

Review Effects of Partial Replacement of Si by Al on Cold Formability in Two Groups of Low-Carbon Third-Generation Advanced High-Strength Steel Sheet: A Review

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Abstract: Partial replacement of Si by Al improves the coatability (or galvanizing property) of Si-Mn advanced high-strength steel (AHSS) sheets. In this paper, the effects of the partial replacement on the microstructure, tensile property, and cold formability are reported for the low-carbon third-generation AHSS sheets, which are classified into two groups, "Group I" and "Group II". The partial replacement by 1.2 mass% Al increases the carbon concentration or mechanical stability of retained austenite and decreases its volume fraction in the AHSSs, compared to Al-free AHSSs. The partial replacement deteriorates the tensile ductility and stretch formability in the AHSSs with a tensile strength above 1.2 GPa. On the other hand, it achieves the same excellent stretch-flangeability as Al-free AHSSs. A complex addition of Al and Nb/Mo further enhances the stretch-flangeability. The cold formabilities are related to the heat treatment condition and microstructural and tensile properties, and the stress state.

Keywords: third-generation advanced high-strength steel (AHSS); Al addition; microstructure; retained austenite; tensile property; cold formability



Since the 1970s, the first-, second-, and third-generation advanced high-strength steel (AHSS) sheets have been developed for weight reduction and enhancement of collision safety of automobiles [1–6]. In these AHSSs, the performance of the mechanical properties are principally enhanced by transformation-induced plasticity (TRIP) [7] and/or twinning-induced plasticity (TWIP) [8] of metastable retained austenite (γ_R), reverted austenite, and/or austenite. The representative characteristic is shown by the product of tensile strength and total elongation (TS×TEl), which increases with an increase in the initial volume fraction of austenite or retained austenite ($f\gamma_0$) (Figure 1) [6]. For a TS level higher than 1.0 GPa, good performance and low production costs can be obtained by using the third-generation AHSSs.

The third-generation AHSSs are classified into "Group I" and "Group II" by a kind of matrix structure and TS level [1,3–6].

Group I: TRIP-aided bainitic ferrite (TBF) steel [9–18]; one-step and two-step quenching and partitioning (Q&P) steels [19–26]; carbide-free bainitic (CFB) steel [27–34]; and duplex type [35–42], laminate type [43,44], bainitic ferrite-type [45], and Q&P-type [46–50] medium-manganese (D-MMn, L-MMn, BF-MMn, and Q&P-MMn) steels.

Group II: TRIP-aided martensitic (TM) steel [15,51–56] and martensite-type mediummanganese (M-MMn) steel [50,57–59].

In Group I, the kind of matrix structure is bainitic ferrite (or bainitic ferrite/martensite mixture) and the TS level is higher than 1.0 GPa, except for D–MMn and L-MMn steels with the duplex structure of an annealed martensite and reverted austenite [35–42] and laminate structure of the δ -ferrite and complex structure of α -ferrite plus reverted austenite [43,44], respectively. On the other hand, the main matrix structure of Group II is martensite, and its TS level is higher than 1.5 GPa [15,51–59]. In Group II, martensite–austenite (MA)



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Copyright: © 2022 by the author. Licensee MDPI, Basel, Switzerland. This article is an open access article distributed under the terms and conditions of the Creative Commons Attribution (CC BY) license (https:// creativecommons.org/licenses/by/ 4.0/). constituent or phase plays an important role in the strain-hardening behavior [52]. The ultimate goals of both group sheets of steel are to achieve the TS×TEl above 30 GPa% and excellent cold- and warm formability. Due to its excellent nature, the heat treatment process of the third-generation AHSSs is also applied to hot and warm stamping products [60–65].



Figure 1. Relationship between the product of tensile strength and total elongation (TS×TEI) and an initial volume fraction for retained austenite, reverted austenite, or austenite ($f\gamma_0$) in the first-, second-, and third-generation (Group I and Group II) advanced high-strength steel (AHSS) sheets. Q&T: quenching and tempering martensitic steel, DP: dual-phase steel, CP: complex-phase steel, TPF, TAM, TBF, and TM: transformation-induced plasticity (TRIP)-aided steels with polygonal ferrite, annealed martensite, bainitic ferrite, and martensite matrix structure, respectively. Q&P: one-step and two-step quenching and partitioning steels, CFB: carbide-free bainitic steel, D-MMn: duplex type medium-Mn steel, L-MMn: laminate type medium-Mn steel, BF-MMn: bainitic ferrite-type medium-Mn steel, Q&P-MMn: Q&P-type medium-Mn steel, M-MMn: martensite-type medium-Mn steel, HMn TWIP: high-manganese TWIP steel, Aus: austenitic steel. This figure is reproduced based on references [6,37].

The TS×TEl of the AHSSs is controlled by various alloying elements such as C, Si, Al, Mn, Cr, Mo, Ni, B, and P [1,5,11–14,21,22,24,26,28,31–36,38,39,42–49,53–56,59], and the heattreatment process [1,4,9–17,19,20,22–39,42–54,56–59]. In this case, low-carbon content below 0.25 mass% is preferred to keep high weldability. The addition of Si and/or Al suppresses the carbide formation and resultantly increases the volume fraction of metastable γ_R during the heat treatment, in the same way as P [66,67]. As Al does not deteriorate the coatability (or galvanizing property), unlike Si [48,49,68–70], it becomes especially advantageous for industrial production in conventional galvanizing lines. However, Al is a weak solidsolution-strengthening element in steel. In addition, the content is limited because Al is a ferrite-stabilizing element [69].

To promote the application of galvanized AHSS sheets to automotive parts, many researchers investigate the effect of partial replacement of Si by Al on the microstructural and mechanical properties, such as tensile property and formability in low-carbon TBF [12–14], Q&P [21,22,24], CFB [28,31,32,34], D-MMn [38,42], BF-MMn [45], Q&P-MMn [47,49], and TM [55] steels, in the same manner as the first-generation AHSSs, such as TRIP-aided polygonal ferrite (TPF) [70–78] and TRIP-aided annealed martensite (TAM) steels [75,78] (Table 1). Unfortunately, most of the mechanical properties are focused on tensile properties, not formability. In this paper, the influences of the partial replacement of Si by Al on the cold formability, such as stretch formability, stretch-flangeability, and bendability of two groups of low-carbon third-generation AHSSs, are summarized, along with the microstructural and tensile properties. For ease of understanding, these characteristics are stated separately for low-carbon "Si-Mn" and "Si/Al-Mn" third-generation AHSSs. Unfortunately, there is not any research on deep drawability in the third-generation AHSSs, except for the first-generation AHSSs such as TPF [79,80] and TAM [79] steels. Thus, the deep drawability of the third-generation AHSSs is omitted in this review.

| Gen. | Steel | Chemical Composition | Property | Ref. |
|-------------|---------|---|------------|------|
| | TPF | 0.25C-1.28Si-1.67Mn-0.03Al, 0.18C-0.02Si-1.56Mn-1.73Al | 1 | [70] |
| | | 0.21C-2.10Si-1.52Mn-0.022Al, 0.22C-0.01Si-1.49Mn-2.02Al | 1, 2 | [71] |
| | | 0.21C-2.10Si-1.52Mn-0.022Al, 0.22C-0.01Si-1.49Mn-2.02Al | 1, 2, 4 | [72] |
| | | 0.19C-1.46Si-1.57Mn-0.06Al, 0.31C-0.34Si-1.57Mn-1.23Al | 1, 2 | [73] |
| 1st Gen. | | (0.14-0.21)C-(0.34-1.47)Si-1.5Mn-(0.03-0.99)Al | 1, 2 | [74] |
| | | 0.20C-(0.49-1.54)Si-1.5Mn-(0.04-0.99)Al | 1, 2, 4, 5 | [75] |
| | | (0.19-0.25)C-(0.09-1.45)Si-1.7Mn-(0.03-1.49)Al | 1 | [76] |
| | | 0.20C-1.87Si-1.99Mn-(0.04-2.0)Al | 1 | [77] |
| | | 0.20C-(0.49-1.50)Si-1.5Mn-(0.04-0.99)Al | 1, 2, 4 | [78] |
| | TAM | 0.20C-(0.49-1.54)Si-1.5Mn-(0.04-0.99)Al | 1, 2, 4, 5 | [75] |
| | | 0.20C-(0.48-1.50)Si-1.5Mn-(0.04-0.99)Al | 1, 2, 4 | [78] |
| | TBF | 0.20C-(0.49-1.54)Si-(1.48-1.51)Mn-(0.04-0.99)Al-(0-0.05)Nb-(0- | 1.0.4 | [10] |
| | | 0.20)Mo | 1, 2, 4 | [12] |
| | | 0.20C-(0.49-1.51)Si-(1.51-2.51)Mn-(0.04-0.99)Al | 1, 2 | [13] |
| | | 0.20C-(0.99-1.54)Si-1.5Mn-(0.04-0.49)Al-(0-0.05)Nb | 1, 2, 4 | [14] |
| | Q&P | 0.24C-1.45Si-1.61Mn-0.30Al, 0.25C-0.55Si-1.70Mn-0.69Al | 1,2 | [21] |
| | | 0.24C-0.12Si-1.60Mn-1.41Al-0.17Mo | 2 | [22] |
| | | 0.30C-(0.48-0.99)Si-(1.86-2.00)Mn-(0.01-1.10)Al-(1.01-2.20)Cr | 1 | [24] |
| | CFB | 0.25C-(0.08-1.09)Si-2.07Mn-(0.021-1.54)Al | 1, 2 | [28] |
| | | 0.25C-2.1Mn-(0.02-1.54)Al | 1,2 | [31] |
| 3rd Gen. | | 0.22C-(1.79-1.82)Si-(1.98-2.04)Mn-(0-0.50)Al-1.0Cr-0.23Mo | 1, 2 | [32] |
| | | 0.2C-1.55Si-2.0Mn, 0.2C-0.77Si-2.0Mn-0.76Al | 1,2 | [34] |
| | D-MMn | (0.1-0.3)C-(0-1.5)Si-(2-5)Mn-(0-1.5)Al-(0-1.5)Cr | 1 | [38] |
| | | (0.18-0.19)C-(7.66-7.93)Mn-(0-2.79)Al | 1, 2 | [42] |
| | BF-MMn | 0.18C-0.23Si-3.6Mn-1.7Al-0.2Mo-0.04Nb | 1 | [45] |
| | Q&P-MMn | 0.173C-4.46Mn-1.47Si-0.03Al, 0.195C-4.52Mn-0.04Si-1.31Al | 1,2 | [47] |
| | | 0.2C-1.50Si-4.02Mn-0.02Al, 0.2C-0.08Si-4.04Mn-1.46Al | 1, 2 | [49] |
| | TM | 0.20C-(0.20-1.50)Si-1.24Mn-(0.02-1.22)Al-0.2Cr-(0.003-0.005) Ti-(0.003-0.005)B | 1, 2, 3, 4 | [55] |

Table 1. Chemical composition (in mass%), measured properties, and references for low-carbon Si/Al-Mn first- and third-generation AHSSs used in various kinds of research.

2. Microstructure and Retained Austenite Characteristics

2.1. C-Si-Mn steel

In general, microstructure and retained austenite properties of the third-generation AHSSs are strongly controlled by the chemical composition and following heat treatment conditions:

- annealing process (austenite reverted transformation process) [35–39,44],
- quenching process [19–21,23,25,50–54],
- ausforming process [27,33],
- isothermal transformation (IT) process (or austempering process and bainite transformation process) [9–17,27–34,40,51–54],
- partitioning process [24,54], etc.

Representative heat treatment diagrams of the third-generation AHSSs are shown in Figure 2. Low-carbon TBF and TM steels can be produced by austenitizing and subsequent IT processes. For the TBF steel, IT processes at temperatures ($T_{\rm IT}$) above $M_{\rm s}$ (IT process (i)) or between $M_{\rm s}$ and $M_{\rm f}$ (IT process (ii)) are conducted. The CFB [27–34] and BF-MMn [45] steels are produced by the same heat treatment as the TBF steel (Figure 2a). The one-step Q&P steel [20] involves the IT process (ii) of the TBF steel (Figure 2b). For the TM steel, the IT process below $M_{\rm f}$ (iii) is applied after austenitizing (Figure 2a) [52]. The heat treatment corresponding to the IT process (iii) contains direct quenching to room temperature and



subsequent partitioning (DQ&P) process. The M-MMn steel is fabricated by the same IT process (iii) as the TM steel [57–59].

Figure 2. Heat treatment diagrams of (**a**) TBF, CFB and TM steels [52] and (**b**) one-step and two-step Q&P steels [20]. T_{IT} , T_Q , T_P , Ac_3 , M_s , and M_f are isothermal transformation temperature, quenching temperature, partitioning temperature, the austenite-finish temperature on heating, martensite-start temperature, and martensite-finish temperatures, respectively.

The major phase of the TBF steel is bainitic ferrite (α_{bf}) for the IT process (i) (Figure 3a) [81], and it a mixture of α_{bf} and primary coarse soft martensite (α_m) for the IT process (ii) (Figure 3b) [81]. In the TBF steel, an initial γ_R fraction ($f\gamma_0$) increases with increasing T_{IT} (Figure 4) [51]. The highest initial carbon concentration ($C\gamma_0$) is obtained in the TBF steel subjected to the IT process at the temperatures between M_s and M_f . In the TBF steel subjected to the IT process (ii), a small amount of MA phase (a mixture of secondary fine hard martensite (α_m^*) and film-like γ_R) exists as the second phase. In addition, only a small amount of carbide (θ) precipitates only in the α_m lath structure [15]. The microstructure of CFB [27–34], one-step Q&P [20], and BF-MMn [45] steels resembles that of TBF steels subjected to the IT processes (i) and/or (ii). In general, the above-mentioned $f\gamma_0$ and $C\gamma_0$ are calculated by the methods proposed by Maruyama [82] and Dyson and Holmes [83], respectively.



Figure 3. Illustration of the typical microstructure of various third-generation AHSSs [81]. (a): TBF, CFB, and BF-MMn steels ($T_{\text{IT}} > M_{\text{s}}$); (b): TBF, CFB, one-step Q&P, and BF-MMn steels ($T_{\text{IT}} = M_{\text{s}} - M_{\text{f}}$), and two-step Q&P and two-step Q&P-MMn steels ($T_{\text{P}} > M_{\text{s}}$); (c): TM and M-MMn steels ($T_{\text{IT}} < M_{\text{f}}$). α_{bf} , α_{m} , α_{m}^* , γ_{R} , θ , and MA represent bainitic ferrite, primary coarse soft martensite, secondary fine hard martensite, retained austenite, carbide, and MA phase (a mixture of α_{m}^* and film-like γ_{R}), respectively.



Figure 4. Variations in initial volume fraction ($f\gamma_0$, •) and carbon concentration ($C\gamma_0$, \bigcirc) of retained austenite as a function of isothermal transformation temperature (T_{IT}) in 0.20C-1.59Si-1.50Mn-0.05Nb (mass%) TBF and TM steels [51]. The holding time of the IT process is 1000 s. This figure is reproduced based on reference [51].

On the other hand, the major phase of TM steel is α_m or auto-tempered primary martensite (Figure 3c) [81]. The TM steel contains a large amount of MA phase and a small amount of θ in the α_m lath structure as the second phase [51–53]. The θ fraction (f_{θ}) increases with decreasing T_{IT} [52], although it is much lower than that of quenching and tempering (Q&T) steel [52,54]. The TM steel is also called "TRIP-aided duplex martensitic steel" because its microstructure consists of α_m matrix structure and a large amount of MA second phase. It is noteworthy that M-MMn steel contains much larger amounts of MA phase and γ_R than TM steel [59].

The two-step Q&P process generally consists of quenching to a temperature (T_Q) between M_S and M_f after austenitizing and subsequent partitioning at a temperature (T_P) higher than M_s (Figure 2b) [19,20,48]. The process forms the microstructure of α_{bf} and α_m matrix and γ_R , similar to the IT process (ii) for TBF and CFB steels and the one-step Q&P process (Figure 3b). The variation in volume fractions of various phases as a function of T_Q is illustrated in Figure 5a. On the quenching to T_Q , a certain amount of austenite transforms to α_m first. The α_m fraction ($f \alpha_m$) can be estimated by the following empirical equation proposed by Koistinen and Marburger [84].

$$f\alpha_{\rm m} = 1 - \exp\left\{-1.1 \times 10^{-2} \left(M_{\rm S} - T_{\rm Q}\right)\right\} \tag{1}$$

If the T_Q is close to M_f , a small amount of carbide (θ) precipitates only in the α_m lath structure [23,49]. During subsequent partitioning at temperatures above M_s , most of the remaining austenite transforms into α_{bf} . At the same time, the α_m softens through carbon migration (carbon enrichment) into untransformed austenite and carbide precipitation [23]. During final cooling to room temperature, a part of unstable austenite transforms into the MA phase. Typical two examples of T_Q dependence of $f\gamma_0$ in two-step Q&P steels with different carbon content are shown in Figure 5b. In these two-step Q&P steels, the optimum T_Q which gives the maximum volume fraction of γ_R is between M_s and M_f [19,20,23].



Figure 5. (a) Illustration of variations in volume fractions of different phases as a function of quenching temperature (T_Q) in two-step Q&P steel. (b) Variations in initial retained austenite fraction with T_Q in 0.2C-4.0Mn-1.6Si-1.0Cr ($M_s = 273 \,^{\circ}$ C) and 0.3C-1.0Mn-1.6Si-1.0Cr ($M_s = 235 \,^{\circ}$ C) two-step Q&P steels [23]. (a) is reproduced based on references [19,20], in which $f\alpha_m$ and $f\gamma$ are volume fractions of primary coarse soft martensite and austenite as functions of T_Q prior to partitioning. The final or initial austenite fraction ($f\gamma_0$) at room temperature is given by a red, bold, solid line. $f\alpha_m' = f\alpha_{bf}$ (bainitic ferrite fraction) + $f\alpha_m^*$ (secondary fine hard martensite fraction). (b) is reprinted with permission from Elsevier, copyright 2022.

The mechanical properties of the third-generation AHSSs are predominantly controlled by the volume fraction and mechanical stability of γ_R . The mechanical stability is mainly related to its $C\gamma_0$, along with the size and stacking fault energy (SFE) of γ_R , matrix structure, deformation temperature, other microalloying elements, etc. [85]. According to Sugimoto et al. [86], the mechanical stability of γ_R can be defined by the following *k*-value (or the strain-induced transformation factor),

$$k = (\ln f \gamma_0 - \ln f \gamma) / \varepsilon \tag{2}$$

where $f\gamma$ is the retained austenite fraction after deformation to a certain plastic strain (ε). Sherif et al. [87] re-expresses the mechanical stability through the chemical free energy available for transformation as follows:

$$\ln f \gamma_0 - \ln f \gamma = k_1 \,\Delta G^{\alpha' \gamma} \,\varepsilon \tag{3}$$

where k_1 is a modified *k*-value. $\Delta G^{\alpha'\gamma}$ (= $G^{\alpha'} - G^{\gamma}$) is the chemical free-energy change for the transformation of austenite to ferrite (martensite) with the same composition (without considering stored energy due to the shape deformation), where $G^{\alpha'}$ and G^{γ} are the chemical free energies of ferrite (martensite) and austenite, respectively.

In general, the *k*- and *k*₁-values decrease with increasing $C\gamma_0$ in the TPF steel. As shown in Figure 6 [88], the *k*-values of low-carbon Si-Mn TBF, TM, D-MMn, and M-MMn steels also decrease with increasing $C\gamma_0$, although the *k*-values are about three times higher than those of TPF and TAM steels because of lower $C\gamma_0$ [89]. It is noteworthy that the *k*-values of 3Mn and 5Mn M-MMn steels are lower than those of TBF and TM steels, like 3Mn D-MMn steel. This is because high solute Mn concentration in the γ_R plays a role in the austenite stabilizer.



Figure 6. Relationships between strain-induced transformation factor (*k*) and initial carbon concentration of retained austenite ($C\gamma_0$) in 0.2C-1.5Si-1.2Mn-0.2Cr-(0.022-1.22)Al (0Al, 0.7Al, 1.2Al) TM steels (•) [55,88], 0.2C-1.5Si-(1.5-5.0)Mn (1.5Mn, 3Mn, and 5Mn) M-MMn (•) [59] and D-MMn (•) [39,40] steels, 0.2C-1.5Si-1.5Mn-(0-1.0)Cr-(0-0.2)Mo (Cr-Mo) TM steels (•) [53] and 0.2C-1.5Si-1.5Mn (1.5Si) TBF steel (Δ) [51], and 0.2C-1.5Si-1.5Mn TPF (\boxtimes) [89] and TAM (\boxtimes) [89] steels.

2.2. C-Si/Al-Mn Steel

In general, austenite-stabilizing elements such as Ni, Mn, C, N, and Cu, lower the critical temperature (T_0) at which austenite and martensite have the same chemical free energy in steel [85]. Only Co increases the T_0 and makes the austenite unstable. Many ferrite-stabilizing elements also stabilize the austenite, although they increase the T_0 . Exceptionally, Cr lowers the T_0 and significantly increases the austenite stability by the addition of several mass% points, although it is a ferrite-stabilizing element [85].

Al makes the austenite unstable as a ferrite-stabilizing element, like the Co of an austenite-stabilizing element [85]. Ehrhardt et al. [90], Sugimoto and Mukherjee [69], and Alza and Chavez [91] summarize the effect of Al on the microstructure in TPF steel (Figure 7a). According to them, the advantage of Al over Si is increasing the driving force for austenite to bainite transformation, which accelerates the bainite transformation kinetics resulting from an increased nucleation rate. Examples of T_0 curves calculated for 19C-1.54Si-1.51Mn-0.04Al (0Al), 0.5Al: 0.20C-0.99Si-1.51Mn-0.49Al (0.5Al), and 0.20C-0.49Si-1.50Mn-0.99Al (1.0Al) TPF steels are shown in Figure 7b [75]. Al shifts the T_0 line to the high-carbon-concentration side. This means that Al also plays a role in lowering the *k*-value or increasing the mechanical stability [88]. Al increases the SFE of γ_R , and Cu and Si [92], which makes TRIP and TWIP difficult [93]. However, there are some disadvantages to adding Al: it can reduce solid-solution-strengthening and raises the M_s [91].

For 0.2C-(0.5-1.5)Si-1.5Mn-(0.04-1.0)Al-(0-0.2)Mo-(0-0.05)Nb TBF steels [12] and 0.25C-(0.55-1.45)Si-(1.61-1.70)Mn-(0.3-0.69)Al Q&P steels [21], partial replacement of Si by Al increases the mechanical stability of γ_R and decreases its volume fraction. In this case, the increased mechanical stability is mainly associated with higher C γ_0 and higher SFE. A similar effect of Al is obtained in 0.2C-(0.2-1.5)Si-1.24Mn-(0.02-1.22)Al-0.2Cr-0.003B TM steels [55,88] (Figure 6), although the mechanical stability and volume fraction are lower than those of 0.2C-(0.5-1.5)Si-1.5Mn-(0.04-1.0)Al-(0-0.2)Mo-(0-0.05)Nb [12] and 0.2C-(1.0-1.544)Si-1.5Mn-(0.04-0.5)Al-(0-0.05)Nb TBF steels [14] and 0.2C-(0.5-1.5)Si-1.5Mn-(0.04-1.0)Al TPF and TAM steels [75,78]. According to Sugimoto et al. [55], this is caused by insufficient carbon enrichment during the IT process at a lower temperature after the DQ process, which leads to a large amount of MA phase. Additionally, a similar effect of Al on the *k*-value or C γ_0 has been reported for 0.2C-(0.08-1.5)Si-2Mn-(0.02-1.46)Al Q&P-MMn [49] steels.



Figure 7. (a) Effects of microalloying elements on the time-temperature transformation curve of steel [90]. (b) Calculated T_0 curves and measured carbon concentration of retained austenite ($C\gamma_0$) on 19C-1.54Si-1.51Mn-0.04Al (0Al), 0.5Al: 0.20C-0.99Si-1.51Mn-0.49Al (0.5Al), and 0.20C-0.49Si-1.50Mn-0.99Al (1.0Al) TPF steels [75]. (a) is reprinted with permission from AIST, copyright 2022. (b) is reproduced with permission from ISIJ, copyright 2022.

The partial replacement of Si by Al refines the prior austenite grain, α_{bf} lath, and film-like γ_R in 0.2C-(0.5-1.5)Si-1.5Mn-(0.04-1.0)Al-(0-0.2)Mo-(0-0.05)Nb [12] and 0.2C-(0.5-1.5)Si-(1.5-2.5)Mn-(0.04-1.0)Al TBF steels [13]. Zhu et al. [28] show that Al addition of 1.0 to 1.5 mass% results in a remarkable refinement of α_{bf} lath, film-like γ_R , and MA island in 0.25C-(0.1-1.09)Si-2.07Mn-(0.02-1.54)Al CFB steels. A similar result was also found by Tian et al. [32]. For θ precipitation, He et al. [31] found that Al addition successfully suppresses the formation of θ in 0.25C-2.07Mn-(0.02-1.54)Al CFB steels through effective carbon enrichment from transformed α_{bf} to the adjacent untransformed austenite [94]. Kaar et al. [47] and Wallner et al. [49] report that a significantly larger amount of triaxial aligned θ is precipitated in the α_m matrix in 0.173C-4.46Mn-1.47Si-0.03Al and 0.20C-4.52Mn-0.04Si-1.31Al two-step Q&P-MMn steels, respectively, although Al exhibits lower suppression of θ precipitation than Si.

For the α_m transformation, Kobayashi et al. [55] found that the α_m size is largely unchanged by Al addition in 0.2C-(0.2-1.5)Si-1.24Mn-(0.02-1.22)Al-0.2Cr-0.003B TM steels, unlike the above results relating to the α_{bf} transformation. In this case, the prior austenitic grain size was nearly the same in both sheets of steel. Kantanen et al. [24] also showed that the α_m size is not influenced by Al content in 0.30C-0.56Si-2.00Mn-1.10Al-2.20Cr two-step Q&P steel.

3. Tensile Properties

3.1. C-Si-Mn Steel

According to Sugimoto et al. [15,54], flow stress (strain), $\sigma(\varepsilon)$, of the AHSSs containing γ_R of 4 to 30 vol% is formulated by

$$\sigma(\varepsilon) = \sigma^{M}(\varepsilon) + \Delta \sigma_{h}(\varepsilon) \tag{4}$$

where $\sigma^{M}(\varepsilon)$ and $\Delta \sigma_{h}(\varepsilon)$ are the flow stress of the matrix and strain hardening increment of the steel, respectively. The $\Delta \sigma_{h}(\varepsilon)$ can be estimated by

$$\Delta \sigma_{\rm h}(\varepsilon) = \Delta \sigma_{\rm i}(\varepsilon) + \Delta \sigma_{\rm t}(\varepsilon) + \Delta \sigma_{\rm f}(\varepsilon) \tag{5}$$

where $\Delta \sigma_i(\varepsilon)$, $\Delta \sigma_t(\varepsilon)$, and $\Delta \sigma_f(\varepsilon)$ represent "the long-range internal stress hardening", "the strain-induced transformation hardening", and "the forest dislocation hardening", respectively, which can be formulated by

$$\Delta \sigma_{i}(\varepsilon) = \{(7 - 5\nu)\mu / 5(1 - \nu)\} f \cdot \varepsilon_{p}^{u}$$
(6)

$$\Delta \sigma_{\rm t}(\varepsilon) = g(\Delta f \alpha_{\rm m}) \tag{7}$$

$$\Delta \sigma_{\rm f}(\varepsilon) = \zeta \mu \ (\mathbf{b} \cdot \mathbf{f} \cdot \varepsilon / 2r)^{1/2} \tag{8}$$

where v is the Poisson's ratio, μ is the shear modulus, ε_p^{u} is "the eigenstrain" [95], f is the volume fraction of the second phase, $g(\Delta f \alpha_m)$ is a function of the strain-induced martensite fraction, ζ is a material constant, b is the Burgers vector, and r is particle radius of the second phase. In the D-MMn steel, the second phase is mainly film-like γ_R . In the TBF, CFB, one-step Q&P, and BF-MMn steels subjected to the IT process at the temperatures above M_s , the second phase corresponds to untransformed carbon-enriched γ_R and strain-induced martensite. In these steels subjected to the IT process at the temperatures between M_s and M_f , the second phases correspond to α_m , untransformed γ_R , and strain-induced martensite, in the same way as two-step Q&P and two-step Q&P-MMn steels. Sometimes, a small amount of MA phase is also classified into the second phase in these steels. On the other hand, the second phase of the TM and M-MMn steels is mainly equivalent to the MA phase. The strain-hardening mechanism of two-step Q&P steel is also suggested by Celada-Casero et al. [25].

Tensile ductility defined by uniform elongation (UEl), TEl, and reduction in area (RA) are mainly controlled by the strain-hardening behavior mentioned above in the third-generation AHSSs, which is related to the chemical composition and heat treatment conditions, such as austenitizing (annealing) temperature, T_Q , T_{IT} , T_P , and these holding times. The effects of T_Q on the tensile properties in (0.2–0.3)C-1.6Si-4.0Mn-1.0Cr steels subjected to the two-step Q&P process are shown in Figure 8 [23]. The largest TEls of both sheets of steel were obtained by quenching at $T_Q = M_s - 100$ °C, which was about 50 °C lower than T_Q for the largest volume fraction of γ_R (Figure 4b). The optimum T_Q s for TEl roughly match those for the minimum TS. On the other hand, the minimum yield stress (YS) was about 50 °C higher than the optimum T_Q for the TEl.



Figure 8. Quenching temperature dependences of yield stress (YS), tensile strength (UTS), and total elongation (TE) of 0.2C-4.0Mn-1.6Si-1.0Cr (0.2C) and 0.3C-4.0Mn-1.6Si-1.0Cr (0.3C) two-step Q&P steels partitioned at T_P = 450 °C for 300 s after quenching [23]. M_s s of the 0.2C and 0.3C Q&P steels are 273 and 235 °C, respectively. This figure is reprinted with permission from Elsevier, copyright 2022.

The effects of $T_{\rm IT}$ on the tensile properties of 0.20C-1.50Si-1.50Mn-0.05Nb TBF and TM steels are shown in Figure 9 [51]. In these steels, the largest UEL, TEL, and TS×TEl can be obtained at $T_{\rm IT}$ above $M_{\rm s}$, although the YS and TS considerably decrease. The optimum $T_{\rm IT}$ for tensile ductility agrees well with one for the largest fraction of $\gamma_{\rm R}$ [51]. In the TM steel subjected to the IT process at temperatures below $M_{\rm f}$, relatively high TS and TS×TEl are

achieved, compared to the TBF steels subjected to the IT process at temperatures between M_s and M_f . In this case, the YS of the TM steel decreases due to a large amount of MA phase, resulting in the continuous yielding [96], in the same way as ferrite-martensite DP steel.

For the TBF, CFB, one-step Q&P, two-step Q&P, BF-MMn, and Q&P-MMn steels, "the long-range internal stress hardening" and "the strain-induced transformation hardening" mainly contribute to the high ductility [10,14]. In the TM and M-MMn steels, a large amount of MA phase mainly increases "the long-range internal stress hardening", with a small contribution to "the strain-induced transformation hardening" because of a small amount of γ_R or small $\Delta f \alpha_m$.

It is interesting that 0.2C-1.5Si-(1.5–5.0)Mn M-MMn [59] and D-MMn [39,40] steels achieve much higher TS×TEls than those of TBF and TM steels (Figure 10b).



Figure 9. Variations in (**a**) yield stress (YS), tensile strength (TS), (**b**) uniform elongation (UEI), and total elongation (TEI); (**c**) product of TS and TEI (TS×TEI) as a function of isothermal transformation temperature ($T_{\rm IT}$) in 0.20C-1.50Si-1.50Mn-0.05Nb TBF and TM steels. This figure is reproduced based on reference [51].



Figure 10. (a) Engineering stress–strain (*σ*-ε) curves of 0.20C-1.50Si-1.24Mn-0.20Cr-0.022Al-0.0028 (0Al), 0.20C-0.73Si-1.24Mn-0.19Cr-0.72Al-0.0027B (0.7Al), and 0.20C-0.20Si-1.24Mn-0.20Cr-1.22Al-0.0026B (1.2Al) TM steels [55,88]. (b) Combination of the tensile strength (TS) and total elongation (TEl) of 0Al, 0.7Al, and 1.2Al TM steels (●) [55,88], 0.2C-1.5Si-1.5Mn-(0-1.0)Cr-(0-0.2)Mo (Cr-Mo) TM steels (▲) [53], 0.2C-1.5Si-(1.5-5.0)Mn (1.5Mn, 3Mn, and 5Mn) M-MMn (■) [59] and D-MMn (□) [39,40] steels, 0.2C-1.5Mn-0.99Si-0.49Al (0.5Al) TBF steel (○), 0.2C-1.5Mn-1.00Si-0.48Al-0.049Nb (0.5Al-0.05Nb) TBF steel (○) [14], and 0.2C-1.5Si-1.5Mn-0.05Nb (1.5Si) TBF steel (△) [51]. (b) is produced based on references [14,39,40,51,53,55,59,88].

3.2. C-Si/Al-MnSsteel

Partial replacement of Si by Al lowers the strain hardening rate and flow stress in 0.20C-(0.20-1.50)Si-1.24Mn-(0.022-1.22)Al-0.20Cr-(0.0026-0.0028)B TM steel (Figure 10a) [55,88],

because solid solution-hardening of Al (24 MPa/at%) is about half that of Si (55 MPa/at%) in Fe-C ferrous steel [97]. Resultantly, the YS, TS, UEl, and TEl decrease with increasing Al content under a constant Si+Al content, although the RA increases with increasing Al content [88]. Additionally, a similar result can be obtained in 0.2C-(0.99-1.54)Si-1.51Mn-(0.033-0.49)Al-(0-0.05)Nb TBF steels subjected to the IT process at temperatures between M_s and M_f [14]. If the addition of Al is below 0.7 mass% in the TBF and TM steels, these steels keep the TS×TEl of 10 GPa%, which is slightly lower than those of Al-free TBF and TM steels [14,88] (Figure 10b). However, Al addition of 1.2 mass% (or the partial replacement of Si by 1.2 mass% Al) considerably decreases the TS×TEl in TM steel in a TS range above 1.2 GPa (see 1.2Al TM steel in Figure 10b).

It is noteworthy that further addition of 0.05 mass% Nb is suitable to increase the TS×TEl in Al-bearing TBF steel [14] (see 0.5Al-0.05Nb TBF steel in Figure 10b). In this case, precipitation hardening by fine NbC contributes to increasing the TS. A similar result is reported in 0.2C-0.77Si-2.0Mn-0.76Al CFB steel [34], except for the TEl. In 0.25C-0.55Si-1.70Mn-0.69Al Q&P steel, 0.195C-4.52Mn-0.04Si-1.31Al Q&P-MMn steel, and 0.2C-0.08Si-4.04Mn-1.46Al Q&P-MMn steel. Low ductility resulting from the partial replacement of Si by Al was also reported by De Moor et al. [21,22], Kaar et al. [47], and Wallner et al. [49]. According to Sugimoto et al. [52] and Pham et al. [53], the volume fraction of θ in α_m lath structure hardly influences the ductility of the Al-added TM steels because the θ fraction is very little. According to Jing et al. [42], 1.39 mass% Al addition achieved the highest TS×TEl in (0.18-0.19)C-(7.66-7.93)Mn-(0-2.79)Al D-MMn steels.

As shown in Figure 11a, the TS×TEls of various TBF and TM steels tend to increase with the volume fraction of γ_R [14,39,40,51,53,55,59,88]. In addition, they increase as the *k*-value decreases (Figure 11b). Therefore, a decrease in TS×TEl of 1.2Al TM steel may be caused by the decreased volume fraction of γ_R and low solid solution-hardening, although the mechanical stability of γ_R increases (or the *k*-value decreases). The TS×TEl− $f\gamma_0$ relation of 5Mn M-MMn steel is superior to those of the other TBF and TM steels. This is associated with the increased mechanical stability (decreased *k*-value), the volume fraction of γ_R , and the MA phase due to its high Mn concentration [88]. 5Mn D-MMn steel has a higher TS×TEl than 5Mn M-MMn steel. This is mainly caused by the high-volume fraction of γ_R .



Figure 11. (a) Relationship between tensile strength (TS) and initial retained austenite fraction ($f\gamma_0$). (b) Relationships between product of TS and total elongation (TS×TEl) and strain-induced transformation factor (k) in 0.20C-1.50Si-1.24Mn-0.20Cr-0.022Al-0.0028 (0Al), 0.20C-0.73Si-1.24Mn-0.19Cr-0.72Al-9.0027B (0.7Al), and 0.20C-0.20Si-1.24Mn-0.20Cr-1.22Al-0.0026B (1.2Al) TM steels (\bullet) [55,88], 0.2C-1.5Si-1.5Mn-(0-1.0)Cr-(0-0.2)Mo (Cr-Mo) TM steels (\blacktriangle) [53], 0.2C-1.5Si-(1.5-5.0)Mn (1.5Mn, 3Mn and 5Mn) M-MMn (\blacksquare) [59] and D-MMn (\square) [39,40] steels, 0.2C-1.5Mn-0.99Si-0.49Al (0.5Al) TBF steel (\bigcirc), 0.2C-1.5Mn-1.00Si-0.48Al-0.049Nb (0.5Al-0.05Nb) TBF steel (\bigtriangledown) [14], and 0.2C-1.5Si-1.5Mn-0.05Nb (1.5Si) TBF steel (\bigtriangleup) [51]. (a,b) are produced from the results in references [14,39,40,51,53,55,59,88].

4. Stretch formability

4.1. C-Si-Mn Steel

The stretch formability is defined by the maximum stretch height (H_{max}), which is usually measured using a ball-head punch. Figure 12a shows the $T_{\rm IT}$ dependence of the $H_{\rm max}$ in 0.20C-1.50Si-1.50Mn-0.05Nb TBF and TM steels subjected to the IT process at $T_{\rm IT}$ = 200 °C to 450 °C for 200 s after austenitizing [51]. For comparison, the T_P dependence of the H_{max} of 0.20C-1.50Si-1.50Mn-0.05Nb TM steel subjected to the DQ&P process after austenitizing and then partitioning at $T_{\rm P}$ = 200 °C to 450 °C for 1000 s is shown in the Figure 12a. The H_{max} s values of the IT-processed TBF and TM steels (red line) increase with increasing T_{IT} . The H_{max} of DQ&P-processed TM steel increases with increasing $T_{\rm P}$, although the $H_{\rm max}$ s are lower than those of IT-processed TBF and TM steels. This result indicates that the IT process is suitable compared to the DQ&P process in TBF and TM steels. According to Kobayashi et al. [51], this result is associated with a larger amount of $\gamma_{\rm R}$. As shown in Figure 12b, the products of TS and $H_{\rm max}$ (TS× $H_{\rm max}$ s) of the IT-processed TBF and TM steels and DQ&P-processed TM steel are much higher than those of 0.2C-1.5Si-(1.5-5.0)Mn M-MMn steels [59], 0.082C-0.88Si-2.0Mn ferrite-martensite DP steel [16,51,52], and 0.23C-0.19Si-1.29Mn-0.21Cr-0.003B 22MnB5 steel subjected to the Q&T process (22MnB5 Q&T steel) [16,52] in a TS range above 1.0 GPa, although they are a little lower than those of 0.2C-1.5Si-(1.5-5.0)Mn D-MMn steels [39]. It is very important to know that large tensile ductility does not necessarily lead to high stretch formability because of the different stress states, although it exhibits a linear relationship with stretch formability. Namely, equi-biaxial tension growing on stretch-forming promotes crack and/or void initiation compared to uniaxial tension [80].

It is noteworthy that the H_{max} of 5Mn D-MMn steel is further increased to 10.5 mm by warm forming at 200 °C, which significantly increases the mechanical stability of γ_{R} [39].



Figure 12. (a) Isothermal transformation (T_{IT}) and partitioning temperature (T_{P}) dependences in maximum stretch height (H_{max}) in 0.20C-1.50Si-1.50Mn-0.05Nb TBF and TM steels subjected to isothermal transformation (IT) process (•) and TM steel subjected to direct quenching and then partitioning (DQ&P) process (•) [51]. (b) Relationship between H_{max} and tensile strength (TS) in the IT-processed TBF and TM steels (•) [51], DQ&P-processed TM steel (•) [51], 0.2C-1.5Si-(1.5-5.0)Mn (1.5Mn, 3Mn, and 5Mn) M-MMn (**□**) [59] and D-MMn (**□**) [39] steels, 0.082C-0.88Si-2.0Mn ferrite-martensite DP steel (\diamond) [14,51,52], and 0.23C-0.19Si-1.29Mn-0.21Cr-0.003B 22MnB5 Q&T steel (**⊠**) [14,52]. (a) is reproduced based on reference [51]. (b) is reproduced based on references [14,39,51,52,59].

4.2. C-Si/Al-Mn Steel

In 0.20C-1.50Si-1.24Mn-0.20Cr-(0.022-1.22)Al-0.00bB TM steels (0Al, 0.7Al, and 1.2Al TM steels) under a condition of Si+Al = 1.5 mass%, the TS× H_{max} decreases with increasing Al content in the same way as the TEl and TS×TEl in (Figure 13a) [88]. It is interesting that partial replacement of Si by 1.2 mass% Al considerably decreases the H_{max} and

TS× H_{max} [55,88], compared to those of 0.2C-1.5Si-1.5Mn-0.05Nb (1.5Si) TBF steel [51] and 0.2%-1.5Si-1.5Mn-(0-1.0)Cr-(0-0.2)Mo (Cr-Mo) TM steels [53].

As shown in Figure 13b, the H_{max} exhibits a linear relationship with the TEl in 0Al, 0.7Al, and 1.2Al TM steels, the Cr-Mo TM steels, and the 1.5Si TBF steels, although the slopes of 5Mn M-MMn and D-MMn steels are lower than those of these steels. Kobayashi et al. [55] propose that lower H_{max} and TS× H_{max} of 1.2Al TM steel may be caused by lower UEl and Tel, resulting in low solution-hardening and low γ_R fraction. According to Sugimoto et al. [59], the small H_{max} of the 5Mn M-MMn steel is caused by the presence of a much larger MA phase, although a large amount of metastable γ_R makes a positive contribution to the H_{max} . In this case, a crack initiation at the interface between the matrix and MA phase is promoted by equi-biaxial tension, as opposed to uniaxial tension (tensile test).

The effect of a complex addition of Al and other alloying elements on the H_{max} has not been investigated for the third-generation AHSSs. It is expected that the complex addition of Al and Nb enhances the H_{max} because it achieves a large TEl in 0.5Al-0.05Nb TBF steel (Figure 10b).



Figure 13. (a) Maximum stretch height–tensile strength (H_{max} −TS) relation in 0.20C-1.50Si-1.24Mn-0.20Cr-0.022Al-0.0028B (0Al), 0.20C-0.73Si-1.24Mn-0.19Cr-0.72Al-9.0027B (0.7Al), and 0.20C-0.20Si-1.24Mn-0.20Cr-1.22Al-0.0026B (1.2Al) TM steels (●) [55,88], 0.2C-1.5Si-1.5Mn-(0-1.0)Cr-(0-0.2)Mo (Cr-Mo) TM steels (▲) [53], and 0.2C-1.5Si-1.5Mn-0.05Nb (1.5Si) TBF steel (△) [51]. (b) Relationship between H_{max} and total elongation (TEl) in these steels and 0.2C-1.5Si-(1.5-5.0)Mn (1.5Mn, 3Mn and 5Mn) M-MMn (■) [59] and D-MMn (□) [39] steels. (**a**,**b**) are produced based on references [39,51,53, 55,59,88].

5. Stretch-Flangeability

5.1. C-Si-Mn Steel

In general, the stretch-flangeability can be evaluated by the following hole-expansion ratio (HER) using a punched-hole specimen:

HER =
$$(d_{\rm f} - d_0)/d_0$$
 (9)

where d_0 and d_f are the original diameter of the punched hole and the hole diameter upon cracking during the hole-expansion test, respectively. In many cases, the hole-punching tests are carried out at a clearance of about 10% [98]. The subsequent hole-expansion tests are conducted using a conical punch tool with a vertical angle of 60 deg. [55,59]. Recently, the conical punch tool has been preferentially used to measure the HER.

The HER increases with increasing $T_{\rm IT}$ in 0.20C-1.50Si-1.50Mn-0.05Nb IT-processed TBF and TM steels and increases with increasing $T_{\rm P}$ in DQ&P-processed TM steel (Figure 14a) [51]. In this case, the TM steel subjected to the IT process at the temperatures of $M_{\rm f}$ (50 to 100) °C and TBF steel subjected to the IT process at the temperatures between $M_{\rm s}$ and $M_{\rm f}$ achieve higher TS×HER (50 to 60 GPa%) than the DQ&P-processed

TM steel. In addition, the IT-processed TBF and TM steels have much higher TS×HER than 5Mn M-MMn [59] and 5Mn D-MMn steels [40], ferrite-martensite DP steel [16,51,52], and 22MnB5 Q&T steel [16,52] (Figure 14b). According to Kobayashi et al. [51], the high TS×HER of the IT-processed TM steel is mainly caused by (i) uniform $\alpha_{\rm m}$ lath structure with low θ fraction and (ii) plastic relaxation of localized stress concentration by the strain-induced transformation of metastable γ_R at the α_m lath boundary and/or in a finely dispersed MA phase, which suppresses the void and/or cracks initiation on punching and void coalescence or cracking on hole expansion. For the IT-processed TBF steel, the excellent stretch-flangeability is also associated with (iii) a uniform fine mixture of α_{bf} and α_m , and the above (i) and (ii). The effects of $T_{\rm IT}$ on the HERs are also reported for 0.20C-1.40Si-1.70Mn-0.045Nb TBF steel [17], 0.2C-0.2Si-2Mn-0.03Ti-0.003B TBF steel [16], 0.13C-1.35Si-2.10Mn-0.98Cr (and 0.18C-1.40Si-2.13Mn-1.00Cr) TBF and DQ-processed TM steels [56], and 0.22C-1.48Si-3.79Mn-0.98Cr Q&P-MMn steel [46]. The effect of T_Q (in the Q&P process) on the stretch-flangeability was reported by Im et al. [25] using 0.18C-1.5Si-2.6Mn steel (M_s = 368 °C). In this case, higher HER was obtained by a lower quenching process at $T_{\rm O}$ (= 280 °C), followed by partitioning at $T_{\rm P}$ = 425 °C.



Figure 14. (a) Isothermal transformation and partitioning temperature ($T_{\rm IT}$ and $T_{\rm P}$) dependences of hole-expansion ratio (HER) of 0.20C-1.50Si-1.50Mn-0.05Nb TBF and TM steels subjected to isothermal transformation (IT) process (•) and TM steel subjected to a direct quenching and partitioning (DQ&P) process (•) [51]. (b) HER–TS (tensile strength) relation in the IT-processed TBF and TM steels (•) [51], the DQ&P-processed TM steel (•) [51], 0.2C-1.5Si-(1.5-5.0)Mn (1.5Mn, 3Mn, and 5Mn) M-MMn (**□**) [59] and D-MMn (**□**) [40] steels, 0.082C-0.88Si-2.0Mn ferrite-martensite DP steel (•) [14,51,52], and 0.23C-0.19Si-1.29Mn-0.21Cr-0.003B 22MnB5 Q&T steel (**∞**) [14,52]. (a) is reproduced based on references [14,40,51,52,59].

5.2. C-Si/Al-Mn Steel

As shown in Figure 15a, partial replacement of Si by Al also keeps high TS×HER in 0.2C-(0.2-1.5)Si-1.24Mn-(0.022-1.22)Al-0.2Cr (0Al, 0.7Al, and 1.2Al) TM steels [55,88]. The TS×HER is at the same level as those of Al-free TM and TBF steels, such as 0.2C-1.5Si-1.5Mn-(0-1.0)Cr-(0-0.2)Mo (Cr-Mo) TM steels [53], 0.2C-1.0Si-1.5Mn-0.5Al (0.5Al) TBF steel [14], 0.2C-1.0Si-1.5Mn-0.5Al-0.05Nb (0.5Al-0.05Nb) TBF steel [14], and 0.2C-1.5Si-1.5Mn-0.05Nb (1.5Si) TBF steel [51]. De et al. [74], Sugimoto et al. [75], and Samek et al. [76] investigated the TS×HER of low-carbon Si/Al-Mn TPF [74–76] and TAM [75] steels. In the TPF and TAM steels, the replacement of Si by Al keeps a high TS×HER in the same way as the low-carbon Si/Al-Mn TBF and TM steels [14].

Differing from the drilled hole samples [69], the HER of the punching hole samples is considerably controlled by the surface-layer damage on hole-punching in TM and TBF steels [14,53,55]. The main surface-layer damage is measured by the size and number of voids and/or cracks in the break section on the punched surface, which are controlled by

clearance between die [98] and punch, punching temperature and speed [89], and retained austenite characteristics [89]. In general, the punching damage can be also evaluated by a ratio of shear section length to sheet thickness (*ss*/*t*), because the larger the *ss*/*t* value, the higher the HER in TBF, TM, and M-MMn steels, as shown in Figure 16a. According to Kobayashi et al. [55], Sugimoto and Kobayashi [88], and Sugimoto et al. [14], the *ss*/*t* value increases with increasing Al content in 0Al, 0.7Al, and 1.2Al TM steels and 0.5Al-0.05Nb TBF steel. As shown in Figure 16b, the large *ss*/*t* of 1.2Al TM steel is connected to a large RA or local ductility. Therefore, the small punching damage may bring on high HER in 1.2Al TM steel, resulting from uniform α_m lath structure, the high mechanical stability of γ_R , and low solution-hardening.

In 0.2C-(1.0-1.5)Si-1.5Mn-(0.04-0.5)Al-(0-0.05Nb) TBF steels subjected to the IT process at the temperatures just below M_s , the partial replacement of Si by Al also achieved high TS×HER (see 0.5Al TBF steel in Figure 15b), in the same way as Al-free (0Al) TBF steel (see 0Al TBF steel in Figure 15b) [14]. Furthermore, the complex addition of 0.5 mass% Al and 0.05 mass% Nb considerably enhances the TS×HER (see 0.5Al-0.05Nb TBF steel in Figure 15b), compared to 0.5Al TBF steel. A similarly high TS×HER was also obtained by the complex addition of Al and Nb/Mo in 0.20C-(0.49-1.54)Si-(1.48-1.51)Mn-(0.04-0.99)Al-(0-0.05)Nb-(0-0.20)Mo TBF steels [12]. According to Sugimoto et al. [12,14], this increased stretch-flangeability by complex addition is mainly associated with the small punching damage due to refined microstructure, stabilized film-like γ_R , and TRIP effect on hole expansion. In this case, the complex addition brings on precipitation-hardening by fine NbC/Mo₂C.



Figure 15. (a) Relationship between hole-expansion ratio (HER) and tensile strength (TS) in 0.20C-1.50Si-1.24Mn-0.20Cr-0.022Al-0.0028 (0Al), 0.20C-0.73Si-1.24Mn-0.19Cr-0.72Al-9.0027B (0.7Al), and 0.20C-0.20Si-1.24Mn-0.20Cr-1.22Al-0.0026B (1.2Al) TM steels (\bullet) [55,88], 0.2C-1.5Si-1.5Mn-(0-1.0)Cr-(0-0.2)Mo (Cr-Mo) TM steels (\blacktriangle) [53], 0.20C-0.99Si-1.51Mn-0.49Al (0.5Al) TBF steel (\bigcirc) [14], 0.20C-1.00Si-1.50Mn-0.48Al-0.049Nb (0.5Al-0.05Nb) TBF steel (\bigtriangledown) [14], and 0.2C-1.5Si-1.5Mn-0.05Nb (1.5Si) TBF steel (\bigtriangleup) [51]. (b) Variations in the product of TS and HER (TS×HER) as a function of isothermal transformation temperature ($T_{\rm IT}$) in 0.19C-1.54Si-1.51Mn-0.04Al (0Al), 0.5Al, 0.21C-1.50Si-1.51Mn-0.04Al-0.048Nb (0Al-0.05Nb), and 0.5Al-0.05Nb TBF steels [14]. (a) is produced based on references [14,51,53,55,88]. (b) is reproduced with permission from ISIJ, copyright 2022.



Figure 16. (a) Relationship between hole expansion ratio (HER) and a ratio of shear section length to sheet thickness (*ss/t*) and (b) relationship between *ss/t* and reduction of area (RA) in 0.20C-1.50Si-1.24Mn-0.20Cr-0.022Al-0.0028 (0Al), 0.20C-0.73Si-1.24Mn-0.19Cr-0.72Al-9.0027B (0.7Al), and 0.20C-0.20Si-1.24Mn-0.20Cr-1.22Al-0.0026B (1.2Al) TM steels (\bullet) [55,88], 0.2C-1.5Si-1.5Mn-(0-1.0)Cr-(0-0.2)Mo (Cr-Mo) TM steels (\bullet) [53], 0.2C-1.5Si-(1.5-5.0)Mn (1.5Mn, 3Mn, and 5Mn) M-MMn steels (\bullet) [59], 0.20C-1.50Si-1.50Mn-0.05Nb (1.5Si) TBF steel (\triangle) [51], and 0.2C-1.00Si-0.48Al-0.049Nb (0.5Al-0.05Nb) TBF steels (\bigtriangledown) [14]. (**a**,**b**) are produced based on references [51,53,55,59,88].

6. Bendability

6.1. C-Si-Mn Steel

There is only a small amount of research on the bendability of the third-generation AHSSs [16,29,52,99]. The bendability is usually evaluated by the average bending angle [16,29] and minimum bending radius (R_{min}) [52,99]. Figure 17a shows the variation in average bending angle as a function of austempering temperature (or T_{IT}) in 0.34C-1.65Si-1.94Mn-1.07Cr CFB steel [29]. In this study, three-point bending tests were conducted to measure the average bending angle following the standard of the Association of German Automobile Industries (VDA 238-100). The largest average bending angle was obtained by the IT process at $T_{IT} = 350$ °C, just higher than M_s (329 °C), as shown in Figure 17a. The optimum T_{IT} corresponds to that of maximum total elongation (TE) (Figure 17b). In this case, the microstructure was a mixture of α_{bf} and γ_R , and the γ_R fraction was relatively high (Figure 17c). According to Rana et al. [29], the higher bendability is connected with a finer overall microstructure, the absence of α_m , and the high mechanical stability of γ_R .

Figure 18a shows the variations in R_{min} as functions of the TS in 0.21C-1.49Si-1.50Mn-1.0Cr-0.05Nb TBF and TM steels, 0.082C-0.88Si-2.0Mn ferrite-martensite DP steel, and 22MnB5 Q&T steel [52]. Small R_{min} was obtained in the TBF steel subjected to the IT process at the temperatures between M_s and M_f and in the TM steel subjected to the IT process at 200 °C ($<M_f = 261$ °C). The high bendability of the TM steel is caused by uniform microstructure, resulting in high local ductility despite a high MA phase fraction [52].

Figure 18b compares the R_{min} s of the TPF, TAM, and TBF steels with the chemistry of (0.1-0.6)C-1.5Si-1.5Mn [99]. Small R_{min} values are achieved in TBF and TAM steels with lath-type uniform microstructure and high mechanical stability of γ_{R} , resulting in high local ductility. In parallel with the bendability, much research to improve the spring back is also conducted for applications of the third-generation AHSSs [100].



Figure 17. Variations in (**a**) average bending angle, (**b**) yield stress (YS), ultimate tensile strength (UTS), and total elongation (TE), and (**c**) volume fractions of retained austenite and martensite as a function of austempering temperature in 0.34C-1.65Si-1.94Mn-1.07Cr CFB steel [29]. M_s of the steel is 329 °C.



Figure 18. Relationships between minimum bending radius (R_{min}) and tensile strength (TS) in (a) 0.21C-1.449Si-1.50Mn-1.0Cr-0.05Nb TBF and TM steels (•), 0.082C-0.88Si-2.0Mn ferrite-martensite DP steel (\diamond), and 22MnB5 Q&T steel (\boxtimes) [52] and (b) (0.1-0.6)C-1.5Si-1.5Mn TPF (\bigcirc), TAM (\square), and TBF (\triangle) steels [99]. (a) is reprinted with permission from AIST, copyright 2022. (b) is reprinted with permission from ISIJ, copyright 2022.

6.2. C-Si/Al-Mn Steel

Unfortunately, there are not any results on the effect of partial replacement of Si by Al on the bendability in the third-generation AHSSs. However, research investigating the effect of Al content on the R_{min} in the first-generation AHSSs was reported by Sugimoto et al. [75]. Figure 19 shows the effect of Al content on $R_{min} - T_{IT}$ relation in 0.2C-(0.5-1.5)Si-1.5Mn-(0.04-1.0)Al TPF and TAM steels with TS between 700 and 1000 MPa [75]. Partial displacement by 0.5 and 1.0 mass% Al under a condition of Si+Al = 1.5 mass% reduced the R_{min} in both sheets of steel. The optimum bendability was obtained in the TAM steel subjected to the IT process at temperatures between 275 and 425 °C, in the same way as the HER. According to Sugimoto et al. [75], the improved bendability of TAM steel is principally owed to the refined annealed martensite lath structure and the increased carbon concentration of γ_R needles. In general, small R_{min} is brought about by large RA. As Al-added TBF and TM steels have large RA values, a small R_{min} is expected to be achieved in the Al-added TBF and TM steels.



Figure 19. Variations in minimum bending radius (R_{min}) as a function of isothermal transformation temperature (T_{IT}) in (**a**) TAM and (**b**) TPF steels with the chemistry of 0.19C-1.54Si-1.51Mn-0.04Al (0Al), 0.20C-0.99Si-1.51Mn-0.49Al (0.5Al), and 0.20C-0.49Si-1.50Mn-0.99Al (1.0Al) [75]. (**a**,**b**) are reprinted with permission from ISIJ, copyright 2022.

7. Summary and Perspectives

Partial replacement of Si by Al improves the coatability (or galvanizing property) of steel sheets, in the same way as P. Therefore, many researchers hope to know the effect of the partial replacement on the cold formability in low-carbon, Si-Mn, third-generation AHSSs, such as the TBF, Q&P, CFB, D-MMn, BF-MMn, and Q&P-MMn steels (Group I) and TM and M-MMn steels (Group II). The effects of the partial replacement of Si by 0.04 to 1.5 mass% Al on the microstructure, tensile properties, and cold formability of the AHSS sheets are summarized as follows.

(1) The partial replacement of Si by Al decreases the volume fraction of γ_R and increases its mechanical stability in the third-generation AHSSs. In the TBF and TM steels, the partial replacement of Si by 0.7 mass% Al keeps the same large TEl and TS×TEl as Alfree steels. The TEL and TS×TEl decrease with increasing Al content, but the replacement of Si by 1.2 mass% Al deteriorates the TS×TEl in the TM steel with TS above 1.2 GPa, accompanied by decreases in YS, TS, and TEl. This is mainly caused by low solid solution-hardening and a decreased γ_R fraction, resulting in decreased flow stress and a decreased strain hardening rate, despite increased mechanical stability of γ_R . Similar results are reported for CFB, Q&P, and Q&P-MMn steels.

(2) The partial replacement of Si by 1.2 mass% Al considerably decreases the stretch formability in the TM steel with TS above 1.2 GPa, although the stretch formability decreases with increasing Al content. This is associated with the fact that equi-biaxial tension growing on stretch-forming plays a role in deteriorating the stretch formability through easy crack and/or void formation. On the other hand, the partial replacement of Si by 0.5 to 1.2 mass% Al results in the same stretch-flangeability as Al-free steels in the TBF and TM steels. This was associated with small punching damage which results from a uniform lath structure and high mechanical stability of γ_R , despite low solution hardening and a decreased γ_R fraction. Unfortunately, there is not any research on the bendability in Al-added AHSSs.

(3) A complex addition of Al and Nb/Mo achieves considerably larger TEl and TS×TEl in TBF steels. In addition, the complex addition also further increases the stretch-flangeability of the TBF steel owing to the small punching damage due to refined microstructure, stabilized film-like γ_R , and TRIP effect on hole-expanding. In this case, the complex addition brings on precipitation hardening by fine NbC/Mo₂C. Further research on the cold formabilities of the AHSSs with a complex addition of Al and other elements (Nb, T, V, Cr, Mo, Ni, B, etc.), resulting in higher tensile strength above 1.5 GPa, is also expected in the future.

(4) In order to apply the third-generation AHSSs to automotive sheet components, many studies on fatigue strength [6,54,81,101–103], toughness [6,29,43,54,57,81,88,104–106], hydrogen embrittlement resistance [13,18,107,108], and weldability [109–119] will be un-

dertaken. It has been reported so far that the partial replacement of Si by Al increases the hydrogen embrittlement resistance in the 0.2C-(0.5-1.5)Si-1.5Mn-(0.04-1.0)Al TBF steel [13]. In the future, further research on the effects of the partial replacement of Si by Al on these mechanical properties of the AHSSs is expected. In parallel with this, many researchers try to apply the AHSSs to automotive hot/warm/cold forging parts [81,120–124]. In this field, further research investigating the effects of a complex addition of Al and other alloying elements on mechanical properties such as tensile properties, fatigue strength, toughness, and wear resistance is hoped for in the future.

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Nomenclature

| AHSS | advanced high-strength steel | TRIP | transformation-induced plasticity | | | |
|----------------------------|---|-------------------------|--|--|--|--|
| TWIP | twin-induced plasticity | HMn TWIP | high Mn TWIP | | | |
| Aus. | austenitic | TBF | TRIP-aided bainitic ferrite | | | |
| Q&P | quenching and partitioning | CFB | carbide-free bainite | | | |
| D-MMn | duplex type medium Mn | L-MMn | laminate type medium Mn | | | |
| BF-MMn | bainitic ferrite type medium Mn | Q&P-MMn | Q&P type medium Mn | | | |
| TM | TRIP-aided martensite | M-MMn | martensite type medium Mn | | | |
| TPF | TRIP-aided polygonal ferrite | TAM | TRIP-aided annealed martensite | | | |
| DP | dual-phase | СР | complex phase | | | |
| Q&T | quenching and tempering | DQ&P | direct quenching and partitioning | | | |
| IT | isothermal transformation | M_s | martensite-start temperature | | | |
| M_{f} | martensite-finish temperature | T_{IT} | isothermal transformation temperature | | | |
| T_O | quenching temperature | T_P | partitioning temperature | | | |
| T_0^{\sim} | critical temperature at which austenite and martensite have the same chemical free energy | | | | | |
| γ _R | retained austenite | $\alpha_{\rm bf}$ | bainitic ferrite | | | |
| α _m | primary coarse soft martensite | $\alpha_{\rm m}^*$ | secondary fine hard martensite | | | |
| MA | MA $(\alpha_m^* + \gamma_R)$ phase | θ | carbide | | | |
| $f\gamma_0$ | initial volume fraction of $\gamma_{\rm R}$ | fγ | volume fraction of $\gamma_{\rm R}$ | | | |
| fα _{bf} | bainitic ferrite fraction | $f \alpha_{\rm m}$ | primary martensite fraction | | | |
| $f\alpha_{\rm m}^{*}$ | secondary martensite fraction | f _{MA} | MA phase fraction | | | |
| $f\alpha_{\rm m}'$ | $f\alpha_{\rm bf} + f\alpha_{\rm m}^{*}$ | fe | carbide fraction | | | |
| $C\gamma_0$ | initial carbon concentration of γ_R | ε | plastic strain | | | |
| $\Delta G^{\alpha'\gamma}$ | chemical free energy change for transformation of γ to α | $G^{\alpha'}$ | chemical free energy of ferrite (martensite) | | | |
| G^{γ} | chemical free energy of austenite | k | strain-induced transformation factor | | | |
| k_1 | modified <i>k</i> -value | SFE | stacking fault energy | | | |
| σ | flow stress of steel | $\sigma^{\mathbf{M}}$ | flow stress of matrix | | | |
| $\Delta \sigma_{\rm h}$ | strain hardening increment | $\Delta \sigma_{\rm i}$ | long-range internal stress | | | |
| $\Delta \sigma_{\rm t}$ | transformation hardening | $\Delta \sigma_{ m f}$ | forest dislocation hardening | | | |
| ν | Poisson's ratio | μ | Shear modulus | | | |
| f | volume fraction of second phase | ε_p^{u} | eigenstrain | | | |
| $\Delta f \alpha_{\rm m}$ | strain-induced martensite fraction | ζ | material constant | | | |
| b | Burgers vector | r | particle radius of second phase | | | |
| YS | yield stress | TS, UTS | tensile strength | | | |
| UEl | uniform elongation | TEl, TE | total elongation | | | |
| RA | reduction of area | TS×TEl | product of TS and TEl | | | |
| H_{max} | maximum stretch height | $TS \times H_{max}$ | product of TS and H_{max} | | | |
| HER | hole expansion ratio | TS×HER | product of TS and HER | | | |
| ss/t | shear section length to sheet thickness | R_{\min} | minimum bending radius | | | |

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