in the bainite transformation temperature range (Figure 2(a)). On intercritical annealing, carbon in austenite phase is enriched to the  $A_{c3}$ line. After that, carbon concentration is progressed to the  $T_0$  or  $T_0$ ' line during austempering, if cementite precipitation is suppressed by addition of Si, Al etc.



Figure 2. Annealing-austempering process and pseudo-binary Fe-C diagram for 1.5Si-1.5%Mn.  $T_{\gamma}$  and  $T_{\alpha+\gamma}$  represent annealing temperatures in  $\gamma$  and  $\alpha+\gamma$  regions, respectively.  $T_A$  and  $t_A$  are austempering temperature and time, respectively.

The volume fraction of retained austenite was quantified from integrated intensity of  $(200)\alpha$ ,  $(211)\alpha$ ,  $(200)\gamma$ ,  $(220)\gamma$  and  $(311)\gamma$  peaks of Mo-K $\alpha$  radiation, as given by the following equation [19].

$$f_{\gamma} = 1 / [Z (P_{\alpha}/P_{\gamma}) + 1]$$
 (1)

where  $P_{\alpha}$  and  $P_{\gamma}$  are X-ray diffraction intensity of  $\alpha$  and  $\gamma$  diffraction planes, respectively. Z is constant given by combination of  $\alpha$  and  $\gamma$  diffraction planes (Table II).

Combination of diffraction plane	Z value
(200)α-(200)γ	2.19
(200)α-(220)γ	1.35
(200)α-(311)γ	1.50
(211)α-(200)γ	1.13
(211)α-(220)γ	0.70
(211)α-(311)γ	0.78

Table II. Z value for combination of diffraction plane [19].

The carbon concentration ( $C_{\gamma}$ , mass%) was estimated by substituting the lattice constant ( $a_{\gamma}$ , 10<sup>-1</sup> nm) measured from (200) $\gamma$ , (220) $\gamma$  and (311) $\gamma$  peaks of Cu-K $\alpha$  radiation into the following equation [20].

$$a_{\gamma} = 3.5780 + 0.0330 \text{ x } (\%C_{\gamma}) + 0.00095 \text{ x } (\%Mn_{\gamma}) - 0.0002 \text{ x } (\%Ni_{\gamma}) + 0.0006 \text{ x } (\%Cr_{\gamma}) + 0.0056 \text{ x } (\%N_{\gamma}) + 0.0028 \text{ x } (\%Al_{\gamma}) - 0.0004 \text{ x } (\%Co_{\gamma}) + 0.0014 \text{ x } (\%Cu_{\gamma}) + 0.0053 \text{ x } (\%Mo_{\gamma}) + 0.0079 \text{ x } (\%Nb_{\gamma}) + 0.0032 \text{ x } (\%Ti_{\gamma}) + 0.0017 \text{ x } (\%V_{\gamma}) + 0.0057 \text{ x } (\%W_{\gamma})$$

$$(2)$$

where  $\%Mn_{\gamma}$ ,  $\%Ni_{\gamma}$ ,  $\%Cr_{\gamma}$ ,  $\%N_{\gamma}$ ,  $\%Al_{\gamma}$ ,  $\%Co_{\gamma}$ ,  $\%Cu_{\gamma}$ ,  $\%Mo_{\gamma}$ ,  $\%Nb_{g}$ ,  $\%Ti_{\gamma}$ ,  $\%V_{\gamma}$  and  $\%W_{\gamma}$  represent solute content (mass%) of the individual alloying elements in retained austenite. For convenience, they were assumed to be equivalent to the added contents.

Retained austenite stability against straining, "mechanical stability", is controlled by carbon concentration, morphology and size of retained austenite, strength of matrix structure, stress

state, temperature and strain rate, etc. For transformation-induced plasticity of low alloy TRIPaided steels, the following three temperature ranges are of importance [18].

- (i)  $M_{\rm S}$   $M_{\rm S}^{\sigma}$  range: Yielding of the austenite is by stress-induced transformation of austenite to martensite at pre-existing nucleation sites. At  $M_{\rm S}^{\sigma}$  the stress needed to initiate the martensitic transformation of the retained austenite equals the yield strength of the parent  $\gamma$  phase. Below this temperature, the retained austenite transforms to martensite via preexisting nucleation sites. As the temperature is increased, the stress needed for the martensitic transformation increases since the chemical driving force decreases.
- (ii)  $M_S^{\sigma}$   $M_d$  range: Above the  $M_S^{\sigma}$  temperature, the austenite is strained. The martensite is now predominantly nucleated at new nucleation sites produced by slip. Note that this martensite is not of the high-C plate type and will not have the brittleness associated with plate type martensite. Yielding of the austenite is by glide. The transformation is mainly strain-induced. Additional nucleation occurs at the intersection of strain-induced deformation bands.
- (iii)  $T > M_d$ : The  $M_d$  temperature is the temperature above which no martensite transformation occurs. The higher temperature results in a higher stacking fault energy and a lower driving force for transformation. Hence, no transformation occurs as a result of straining.



Figure 3. Relationship between stress-assisted and strain-induced martensite [21].  $\Delta$   $G^{\sigma^{=0, Ms}}$  is the energy for transformation of austenite to martensite.

As mentioned above, mechanical stability of retained austenite varies as a function of forming temperature. If the mechanical stability is defined by the following k value [6,7], typical forming temperature dependence of k value is shown in Figure 4.

$$\ln f_{\gamma 0} - \ln f_{\gamma} = k \varepsilon \tag{3}$$

where  $f_{\gamma}$  and  $f_{\gamma 0}$  are volume fraction of retained austenite after strained to  $\varepsilon$  and initial one. Bhadeshia et al. [22] rewrite equation (3) as follows,

$$\ln f_{\gamma 0} / f_{\gamma} = k_1 \Delta G^{\alpha' \gamma} \, \varepsilon \tag{4}$$

where  $\Delta G^{\alpha'\gamma}$  is the chemical free energy change  $\Delta G^{\alpha'\gamma} = \Delta G^{\gamma} - \Delta G^{\alpha'}$  for the transformation of austenite to ferrite of the same composition (without considering stored energy due to the shape deformation).

The k value becomes minimum at about  $T_{\rm S}$ =100-150 °C [7,9,23]. The strain-induced martensite transformation (SIMT) preferentially takes place below  $T_{\rm S}$ . On the other hand, strain-induced

bainite transformation preferentially occurs above  $T_S$ . The lower the strain rate, the higher the k-value [7]. Equi-biaxial deformation raises the k-value, compared to uniaxial deformation [23]. And, Even if matrix structure of TRIP-aided steels is changed, the  $T_S$  temperature responsible to minimum k value is nearly constant, although these k values are different each other. In TRIP steels such as Fe-Cr-Ni [1,24] and Fe-Ni-C [24] metastable austenitic stainless steels, the k value becomes zero above  $M_d$  temperature.



Figure 4. Typical variations in volume fraction of retained austenite ( $f_{\gamma}$ ) as a function of plastic strain (0.4%C-1.5%Si-1.5%Mn TPF steel) [7] and illustration of forming temperature dependence of k value of TRIP-aided steel [7,9,23]. SIMT: strain-induced martensite transformation, SIBT: strain-induced bainite transformation.



Figure 5. Structural changes of 0.4%C-1.5%Si-1.5%Mn TPF steel after strained up to 10% deformation at (a) 27 °C, (b) 150 °C or (c) 350 °C and at strain rate of  $2.8 \times 10^4$ /s [7]. (a): strain-induced twinned martensite transformation, (b): deformation twin, (c): strain-induced bainite transformation (strain-induced decomposition).

## Formability

## Replacement of matrix structure

By replacing the prior structure of ferrite and pearlite with a martensite or bainite structure, a new TRIP-aided steel with annealed martensite matrix structure (TAM steel) can be produced by austempering treatment after intercritical annealing [4,10,12]. On the other hand, if annealing is conducted in  $\gamma$  region, followed by immediately austempering, other TRIP-aided steel with bainitic ferrite matrix structure (TBF steel) can be fabricated [9,11,13]. These typical microstructures are shown in Figures 6 and 7. In both steels, retained austenite morphology changes to needle or film type. In addition, these matrix structures are characterized by uniform and fine lath structure and small inter particle path of retained austenite, although dislocation density of the matrix is different each other (Fig. 7). The microstructure of this TBF steel

resembles steel subjected to quenching and partitioning (Q&P) process proposed by Speer et al. [25].



Figure 6. Typical light micrographs of (a) 0.2%C-1.5%Si-1.5%Mn TPF ( $t_A$ =1000s), (b) TAM ( $t_A$ =100s) and (c) TBF ( $t_A$ =200s) steels austempered at 400 °C, in which gray, black and white regions represent matrix ((a) polygonal ferrite, (b) annealed martensite, (c) bainitic ferrite), bainite and retained austenite and/or martensite, respectively.



Figure 7. Typical transmission electron micrographs of (a) 0.2%C-1.5%Si-1.5%Mn TAM and (b) TBF steels austempered at 400 °C for 100s and 450 °C for 200s, respectively.  $\alpha_{am}$ ,  $\alpha_{bf}$  and  $\gamma_R$  represent annealed martensite, bainitic ferrite and retained austenite, respectively.

Figure 8 compares retained austenite characteristics of three kinds of TRIP-aided steels with a different matrix structure. TAM steel is characterized by higher volume fraction and carbon concentration of retained austenite. On the other hand, volume fraction of retained austenite of TBF steel is nearly the same as TPF steel or decreases a bit, although its carbon concentration is as high as TAM steel. The carbon concentrations of TAM and TBF steels nearly agree with  $T_0$  line at which ferrite and austenite of the same chemical composition have identical chemical free energy. So, this may be caused in that these steels do not contain cementite. In general, optimum austempering time ( $t_A$ ) of TAM and TBF steels are shorter than TPF steel, despite higher carbon concentration of retained austenite. This is associated with martensite structure (TAM steel) and austenite structure (TBF steel) containing a large amount of solute carbon, which brings on rapid carbon diffusion on austempering.

Figure 9 shows typical flow curves and yield ratio of three types of TRIP-aided steels (0.1-0.6%C-1.5%Si-1.5%Mn). Figure 10 shows total elongation and stretch-flangeability of the steels (heat treatment conditions are shown in Ref. [12]). TAM steel is characterized by extremely

large total elongation and good stretch-flangeability and bendability in a tensile strength range of 800-1000 MPa. And, the yield ratio increases to 0.7 due to higher strength of matrix. On the other hand, TBF steel possesses the best stretch-flangeability. The excellent stretch-flangeability may be associated with uniform fine lath structure, a small strength ratio of second phase to matrix and relatively stable retained austenite. It is noteworthy that only TBF steel achieves ultra high strength of 1000-1500 MPa due to harder bainitic ferrite and/or martensite matrix. Moreover, the TBF steels possess high yield ratios of about 0.8.



Figure 8. Variations in (a) initial volume fraction  $(f_{\gamma 0})$  and (b) initial carbon concentration  $(C_{\gamma 0})$  of retained austenite as a function of carbon content in 0.1-0.6%C-1.5%Si-1.5%Mn TRIP-aided steels.



Figure 9. Comparison of (a) nominal stress-strain curves of 0.2%C-1.5%Si-1.5%Mn TAM, TBF and TDP steels and variations in (b) yield ratio (*YR*) as a function of tensile strength (*TS*) in 0.1-0.6%C-1.5%Si-1.5%Mn TRIP-aided steels.



Figure 10. Variations in (a) total elongation (*TEl*) and (b) hole expanding ratio ( $\lambda$ ) as a function of tensile strength (*TS*) in 0.1-0.6%C-1.5%Si-1.5%Mn TRIP-aided steels.

#### Warm forming (and strain rate controlling)

As mentioned previously, mechanical stability of retained austenite varies with forming temperature and strain rate [23,26,27] (Figure 4). Figures 11 and 12 show changes in typical flow curves and the effects of forming temperature and strain rate on tensile properties of 0.4%C-1.5%Si-1.5%Mn TPF steel [7]. At 150 °C, total elongation becomes maximum with serration and tensile strength becomes minimum. In this case, *k* value becomes minimum at about 150 °C, although it decreases with increasing strain rate. However, this behavior is not clear at high strain rate. According to Sugimoto et al [28], very high internal stress develops when TPF steel is strained at 150 °C. And, this is explained by a large difference in flow stress between matrix and second phase and suppression of strain-induced transformation. So, extreme large total elongation at 150 °C of TPF steel may be associated with stabilization of retained austenite (effective strain-induced transformation in a large strain range) and the resultant high long-range internal stress due to difference in flow stress between soft matrix and hard-retained austenite.

Figure 13 shows tensile properties and forming temperature dependence of formabilities for 0.2%C-1.5%Si-1.5%Mn TPF and TBF steels [29]. In TPF steel, limiting drawing ratio (*LDR*) and hole-expanding ratio ( $\lambda$ ) are improved by warm forming in the same way as total elongation. However, these forming temperature dependences are different from that of TEL and the peak values appear at temperatures somewhat higher than 150 °C. On the other hand, in TBF steels peak temperatures of total elongation and formability are about 0-100 °C, although tensile strength becomes minimum at about 100 °C, similar to TPF steel. It is noted that TBF steel austempered at 375 °C possesses higher hole-expanding ratio than steel austempered at 450 °C, although the steel has higher tensile strength. This is caused by finer lath structure and higher retained austenite stability. If hole expanding is punched at 150 °C, followed by hole expanding at room temperature, the hole-expanding ratio is considerably improved because strain-induced transformation and void initiation on punching is effectively suppressed and the TRIP effect efficiently operates on hole-expanding. Also, TAM steel shows the same tendency as TBF steel.



Figure 11. Changes in flow curves with testing temperature and strain rate in 0.4%C-1.5%Si-1.5%Mn TPF steel ( $f_{\gamma 0}$ =14.7vol%,  $C_{\gamma 0}$ =1.20mass%)



Figure 12. Variations in tensile strength and total elongation with testing temperature and strain rate in 0.4%C-1.5%Si-1.5%Mn TPF steel.



Figure 13: Forming temperature dependences of (a) tensile strength (*TS*), (b) total elongation (*TEl*), (c) limiting drawing ratio (*LDR*) and (d) hole-expanding ratio ( $\lambda$ ) of 0.2%C-1.5Si-1.5%Mn TPF and TBF steels austempered at *T*<sub>A</sub>=375-450 °C. Strain rate of tensile tests is 2.8x10<sup>-4</sup>/s.

TPF steel ( $T_A$ =400 °C):  $f_{\gamma 0}$ =18.2mass%,  $C_{\gamma 0}$ =1.00mass%, TBF steel ( $T_A$ =375 °C):  $f_{\gamma 0}$ =9.8mass%,  $C_{\gamma 0}$ =1.16mass%, TBF steel ( $T_A$ =450 °C):  $f_{\gamma 0}$ =11.2mass%,  $C_{\gamma 0}$ =0.96mass%.

#### **Alloying Elements**

Many researchers reported about the effects of alloying elements such as C, Al, P, Si, Mn, etc. on tensile strength and total elongation of TPF steels [3,13-17,27, 30-39]. In addition, the effects of austempering temperature and holding time on these properties have been examined because good strength – elongation balance is required at 460 °C corresponding to hot-dip galvanizing [6,7,9,11,13,29]. Unfortunately the optimum austempering temperatures were between 400 and 450 °C in these studies. It is very important that TAM and TBF steels are beneficial to raise the optimum austempering temperature because these carbon concentrations of retained austenite are relatively high even at 460 °C due to carbide free [28,40]. So, the effects of alloying elements on strength – elongation balance and optimum austempering temperature were investigated for TAM and TBF steels, as well as mutual effect of alloying element and warm forming of TPF steel.

#### TPF steel

Figure 14 shows flow curves of Al-Nb-Mo bearing TPF steel (0.4%C-0.5%Si-1.5%Mn-1.0%Al-0.05%Nb-0.2%Mo) [41]. At 20 °C testing, the total elongation are about 20%, but it increases to 78% at 150 °C and strain rate of  $3.3\times10^{-3}$ /s under tensile strength of 1000 MPa. The total elongation considerably exceeds that of 0.4%C-1.5%Si-1.5%Mn base steel (Figure 12). It is supposed that this reason is associated with complex additions of alloying elements. It is noteworthy that this strength – elongation balance is at the same level as that of Fe-Mn-Al-Si TWIP steel [42] (Figure 15).

In Figure 14, while fine serrations were observed during low strain rate tests  $(3.3 \times 10^{-4}/\text{s} \text{ and} 3.3 \times 10^{-5}/\text{s})$  at 25 °C, the fluctuations are much more prominent at 150 °C, with intense serrations near the yield point. Load drops continue even up to peak load. Deformation bands are observed in all steel samples tested at strain rates of  $3.3 \times 10^{-4}/\text{s}$  and  $3.3 \times 10^{-5}/\text{s}$  at 150 °C. The bands originate from one end of the gage length and are oriented at an angle of ~45° to the tensile axis and then at higher strains two sets of mutually perpendicular shear bands are observed.



Figure 14. Nominal stress – strain curves of 0.4%C-0.5%Si-1.5%Mn-1.0%Al-0.02%Nb-0.1%Mo TPF steel tested at (a) 25 or (b) 150 °C and at strain rate from  $3.3 \times 10^{-2}$  to  $3.3 \times 10^{-5}$ /s. ( $f_{\gamma 0}$ =25vol%,  $C_{\gamma 0}$ =0.77mass%)



Figure 15. Tensile strength – total elongation balance of high- and ultra high strength steels.

## TAM steel

According to Aaronson [43], annealed martensite lath structure is maintained for II and III groups in the following three groups, because alloying elements of II and III groups promote austenite nucleation along prior martensite lath boundary during intercritical annealing.

I: Fe-C, Fe-C-Mo, Fe-C-Cr II: Fe-C-Mn, Fe-C-Ni, Fe-C-Cu III: Fe-C-Si, Fe-C-Al, Fe-C-Co

Addition of 0.5-1.0%Al (and simultaneous Si removal of the same amount) to 0.2%C-1.5%Si-1.5%Mn TAM steel, brings on a large total elongation and high optimum austempering temperature for total elongation of 450-475 °C, as well as increase in stretch-flangeability and decrease in tensile strength [30](Figure 16). As shown in Figures 17 and 18, Al addition refines annealed martensite lath structure and retained austenite needles. Furthermore, Al increases carbon concentration of retained austenite even at  $T_A$ =500 °C where there is no retained austenite in TPF steel. In general, good galvanization is attained by decreasing the Si content because formation of insoluble silicon oxide layer is suppressed. Unfortunately, we found that Nb addition is not effective to improve formability in TAM steel.



Figure 16. Variations in (a) total elongation (*TEl*) and (b) hole-expanding ratio ( $\lambda$ ) as a function of austempering temperature ( $T_A$ ) in 0.2%C-(0.5-1.5)%Si-1.5%Mn-(0-1.0)%Al TAM steels.



Figure 17. Light micrographs of (a) 0.2%C-1.5%Si-1.5%Mn-0.04%Al and (b) 0.2%C-0.5%Si-1.5%Mn-1.0%Al TAM steels austempered at 475 °C.



Figure 18. Variations in (a) initial volume fraction  $(f_{\gamma 0})$  and (b) initial carbon concentration  $(C_{\gamma 0})$  as a function of austempering temperature  $(T_A)$  in 0.2%C-(0.5-1.5)%Si-1.5%Mn-(0-1.0)%Al TAM steel.

## TBF steel

In TBF steel, Al addition is not effective to improve total elongation and raise the optimum austempering temperature. However, further complex additions of 0.05%Nb and 0.2%Mo are very effective to attain large total elongation and high optimum austempering temperature (450-500 °C), although stretch-flangeability is not improved by these additions, as shown in Figure 19. This is mainly owing to the change of morphology of matrix structure and retained austenite. The typical example of microstructure is shown in Figure 20. In the figure, Al addition makes fine lath structure and retained austenite films. On the other hand, further additions of Nb and Mo change lath structure and retained austenite into quasi-ferrite and blocky phase, respectively.

Complex additions of 0.02-0.08%Nb and 0.1-0.2%Mo increase yield stress and tensile strength by 100-200 MPa due to fine NbMo(C,N) precipitates (Figure 21). According to Drewes and Walker [44], Nb plays the following roles in conventional TRIP-aided steels (TPF steel).

- (1) Nb in solid solution lowers the martensite start temperature. Furthermore, in dissolved form it suppresses carbide precipitation in the temperature range of bainite formation, with consequent increase in the residual austenite content and its carbon content.
- (2) In the precipitated form as carbides, nitrides or carbonitrides, Nb has a grain refining effect, increasing the tensile strength. Further more, niobium suppresses the formation of undesired pearlite in the TRIP-aided steels.

So, high retained austenite volume fraction of this Al-Nb-Mo bearing TBF steels may be caused by Nb in dissolved form, which suppresses cementite precipitation in the temperature range of bainite formation. Nb in precipitated form as Nb(C,N) contributes to refinement of retained austenite phase through refinement of prior austenitic grain. It is considered that the Nb(C,N) changes the retained austenite morphology from filmy to blocky. Consequently, this may increase the volume fraction of retained austenite by suppressing formation of bainite with cementite, and thus maintains high carbon concentration of retained austenite in an austempering temperature range between 450 and 500  $^{\circ}$ C.



Figure 19. Variations in (a) total elongation (*TEl*) and (b) hole-expanding ratio ( $\lambda$ ) as a function of austempering temperature ( $T_A$ ) in (a) base (0.2%C-1.5%Si-1.5%Mn), (b) 0.5Si-1.0Al and (c) 0.5Si-1.0Al-0.05Nb-0.2Mo TBF steels.

Recently, we found that Nb addition and complex additions of Nb and Mo without Al considerably enhances stretch-flangeability at low austempering temperature range, namely in ultra high tensile strength range [45]. This relationship is under consideration.

Figure 22 shows the contribution of three techniques to total elongation and stretch-flangeability of three kinds of TRIP-aided steels. It is found that warm forming is the best effective to improve total elongation. On the other hand, replacement of matrix structure extremely increases stretch-flangeability. Recently, the hot stamping technique was tried for 1470 MPa grade impact members. However, formable TBF steel does not require heating during stamping and maintains a clean surface. In addition, the steel possesses high delayed fracture strength because retained austenite films trap most of the solute hydrogen [46,47]. It is planned that 1470 MPa grade TBF steel may be supplied to automobile makers up to the end of this year.



Figure 20. Typical scanning electron micrographs of steels (a) base (0.2%C-1.5%Si-1.5%Mn), (b) 0.2%C-0.5Si-1.5%Mn-1.0Al and (c) 0.2%C-0.5Si-1.5%Mn-1.0Al-0.05Nb-0.2Mo TBF steels austempered at  $T_A$ =450 °C.



Figure 21. Transmission electron micrographs of 0.2%C-1.5%Si-1.5%Mn-0.05%Nb-0.2%Mo TBF steel austempered at  $T_A$ =475 °C for  $t_A$ =200s.



Figure 22. Contribution of three improvement techniques to (a) total elongation (*TEl*) and (b) stretch-flangeability ( $\lambda$ ) of TRIP-aided steel.

#### Summary

It was introduced that replacement of the matrix structure, warm forming (& strain rate controlling) and the addition of alloying elements were effective in improving the formability of TRIP-aided steels. In regards to the replacement of the matrix structure, TAM steel achieved the best formability balance in a tensile strength range of 780-980 MPa. On the other hand, TBF steel was characterized by the best stretch-flangeability in a tensile strength range of 980-1470 MPa. Warm forming and strain rate controlling brought on excellent strength – elongation balance (tensile strength of 1000 MPa and total elongation of 78%) in TPF steel when alloying elements of Al, Nb etc. were added.

The following interesting results have been reported by authors [46-51]:

- (1) TBF steel possessed superior hydrogen embrittlement resistance in a tensile strength range from 980 to 1470 MPa [46,47], even after forming.
- (2) Also, TBF steel exhibited large cyclic hardening [48,49], high fatigue strength [50] and Charpy absorbed energy [51].

Thus, it is expected that the TBF steels are applied to not only impact members and under body parts, but also driving parts and bolts with ultra high-tensile strength in future.

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