Topical Review

Recent developments and perspectives of advanced high-strength medium Mn steel: from material design to failure mechanisms

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Abstract
Advanced high-strength steels are key structural materials for the development of next-generation energy-efficient and environmentally friendly vehicles. Medium Mn steel, as one of the latest generation advanced high-strength steels, has attracted tremendous attentions over the past decade due to its excellent mechanical properties. Here, the state-of-the-art developments of medium Mn steel are systematically reviewed with focus on the following crucial aspects: (a) the alloy design strategies; (b) the thermomechanical processing routes for the optimizations of microstructure and mechanical properties; (c) the fracture mechanisms and toughening strategies; (d) the hydrogen embrittlement mechanisms and improvement strategies.

Keywords: medium Mn steel, alloy design, mechanical property, fracture mechanism, hydrogen embrittlement

1. Introduction
Driven by increasingly serious environmental issues, the development of next generation energy-efficient and environmentally friendly vehicles has become the trend in the automobile industry. To achieve this goal, one of the critical tasks is to develop strong and lightweight materials for making the automobile body-in-white. It has been shown that every 10% reduction in vehicle weight can generally reduce fuel consumption and emission by 6%–8% [1]. Currently, advanced high strength steels (AHSS) and aluminum alloys are the most promising lightweight structural materials for large-scale applications in vehicles. Compared to AHSS, aluminum alloys offer a larger room for weight reduction (up to \sim 30\%–40\%), yet the energy consumption and emission during the material manufacturing are much higher. Besides, the forming and welding of aluminum alloys are generally more

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Future perspectives

Advanced high-strength medium Mn steel (MMS) is a promising structural material for next-generation lightweight and energy-efficient vehicles. The metastable retained austenite with enriched Mn plays a critical role in the outstanding mechanical properties, which can be optimized by tailoring the volume fraction, mechanical stability, defect density, and grain morphology of retained austenite via various thermomechanical processing routes, such as intercritical annealing (IA), quenching and partitioning (Q&P), chemical patterning, warm rolling (WR), warm stamping (WS), deforming and partitioning (D&P), etc. The deformation-induced martensitic transformation (DIMT) of the retained austenite has both positive and negative effects on the mechanical properties: (a) DIMT can enhance work hardening rate and uniform elongation by postponing the onset of necking; (b) DIMT can toughen the material at the early stage of deformation by inhibiting crack initiation and growth, but it can also deteriorate the toughness at the later stage of deformation due to the brittle fresh martensite, which is susceptible to cleavage cracks; (c) stable austenite without DIMT can act as a strong hydrogen trapping site and crack blunting site, thereby enhancing the hydrogen embrittlement (HE) resistance, whereas the metastable austenite with DIMT, on the other hand, would worsen the HE resistance. Future development of high-performance MMS with simultaneously high strength, good ductility, high fracture toughness, and good HE resistance requires new strategies for maximizing the benefits of DIMT and minimizing its drawbacks.

challenging and expensive. In the automobile industry, AHSS are still the primary option for building the body-in-white nowadays, which provides the driving force for the related fundamental research.

Three generations of AHSS have been developed over the last four decades in order to meet the demands in the automotive industry (figure 1). The first generation AHSS was proposed in the 1980s, which includes interstitial-free steel [2], dual-phase (DP) steel [3–5], complex phase (CP) steel [6–8], martensitic (MART) steel [9–11], high strength low alloy (HSLA) steel [12–14] and conventional transformation-induced plasticity (TRIP) steel [15–17]. The microstructure of the first generation AHSS consists mainly of body-centered cubic (BCC) phases, such as ferrite, martensite or bainite, with a relatively small fraction (typically ≤10%) of face-centered cubic (FCC) phase, i.e. the austenite. The first generation AHSS typically exhibits an ultimate tensile strength (UTS) larger than 600 MPa, with a total elongation (TE) up to 25%, resulting in a product of strength and elongation (PSE) of ~10–20 GPa%.

Due to the excellent mechanical performance, the first generation AHSS has been widely used in vehicles, with DP steel being the flagship.

The second generation AHSS represented by the twinning-induced plasticity (TWIP) steel, emerged in the 1990s [18–20]. TWIP steels have a fully austenitic microstructure, which is metastable. Depending on the temperature, significant deformation twinning can occur during plastic deformation. The UTS of TWIP steels is ~600–1100 MPa, and TE is ~60%–95%, resulting in a PSE of ~50–80 GPa%, which is five times higher than that of the first generation AHSS. Despite the exceptional mechanical properties, the applications of TWIP steels are limited by the challenges in casting, welding and delayed fracture due to the high Mn and C contents.

In order to achieve superior mechanical properties, low cost and easy manufacturing properties, the development of the third generation AHSS began in the 2000s, which includes Q&P steel [21–23], MMS [24–26], and carbide-free bainitic (CFB) steel [27–29]. The microstructure of the third generation AHSS is typically composed of martensite/ferrite/bainite plus metastable retained austenite. The fraction of retained austenite is typically larger than 20%, which is produced by substantial partitioning of C and Mn into austenite during heat treatment. The large fraction of metastable retained austenite provides TRIP and/or TWIP effect during plastic deformation, which leads to the excellent strength and ductility combination. Typically, the UTS of the third generation AHSS is ~1000–1500 MPa, and TE is ~30%–60%, resulting in a PSE of ~30–60 GPa%.

As one of the most promising third generation AHSS, the MMS, with a Mn content of ~3–12 wt.%, has received tremendous attentions over the last decade. Here, we reviewed the state-of-the-art knowledge of MMS, covering aspects of alloy design strategies, thermomechanical processing, microstructure, deformation behavior, fracture mechanisms, and HE.

2. Alloy design

Numerous studies have demonstrated that the excellent combination of high strength and good ductility in MMS is attributed to the metastable retained austenite. Therefore, the design of a strong and ductile MMS in many cases, is focused on the following two crucial questions: (a) how to retain austenite in the final microstructure; (b) how to control its fraction and stability for providing prolonged TRIP and/or TWIP effects during deformation.

In general, the major alloying elements in MMS are C, Mn and Al, while Si, Nb, V, Ti, and B are usually added as minor alloying elements. Mn is used primarily for stabilizing austenite, with a typical concentration of ~3–12 wt.%. The increase of Mn concentration leads to an expansion of the α/γ

![Figure 1. Total elongation versus ultimate tensile strength of various AHSS grades.](image-url)
Figure 2. (a) Equilibrium austenite fraction in 0.1C–1.5Mn and 0.1C–6Mn steels overlapped with recrystallization fraction of cold-rolled microstructure; and (b) influence of Al and Si on austenite fraction. Reprinted from [30]. Copyright (2017), with permission from Elsevier (c) Composition-dependent stacking fault energy (SFE) maps of Fe–C–Mn system and (d) Fe–C–Mn–1.5Al system. Reproduced from [31]. CC BY 4.0.

two-phase region and a decrease of the austenitization temperature, as shown in figure 2(a). The presence of a large two-phase region facilitates the control of the volume fraction of austenite via intercritical annealing (IA) at a specific temperature within the two-phase region. Lower austenitization temperature is beneficial for reducing surface oxidation during thermomechanical processing. During intercritical annealing, ferrite recrystallization and austenite reverse transformation (ART) occur, resulting in an ultrafine-grained duplex microstructure. The ferrite recrystallization kinetics of a cold-rolled (CR) 0.1C–1.5Mn (wt.%) steel at a heating rate of 5 °C s$^{-1}$ is shown by the dashed line in figures 2(a) and (b), which indicates that ferrite recrystallization can occur before or during the ART, depending on the concentration of Mn and Al [30]. Due to the relatively high annealing temperature, there is a strong Mn partition from ferrite to austenite, which further stabilizes austenite and results in a substantial amount of retained austenite at room temperature. The ferrite/austenite duplex microstructure does not guarantee good mechanical properties. The stability of austenite, which can be evaluated by the stacking fault energy (SFE), needs to be tuned carefully for activating the TRIP and/or TWIP effects during deformation. It is demonstrated in figures 2(c) and (d) that SFE increases with the concentrations of Mn, Al, and C, which results in the transition of the deformation mechanism from TRIP to TWIP [31].

Aluminum in MMS to some extent, plays the opposite role compared to Mn. Al increases the stability of ferrite, while reducing that of austenite, leading to the enlargement of the ferrite region and the narrowing of the austenite region, as illustrated in figure 2(b). In other words, Al has significant influence on the volume fraction and stability of ferrite, which is crucial for the ductility of MMS. Besides, Al also influences the SFE of austenite, thereby affecting the hardening mechanisms, as shown in figures 2(c) and (d). It is noted that a high Al concentration (e.g. ≥6 wt.%) leads to the formation of δ-ferrite, which may deteriorate the mechanical properties such as fracture toughness and tensile elongation. Therefore, the concentrations of Al and Mn need to be considered together for optimizing the volume fraction and stability of retained austenite. Numerous studies have demonstrated that a Mn/Al concentration ratio between 1.5 and 4 is usually a useful index for optimizing the mechanical properties.

Carbon is the most important interstitial solid solution element in MMS for effectively enhancing the stability of austenite and thus decreasing the martensite start ($M_s$) temperature. According to literature [32], the $M_s$ temperature has an empirical relation with the alloying contents as: $M_s$ ($°C$) = 539−423C−30.4Mn−7.5Si+30Al (in wt.%). The increase of C by 1 wt.% results in the decrease of $M_s$ by 423 °C. The C concentration also affects the SFE of austenite and the resultant occurrence of TRIP and/or TWIP effects during deformation, as shown in figures 2(c) and (d). It is noted that an excessive C concentration may lead to the formation of large cementite during casting and thermomechanical processing, which deteriorates mechanical properties. Moreover, increasing the C concentration affects the weldability. In short,
the C concentration needs to be chosen carefully to balance all the aforementioned influences. Typically, the C concentration in MMS is ~0.05–0.6 wt.\%.

The high contents of C and Mn will degrade the weldability of MMS, particularly when using conventional fusion welding (FW) techniques such as inert gas welding [33, 34], laser welding [35], and resistance spot welding [36, 37], which will result in a brittle fusion zone that deteriorates the mechanical properties. The conventional FW process involves a high temperature at fusion zone that typically exceeds 1000 °C (i.e. the austenitic single-phase region). During the rapid air cooling of the austenitic fusion zone, martensitic transformation occurs, along with Mn segregation at the martensite laths. The fusion zone with martensite microstructure is extremely brittle due to the high carbon content and Mn segregation. Comparing to the conventional FW techniques, the state-of-the-art friction stir welding (FSW) technique has been reported to be more suitable for MMS due to the relatively low welding temperature that is typically below 800 °C (i.e. the ferrite-austenite two-phase region) [38, 39]. To enhance the mechanical properties of the welded MMS, a post welding heat treatment (e.g. annealing at 630 °C for 30 min) is usually performed [34, 39].

Overall, in order to improve the weldability of MMS, the C and Mn contents should be controlled carefully, and the FSW can be used.

Generally, Si, Nb, V, Ti are added as the minor alloying elements in some MMS. Similar to Al, Si is also a stabilizing element of ferrite. In addition, Al and Si both increase the carbide formation temperature, and thus suppress the formation of carbide during thermomechanical processing. The microalloying of Nb, V, Ti leads to nanoscale precipitation, which can increase the strength and improve the resistance to HE.

3. Thermomechanical processing, microstructure and mechanical properties

In addition to alloy design, a proper thermomechanical processing route is also crucial for producing the desired microstructure and thus the optimal mechanical properties. In this section, we summarized five thermomechanical processing routes (figure 3) for fabricating high-performance MMS, along with the associated microstructures and mechanical properties.

3.1. Intercritical annealing

IA, or so-called ART treatment, is the most widely-used thermomechanical processing route for MMS. During the IA process, the steel is annealed at a temperature in the ferrite-austenite two-phase region (~600 °C–900 °C), i.e. between Ac1 and Ac3. Recrystallization of ferrite and ART occur simultaneously. Besides, there is substantial diffusion of Mn and C from ferrite to austenite during IA, which enhances the stability of austenite so that it can be retained at room temperature. The microstructure prior to IA, the temperature and duration of IA process, all have significant influence on the final microstructure and mechanical properties of MMS. Here, we summarized three different IA-based processing routes that have been successfully used in previous studies for fabricating high-performance MMS, as shown in figure 3(a) [40, 41]. For the first two IA processing routes, the steel is either in the hot-rolled (HR) or cold-rolled (CR) states before the prolonged IA process, which is denoted as HR-IA and CR-IA, respectively. For the third IA processing route, the CR steel is subjected to a flash annealing (FA) process, which is denoted as CR-FA. The microstructures and mechanical properties of MMS produced by HR-IA, CR-IA and CR-FA processing routes are summarized in figure 4. Specifically, the HR-IA MMS has a composition of 0.2C–7Mn–3Al (wt.%). The HR steel was intercritically annealed at 700 °C for 1 h, followed by water quenching to room temperature. The HR-IA MMS exhibits a duplex lamellar microstructure with tempered martensite (α′–γ) and retained austenite (γR), as shown in figure 4(a). This lamellar morphology is the result of active ART but absence of ferrite recrystallization during the HR-IA process. Besides, substantial diffusion of Mn and Al takes place during HR-IA. As illustrated in figure 4(a4), the austenite shows dramatically higher Mn concentration and lower Al concentration than the tempered martensite. The CR-IA and CR-FA MMS have a composition of 0.23C–5.38Mn–1.7Al (wt.%). For the CR-IA MMS, the CR steel was conventionally intercritically annealed at 660 °C for 1 h and water quenched to room temperature. For the CR-FA MMS, the CR steel was flash-heated to 760 °C and then directly quenched to room temperature. The CR-IA MMS exhibits a duplex equiaxed microstructure with recrystallized ferrite (α) and austenite (γ), as shown in figure 4(b). This implies that CR accelerates the recrystallization kinetics, compared with the microstructure of HR-IA MMS. The diffusion of Mn and Al during CR-IA is similar to that of HR-IA, shown in figure 4(b4). The CR-FA MMS has a heterogeneous microstructure with partially recrystallized ferrite and retained austenite, as shown in figure 4(c). The non-recrystallized ferrite with lamellar morphology and recrystallized ferrite with equiaxed morphology can be observed in figure 4(c2). The diffusion of Mn and Al is similar to that of HR-IA and CR-IA. Figure 4(d) displays the engineering stress–strain curves of HR-IA and CR-FA MMS with composition of 0.2C–7Mn–3Al (wt.%). Figure 4(e) show the engineering stress–strain curves of CR-IA and CR-FA MMS with a composition of 0.23C–5.38Mn–1.7Al (wt.%). It is found that the CR-FA MMS exhibits significantly increased yield and UTS yet negligible ductility loss, in comparison to the CR-IA and CR-FA MMS. This extraordinary strength-ductility synergy in the CR-FA MMS is attributed to the heterogeneous microstructure with partial recrystallized ferrite grains, leading to the pronounced heterogeneous deformation-induced (HDI) hardening, as shown in figure 4(f).

3.2. Quenching and partitioning

Although the MMS produced by IA process exhibits high UTS and good ductility, the yield strength (YS) is relatively low, due to the presence of large fraction of recrystallized and soft ferrite. In this case, Q&P treatment was proposed to improve the YS of MMS by forming strong martensite during quenching. The schematic Q&P processing route is illustrated...
Figure 3. Thermomechanical processing routes of high-performance MMS. (a) Intercritical annealing; (b) quenching and partitioning; (c) chemical patterning; (d) warm rolling and warm stamping; (e) deforming and partitioning.

in figure 3(b). The HR steel is first subjected to austenitization at a temperature slightly above Ac$_3$, followed by quenching to a temperature between the M$_s$ and martensite finish (M$_f$) temperatures. The microstructure after quenching consists of martensite and retained austenite. The quenched steel is then tempered at a relatively low temperature (250 °C–500 °C) for relieving the residual stress of the as-quenched martensite and promoting the carbon partitioning from martensite to austenite. It is noted that, due to the relatively high concentration of Mn, the Ac$_3$, M$_s$ and M$_f$ of MMS are all lower than the first generation AHSS, which facilitates the Q&P processing route. For example, the MMS with a composition of 0.2C–10Mn–2Al–0.1V (wt.%) has a Ac$_3$ of ~760 °C, a M$_s$ of ~132 °C, and a M$_f$ below the room temperature [42, 43]. Thus, this MMS can be directly quenched to room temperature after austenitization at 850 °C, while still retaining significant amount of austenite. Then the partitioning is conducted at 300 °C for 10 min. This novel processing route of MMS is referred as room temperature Q&P (RT-Q&P), which is a cost-effective heat treatment for producing MMS with high YS, high work hardening, and good ductility. The microstructure of this RT-Q&P MMS is composed of lath martensite matrix ($\alpha'$) and blocky retained austenite ($\gamma$), with the prior austenite grain boundary (PAGB) denoted by the dashed line, as shown in figures 5(a)–(c). The engineering stress–strain curve of this RT-Q&P MMS is shown in figure 5(d), demonstrating that this RT-Q&P steel exhibits exceptional tensile properties, in comparison to the commercial dual phase steel and press hardening steel.

3.3. Chemical patterning

Sun et al [45, 46] proposed a novel thermomechanical process called chemical patterning, to control the stability of retained austenite and the distribution of stability in MMS. This novel process is intended to create a chemical pattern of Mn in austenite by the flash re-austenitization of a pearlite microstructure. The detailed processing route is shown in figure 3(c). The HR steel is first subjected to austenitization, followed by cooling to a temperature between 500 °C and 600 °C for pearlite transformation, during which lamellar cementite with significant Mn enrichment is formed, as shown in figures 6(a) and (c). Then the steel is subjected to flash re-austenitization for a short time, to obtain a fully austenitic microstructure with a chemical patterning of Mn that results from its previous enrichment in cementite and the insufficient time for homogenization during the flash re-austenitization process. After cooled down to room temperature, this steel forms a lamellar
Figure 4. Microstructures and mechanical properties of MMS produced by different IA processes. Microstructure of (a) HR-IA, (b) CR-IA, (c) CR-FA, characterized by scanning electron microscope (SEM) (a1), (b1), (c1), electron backscatter diffraction (EBSD) phase mapping (a2), (b2), (c2), transmission electron microscope (TEM) (a3), (b3), (c3), and energy dispersive x-ray spectroscopy (EDXS) line profiling (a4), (b4), (c4). Note here the HR-IA MMS has a composition of 0.2C–7Mn–3Al (wt.%), and the CR-IA and CR-FA MMS have a composition of 0.23C–5.38Mn–1.7Al (wt.%). (d) Engineering strain-stress curves of HR-IA and CR-FA MMS with composition of 0.2C–7Mn–3Al (wt.%). (e) Engineering strain-stress curves and (f) HDI stress with increasing true strain of CR-IA and CR-FA MMS with composition of 0.23C–5.38Mn–1.7Al (wt.%). Reprinted from [40, 41]. Copyright (2021), with permission from Elsevier.
microstructure of martensite (with low Mn) and retained austenite (with high Mn) due to the chemical patterning of Mn, as shown in figures 6(b) and (d). Finally, a low temperature tempering is performed to relieve the residual stress in martensite. The engineering stress–strain curves and the corresponding true strain hardening rate curves of the initial pearlite MMS and the chemically patterned MMS are shown in figures 6(e) and (f). The chemically patterned MMS exhibits an UTS of 1600–2000 MPa and a TE of 7%–10%. This novel thermomechanical and microstructure design concept can also be used for steels with relatively low concentrations of Mn and C, such as conventional Q&P steel [47]. In this case, instead of forming a fully pearlitic microstructure, a hypo-eutectoid ferrite and pearlite microstructure is generated, which can still be used to create the chemical patterning of Mn.

3.4. Warm rolling and warm stamping

A WR process has been introduced for producing MMS with improved yield and UTS while maintaining the uniform elongation [48–50]. As shown in figure 3(d), the HR steel is first heated to the intercritical temperature region (~600 °C–900 °C), and then subject to a WR process. Note that the finishing temperature of WR is typically higher than 300 °C. In some cases, an additional annealing or tempering treatment is performed after the WR process [50, 51]. Nevertheless, it has been shown that the additional annealing or tempering process has negligible influence on the mechanical properties and can therefore be removed [48, 52]. During the WR process, a heterogeneous microstructure with lamellar and granular ferrite and retained austenite grains are formed via partial recrystallization [53–55]. The heterogeneous retained austenite grains with various mechanical stability lead to a gradual and steady martensitic transformation (i.e. TRIP effect) throughout the plastic deformation, resulting in excellent work hardening ability and ductility [50]. In addition to the TRIP hardening, the heterogeneous microstructure also leads to more significant deformation incompatibility and thus enhanced generation of geometrically necessary dislocations (GNDs), which also contributes to the strain hardening [55]. Further, the WR process also introduces a high density of dislocations into both ferrite and austenite [43, 49]. The higher dislocation density improves the YS, the austenite stability as well as the TRIP hardening. For example, figures 7(a)

Figure 5. Microstructure and mechanical property of MMS produced by Q&P process. (a) EBSD phase map and (b) EBSD inverse pole figure (IPF) map (c) SEM image of a typical RT-Q&P MMS with a composition of 0.2C–10Mn–2Al–0.1V (wt.%). Reproduced from [44]. CC BY 4.0. (d) Engineering stress–strain curves of the present RT-Q&P MMS, commercial dual phase steel and press hardening steel. Reproduced from [42], with permission from Springer Nature.
Figure 6. Microstructures and mechanical properties of MMS produced by chemical patterning process. (a) TEM image of the initial pearlite. (b) TEM image of the chemical patterned MMS. (c) EDXS line profile of Mn distribution of the initial pearlite. (d) EDXS line profile of Mn distribution of the chemical patterned MMS. (e) Engineering stress–strain curves and (f) corresponding true strain hardening rate curves of the initial pearlite steel and chemical patterned MMS with different processing parameters. Note that this MMS has a composition of 0.51C–4.35Mn (wt.%). Reprinted from [45]. Copyright (2018), with permission from Elsevier.

and (b) show the microstructures of the WR 0% (without WR) and WR 89% MMS with a composition of 0.22C–4.88Mn–3.11Al–0.62Si (wt.%). In comparison to the WR 0% MMS, the WR 89% MMS exhibits a more heterogeneous microstructure with lamellar and granular ferrite grains and austenite grains with high dislocation densities, resulting in a much higher yield and UTS but marginal ductility loss, as shown in figure 7(d).

WS is another novel process that is similar to WR and hot stamping (or hot press hardening) [56, 57], as shown in figure 3(d). WS is a promising process that can be used for large-scale production of high-strength automobile components such as B-pillar based on the currently-available hot stamping facilities [58, 59]. In comparison to hot stamping, the WS introduces more heterogeneous microstructure and significantly enhances the mechanical properties, similar to the
advantages of WR [60, 61]. For example, the microstructure of a WS MMS with a composition of 0.25C–6.94Mn–0.23Al–0.38Si–3.29Cr–0.12Nb (wt.%) consists of martensite, retained austenite and dispersive carbide (figure 7(c)). The engineering stress–strain curves of the baked and non-baked WS MMS are shown in figure 7(e), demonstrating that the UTS, work hardening ability and ductility of the WS MMS are simultaneously improved, in comparison to the commercial 22MnB5 hot stamping steel.

### 3.5. Deforming and partitioning

A novel thermomechanical process called D&P has been developed for fabricating MMS with ultrahigh strength, large ductility, and excellent fracture toughness [62, 63]. As shown in figure 3(e), the D&P process is based on the WR process, with an additional CR process and a partitioning treatment afterward. The deforming processes, including WR and CR, are designed to generate an ultrahigh density of dislocations, via severely plastic deformation and martensitic transformation. The partitioning treatment is used to relieve the residual stress in fresh martensite and stabilize the retained austenite via carbon enrichment. This novel processing route is inspired by the dislocation engineering concept [64]. Specifically, the ultrahigh density of dislocations provides a tremendous number of both plasticity carriers and gliding obstacles, resulting in the large ductility and ultrahigh strength, respectively. The effects of processing parameters on the mechanical properties of the D&P steels have been systematically investigated in [52, 65]. The WR process is critical for generating high density of mobile dislocations to ensure the high ductility. The IA treatment has negligible influence on the microstructure and mechanical properties and is thus unnecessary. The CR process is crucial for acquiring high strength, and the thickness reduction of CR is critical for the strength-ductility synergy of D&P steel. The partitioning process is important for achieving large uniform elongation in the D&P steel. Here, a D&P MMS with a composition of 0.47C–10Mn–2Al–0.7V (wt.%) is used as an example for illustration. This
D&P steel has a heterogeneous lamellar duplex microstructure with retained austenite embedded in a martensite matrix, as shown in figures 8(a)–(e). The dislocation density of martensite matrix was estimated to be $\sim 1.27 \times 10^{16} \text{m}^{-2}$, which is two orders of magnitude higher than that of the conventional thermally transformed martensite. This ultrahigh dislocation density of martensite was also confirmed by the formation of dislocation cells, as shown in figure 8(c). The D&P steel exhibits an ultrahigh YS of 2.2 GPa and a large uniform elongation of 16%, which evades the strength-ductility trade-off and surpasses most of existing high-strength alloys, as shown in figures 8(f) and (g).

3.6. Perspectives of new thermomechanical processing routes

By comparing and summarizing the aforementioned thermomechanical processing routes (figure 3), including IA, Q&P, chemical patterning, WR, WS, and D&P, here we propose the following three design strategies for producing high-performance MMSs. (a) Tailoring the volume fraction and stability of retained austenite, such that to best utilize the TRIP effect to improve the mechanical properties of MMSs. This strategy includes IA, Q&P, and chemical patterning processing routes. (b) Tailoring the dislocation density and morphology...
or the so-called dislocation engineering. High density of dislocations not only enhance the strength, but also stabilize the retained austenite of MMSs. This strategy includes WR, WS, and D&P processing routes. (c) Tailoring the microstructure heterogeneity, including morphology heterogeneity and chemical heterogeneity, which are significantly beneficial to the mechanical properties. The morphology heterogeneity can be produced by CR-FA, WR, WS, and D&P processing routes. The chemical heterogeneity can be produced by chemical patterning route. It has also been reported that chemical heterogeneity can be easily obtained by additive manufacturing, such as laser powder bed fusion [66, 67] and electro-deposition [68].

By combining and optimizing two or three design strategies, new thermomechanical processing routes can be proposed for developing novel high-performance MMSs. For example, a MMS with chemical heterogeneity is first fabricated via chemical patterning or via additive manufacturing, and then the steel is subjected to WR, or WS, or D&P to produce morphology heterogeneity and high density of dislocations, along with large fraction of retained austenite. Another feasible way is to combine chemical patterning and IA or Q&P, which would efficiently enhance the stability and fraction of retained austenite. More novel thermomechanical processing routes can be proposed by taking advantage of different design strategies in the future.

4. Fracture mechanisms

4.1. Quantitative measurement of fracture toughness

The resistance to fracture of MMS is crucial for its applications in vehicles concerning the forming of car body components and the safety during car crash. Although the tensile properties of MMS are extensively investigated, it is noted that they cannot be directly to evaluate the fracture toughness. The proposal that PSE represents the toughness in certain cases is questionable. In fact, a steel with superior tensile properties can have relatively low toughness and vice versa, which is elaborated in the discussion of a high-strength Q&P steel [23].

Up to now, there are still challenges in the quantitative measurement of fracture toughness. The two main approaches are the Charpy impact test and the J-integral resistance curve (J-R curve). The Charpy impact test aims to measure the energy absorbed by the material during impact (i.e. the impact toughness) using notched samples with a standard size (e.g. $55 \times 10 \times 10 \text{ mm}^3$). When the samples are subjected to impact loadings, transverse stresses develop due to the existence of notch [69]. This approach finds its wide application since it is time and cost efficient and can be conducted conveniently at various temperatures. However, the high strain rate during impact renders it challenging to separate the initiation and growth of cracks, as well as to perform interrupted tests. Besides, the absorbed energy is an empirical quantity that is difficult to analyze using fracture mechanics.

In contrast, the resistance curve (R curve) method is capable of measuring the quasi-static fracture properties of elastic (K-R curve) or elastic-plastic materials (J-R curve) quantitatively, which provides important information such as tearing modulus, crack initiation and growth toughness [70]. Precracking is required for obtaining an ideally sharp crack that fulfills the requirements of fracture mechanics. A primary advantage of the R curve method is that the size-independent plane strain crack initiation toughness values can be obtained once specific conditions (e.g. straight crack fronts) are satisfied. This parameter can be regarded as an intrinsic property of materials. The R curve method however, has the drawbacks of the laborious pre-cracking process and the requirement to mitigate the crack-tunneling effect. It is noted that a general relationship between the toughness from the Charpy impact test and the R curve test has not yet been derived despite several attempts [71, 72].

4.2. Fracture modes

As a result of various chemical compositions and phases with different mechanical properties, all three common fracture modes, namely the ductile, cleavage and intergranular fracture, have been observed in MMS.

Ductile fracture is featured by void formation and coalescence. In traditional HSLA steels and some single-phase steels, voids usually initiate near precipitates or impurity inclusions due to local stress concentration [73]. For steels with multiple phases such as MMS, DP and Q&P steels, a high density of GNDs are generated near phase boundaries during deformation due to plastic incompatibility, which can easily lead to void formation. For example, the MMS with a chemical composition of 0.2C–10.4Mn–2.9Al (wt.%) produced by CR-IA processing route, has an ultrafine-grained microstructure of ferrite and metastable retained austenite [74]. The fracture surfaces of tensile samples showed that the number of voids formed at the ferrite/martensite ($\alpha'/\alpha$) interfaces is at least five times higher than that inside ferrite and martensite grains, as shown in figure 9, indicating the interphase decohesion is the main reason for voids initiation.

Brittle cleavage fracture of steels is defined as the separation of specific grain surfaces; in other words, the crack grows along specific crystallographic planes [75]. Though the crack propagates through the grains interior, minimal plasticity is involved during the process, leading to lower energy absorption and thus fracture toughness. Cleavage is observed less commonly in austenite than in fresh martensite and ferrite, especially $\delta$-ferrite. In the cases of martensite and ferrite, cleavage usually occurs on the (100) plane due to the low packing density [76]. In recent years, lightweight high-strength automobile steels with excess Al and Si concentration have drawn more attention. It has been reported that the addition of 1 wt.% of Al and Si can reduce the density by 1.3% and 0.8%, respectively [77]. However, the high contents of Al and Si usually deteriorate the mechanical properties due to the formation of coarse-grained $\delta$-ferrite that is susceptible to cleavage fracture. For example, Choi et al observed cleavage crack in $\delta$-ferrite in an MMS with 3% Al and 1.5% Si even though the $\delta$-ferrite is the softest phase in this steel [78], as presented in figure 10.
Figure 9. SEM observations near the fracture surface of the tensile samples, showing the void formation at different sites (marked by red elliptical frames) in an MMS with a chemical composition of 0.2C–10.4Mn–2.9Al (wt.%) produced by CR-IA processing route. IA temperature: (a)–(c) 700 °C, (d), (e) 750 °C. Reprinted from [74]. Copyright (2019), with permission from Elsevier.

Figure 10. EBSD IPF maps of the fractured samples showing the long cleavage cracks in coarse-grained δ-ferrite (indicated by white arrows) in an MMS with a chemical composition of 0.3C–6Mn–3Al–1.5Si (wt.%) produced by HR-IA processing route. IA temperature: (a) 760 °C, (b) 800 °C, (c) 840 °C, (d) 880 °C. Reprinted from [78]. Copyright (2017), with permission from Elsevier.
Martensite, one of the strongest phases in steels, is usually utilized to enhance the strength of MMS, by some processing routes, such as Q&P and D&P. However, martensite with high carbon content, or at cryogenic temperature, is susceptible to cleavage fracture due to the heavily distorted lattice. For example, Maede et al. [79] reported an MMS with a chemical composition of 0.1C–5Mn (wt.%) produced by two processing routes: (a) homogenization and water quenching (AsQ sample); and (b) warm rolling and water quenching (RQ sample). It was found that both MMS samples were dominated by cleavage fracture when subjected to Charpy impact tests at 77 K, as shown in figures 11(a) and (b). This brittle cleavage fracture mode leads to an extremely low impact toughness (less than 5 J cm\(^{-2}\)) [79]. The EBSD IPF maps of the cross-section of the fracture surfaces of both MMS samples after impact fracture at 77 K, confirm that the cleavage cracks grow parallel to the \{100\} planes of the martensite packets, as shown in figures 11(c) and (d). The white arrows indicate the \(<100>\) orientations of each packet. In addition, the twin boundaries also contribute to the cleavage fracture [80]. Even though twinning occurs less commonly in MMS than in high alloy steels due to its low SFE, the local chemical segregations may lead to the activation of twinning in some cases.

Intergranular fracture can also occur in MMS, especially when there is intergranular precipitation or segregation of solute atoms at grain boundaries, which will deteriorate the cohesion of grain boundaries. High stress-triaxiality has also been observed to induce intergranular fracture in Q&P steel as a result of deteriorated austenite stability and high maximum principal stress [81]. Of particular importance is the ductile-to-brittle transition phenomena that occurs at cryogenic temperatures. Low temperature will enhance the YS, reduce the grain boundary cohesion and promote the possibility of intergranular fracture. For example, an MMS with a chemical
composition of 0.2C–9.4Mn–2.1Al (wt.%) and an ultrafine-grained microstructure of ferrite and austenite, fails by microvoid formation during Charpy impact test at room temperature, while intergranular fracture dominates at $-90^\circ$C, as presented in figure 12. As a result, the absorbed energy at $-90^\circ$C is less than 10% of that at room temperature.

4.3. Toughening mechanisms

Similar to the strength-ductility trade-off, the intrinsic conflict between strength and toughness is also one of the biggest challenges for alloy design. The strengthening of steels is generally achieved by blocking or inhibiting dislocation movement with defects at different length scales [62, 82], which inevitably causes strain concentration and thus microcrack formation once dislocations are entangled. The Ashby maps of YS versus fracture toughness or impact toughness shown in figure 13 clearly demonstrate the strength-toughness trade-off [62, 83].

Many promising approaches have been proposed to achieve an excellent synergy of strength and toughness [84, 85]. Among these approaches, tailoring the fraction and stability of metastable retained austenite (i.e. engineering the martensitic transformation or TRIP effect) has been shown to be a feasible and effective way. The effects of martensitic transformation on the toughness should be similar to that in low-alloy steels, except that MMS usually contains a higher fraction of retained austenite. The beneficial effects of martensitic transformation can be explained by: (a) the additional energy absorbed when metastable austenite transforms into martensite at the crack tip [86]; and (b) the volume expansion during martensitic transformation that leads to compressive stress for suppressing crack initiation and growth [87]. For example, Antolovich and Singh [86] estimated that martensitic transformation can contribute up to 39%–60% of fracture toughness.

However, martensitic transformation can also be detrimental for fracture toughness. Such opposite effects are due to...
to the brittleness of martensite, which becomes more profound with a higher carbon content or a larger austenite volume fraction. For example, Jacques et al reported that TRIP effect can improve ductility by suppressing plastic localization, yet it deteriorates fracture toughness since premature cracking frequently occurs in fresh martensite and voids can easily initiate at martensite/martensite and martensite/ferrite interfaces, as shown in figures 14(b)–(d) [88, 89]. Additionally, figure 14(a) indicates that the crack path follows the network of second-phases that percolate along the grain boundaries of ferrite. Wu et al calculated the influence of TRIP effect on toughness and found that the contribution became negative if the carbon concentration in austenite was higher than 1.1 wt.% [90]. The conflicting effects of retained austenite on fracture toughness emphasize the importance of tailoring its volume fraction, morphology and mechanical stability in MMS so that the beneficial contribution can be fully exploited while minimizing the deleterious effects [88].

In addition to the TRIP effect, delamination can also be employed to enhance static crack initiation toughness and impact toughness. Some interfaces in materials, such as PAGB, are relatively weak due to the chemical segregation or heterogeneous deformation, and thus have a higher probability of separating and forming secondary cracks upon loading. The beneficial effect of delamination is three-fold: (a) it can effectively blunt crack tip or deflect crack path, thereby inhibiting its further growth; (b) it can change the stress state of the main crack from plane strain to plane stress with the aid of free surfaces created by delamination; (c) it takes additional energy to form new surfaces. For example, Liu et al reported that the MMS produced by D&P processing route achieves an ultrahigh YS of 2.2 GPa while also possessing a high fracture initiation toughness ($K_{IC}$) of 101 MPa√m [63]. This exceptional toughness mainly originates from the high-strength induced multi-delamination at weakened PAGBs due to Mn segregation, which considerably improves fracture toughness by developing plane stress state near the main crack, as represented in figure 15. Zou et al reported a low-carbon MMS produced by non-recrystallization WR, which possesses a high upper shelf energy (>450 J) and low-temperature impact toughness (∼105 J at −196 °C) without sacrificing tensile strength (1145 MPa) [91]. It was proposed that elongated laminated microstructure as well as the cleavage planes lead to the occurrence of delamination, which improved the toughness considerably. Besides, the TRIP effect is also considered to play a role in the enhancement of fracture toughness, especially at cryogenic temperature [63, 91].

5. Hydrogen embrittlement
HE refers to the phenomena that materials lose load-bearing capacities or damage resistances abruptly under the influence of hydrogen. HE was first reported in iron and steel in year 1875 [92], while thence found in Ti alloys, Al alloys, super-alloys, etc [93–96]. HE is particularly dangerous and often causes catastrophic failure of engineering structures without any sign because of its unavoidable and unpredictable invasion.

Figure 14. SEM micrographs taken close to the crack tip of TRIP steel; (a) crack propagation path; (b) decohesion between two martensite grains; (c) decohesion between ferrite and martensite grains; and (d) cleavage of a martensite grain. Reprinted from [89]. Copyright (2008), with permission from Elsevier.
Figure 15. Delamination toughening mechanism of MMS steel with a composition of 0.44C–9.95Mn–1.87Al–0.67V (wt.%) produced by D&P processing route. (a) Schematic diagram of fractured C(T) sample with extensive long and short delamination cracks. (b) and (c) Images of the fracture surface of C(T) sample showing numerous thin-layer delamination cracks. (d) and (e) SEM image and schematic diagram of the cross-section of the fracture sample showing the development of delamination cracks along the elongated PAGBs. The blue arrows indicate short delamination cracks, the black arrows indicate the long delamination cracks, the pink arrows indicate the secondary delamination cracks that usually observed in the vicinity of long delamination cracks. From [63]. Reprinted with permission from AAAS.

into materials during manufacturing and service [97–99]. It has become a significant concern in AHSS with increased strength over the last few decades [100, 101]. As the 3rd generation AHSS, MMS are particularly prone to HE essentially due to the complex microstructures and deformation behaviors [102, 103]. In MMS, multiple phases, including ferrite, martensite and retained austenite, have different intrinsic strength and toughness [58, 104, 105], leading to deformation incompatibility [55]. The H diffusion rate and concentration are distinct in different phases, which would lead to various H distribution and interaction with microstructures [106]. During deformation, microstructure evolution, such as
DIMT, deformation twinning, and deformation incompatibility can alter H distribution and H-microstructure interaction, thus changing the predominant HE mechanism and significantly influencing the mechanical response [107–110]. Thus, the fundamental mechanisms of HE in MMS are complicated and have not yet been fully understood. Disagreements remain among existing models [100, 103], including the hydrogen-enhanced decohesion (HEDE), hydrogen-enhanced localized plasticity (HELP), hydrogen-enhanced strain-induced vacancies and adsorption-induced dislocation emission (AIDE), etc. The interplay of these HE models has been directly observed by experiments and numerically verified by simulations, even though a unified theory is still lacking [111–115].

5.1. Effects of microstructures on H distribution and HE characteristics

Retained austenite is of great importance in MMS due to its exceptional TRIP effect for enhancing the work hardening, whilst it has two orders of magnitude higher H solubility than that of ferrite or martensite [116, 117]. Therefore, retained austenite can serve as an H trapping site during H diffusion and uptake. In particular, the desorption energy obtained from TDS indicates retained austenite is a strong reversible H trapping site, i.e. with a typical value of 48 kJ mol$^{-1}$ [116]. Ryu et al [116] conducted pre-strain on a 0.12C–4.6Mn–0.55Si–1.1Al (wt.%, L–Al) MMS and a 0.12C–5.8Mn–0.47Si–3.1Al (wt.%, H–Al) MMS, and investigated H desorption by TDS. As figures 16(a) and (b) shows, the spectrum can be decomposed into two peaks associated with austenite and ferrite-defects. After pre-straining, the austenite fraction decreases and the defect density in ferrite increases, resulting in the peak shifting in figures 16(c) and (d). Moreover, figures 16(e) and (f) validates that pre-strain after pre-charge can lead to H redistribution due to the change of phase fraction. As a trapping site, austenite can enhance the HE resistance of materials. It is reported that HE sensitivity decreases with an increasing austenite fraction in a series of quenched and annealed low carbon MMS, where the austenite is stabilized by Mn segregation [109].

H diffusion in FCC austenite phase is around two orders of magnitude slower than that in BCC ferrite and martensite phases [118]. This conspicuous contrast can lead to vast variations in H diffusion in MMS with multiphase microstructure [108, 117]. Besides, studies also show the effect of austenite grain morphology on H diffusion, either in MMS [108] or in other material [119]. Moreover, the percolation of austenite phase also plays a key role on the H distribution and concomitant HE phenomena of MMS. For example, Sun et al [117] studied the effects of phase fraction and their percolation on the H migration and trapping behavior in an MMS with a composition of 0.2C–10.2Mn–2.8Al–1Si (wt.%) fabricated by low-temperature (IA700) and high-temperature (IA800) IA processing. The IA700 sample has a ferrite matrix ($\sim$74 vol.%) with embedded austenite, whereas the IA800 sample has an austenite matrix ($\sim$59 vol.%) with embedded ferrite. As illustrated in figure 17, less H are trapped in austenite and austenite/ferrite boundaries in the IA700 MMS as the tunnels of ferrite matrix with a high dislocation density act as highway for H diffusion. In contrast, most H are trapped in austenite and austenite/ferrite boundaries in the IA800 MMS. The different H distributions result in different HE characteristics and corresponding mechanisms. The IA700 MMS fractures due to H-enhanced localized plasticity in high dislocated ferrite (i.e. HELP mechanism), while the IA800 MMS is embrittled by boundary decohesion (i.e. HEDE mechanism).

Carbides and nitrides are effective H trapping sites and have been proven to be beneficial to mitigating HE susceptibility,
e.g. in martensitic steels [120–122]. However, there are few studies on the effect of micro-alloy precipitates on the HE behavior of MMS. Park et al [123] evaluated the possibility of micro-alloying additions of Nb, Ti and V on enhancing HE resistance in a 0.08C–6Mn (wt.%) base MMS, and concluded that the influence is negligible. As shown in figure 18(a), H-charged micro-alloyed MMS show either limited improvement or even deterioration in terms of elongation loss, compared to the base MMS. Meanwhile, the H desorption curves are almost identical, as shown in figure 18(b). They attributed the trivial influence of precipitates on the HE behavior to the nature of two-phase microstructure of MMS, i.e. precipitates forming mainly in ferrite while H preferentially trapped in austenite, which is verified by the H distribution via Ag decoration technique, as shown in figures 18(c) and (d).

In addition, dislocations also affects the H distribution and thus the HE characteristics of MMS. Although dislocations essentially show a relatively lower H desorption energy (∼20 kJ mol⁻¹ [117]) than other strong H-trapping sites, they cannot be ignored in terms of storing reversible Wang et al [124] performed cold rolling in a 0.25C–8.67Mn–2.69Al–0.54Si–0.004P–0.004 S (wt.%) MMS, and found that the reversible H content increases significantly with pre-strain, while the irreversible H remains relatively constant. The effect of pre-strain on the reversible H absorption is attributed to the increase of dislocation density and the H diffusion coefficient because of deformation-induced martensite.

5.2. Effects of mechanical behaviors on HE resistance

Deformation behavior often alters the microstructure and has a huge influence on the H distribution and resultant HE phenomena. In MMS, one remarkable microstructural change is the deformation-induced martensite (DIM) transformed from retained austenite [102]. Owing to the drop in H solubility and rise in H diffusion rate, the H inherited from retained austenite will redistribute and thus deteriorate HE resistance after the formation of DIM [117]. Zhang et al [125] compared the HE susceptibility of the 0.2C–4.88Mn–3.11Al–0.62Si (wt.%) MMS processed with different IA temperatures, and proposed that the stability of austenite plays a critical role in governing the HE behavior. With increasing IA temperature, the austenite increases in both amount and grain size. However, it becomes more unstable with concomitant reduction in Mn and C content. Further to an IA temperature of 850 °C, the austenite cannot even sustain at ambient temperature and transform to martensite. They reported that brittle fracture fraction increases obviously due to the destabilization of austenite.
Figure 19. The microstructures of (a) 770 °C and (b), (c) 850 °C intercritically annealed MMS after 5% tensile strain. Note that the MMS has a composition of 0.2C–4.88Mn–3.11Al–0.62Si (wt.%). White and yellow arrows indicate micro-cracks at $\alpha/\alpha'$ and $\alpha'/\alpha'$ boundaries, respectively. Reprinted from [125]. Copyright (2019), with permission from Elsevier.

Figure 20. Slow strain rate tensile (SSRT) curves of the MMS intercritically annealed at 750 °C for (a) 10 min (b) 60 min and (c) 360 min. (d) Variations of index of relative susceptibility to HE (HEI) with intercritical annealing time. Note that the MMS has a composition of the 0.2C–5Mn–3Al–0.6Si (wt.%). Reprinted from [126]. Copyright (2018), with permission from Elsevier.

an extreme case, the 850 °C annealed MMS presents prevailing brittle intergranular fracture at the crack initial region due to HEDE. As demonstrated in figure 19, crack nucleates at $\alpha/\alpha'$ and $\alpha'/\alpha'$ boundaries in the 850 °C annealed MMS at 5% tensile strain, while no crack is observed in the 770 °C annealed MMS. In another work by Shao et al [126], the austenite grains grow with increasing IA time and consequently become unstable. Therefore, the elongation loss shows an increasing trend as seen in figure 20.

The Lüders banding is a commonly observed plastic instability phenomenon in MMS [55]. Zhang et al [127] revealed the adverse effect of the Lüders band on resisting HE in a 0.14C–7.65Mn–1.49Al (wt.%) MMS, where the higher fraction of martensite formation induced by the larger Lüders strain and the associated higher local flow stress both contribute to the deterioration of HE resistance, as shown in figure 21.

Naturally, it is inspired to enhance HE resistance by enhancing the austenite stability. In a recent work by Sun et al [128],
a novel strategy, chemical heterogeneity, was proposed to enhance the crack resistance and local hydrogen trapping. In this work, the chemical heterogeneity was achieved by applying a short time IA on cold rolled 0.2C–10.2Mn–2.8Al–1Si (wt.%) MMS, which causes incomplete partitioning/homogenization of Mn. As illustrated in figure 22(a), the well-tuned Mn distribution constructs rich and lean regions in the MMS. During H-induced crack propagation, the Mn-lean austenite at crack tip will transform to DIM and then fracture. Once encountering with Mn-rich austenite, the crack can be arrested through two effects: (a) H trapping by the stable austenite; and (b) the enhanced local plasticity of stable austenite. An example of the crack blunting scene is captured in figure 22(b), and the chemical heterogeneity is confirmed by APT in figure 22(c). This strategy can generally improve the HE resistance up to 122%, compared with other strategies, such as carbide precipitation and grain refinement.

Given the significant contrast of mechanical properties among austenite, ferrite and martensite in MMS, these phases react differently to applied load, generally leading to strain partitioning between phases and stress concentration at phase interfaces, both of which promote crack initiation [129]. To minimize the negative effects of deformation incompatibility, Li et al [130] designed a co-precipitation hardened MMS with NiAl in ferrite and Cu-rich particles in austenite. Two MMS, 0.01C–7Mn–1.5Al–2.5Ni (wt.%) and 0.01C–7Mn–1.5Al–2.5Ni–1.5Cu (wt.%), have been fabricated and noted as Cu-Free and Cu-Added, respectively. They exhibit similar mechanical properties but different HE resistance, as shown in figure 23(a). Through investigation of lattice strain of austenite and ferrite, it is proven that Cu-added MMS has a better strain compatibility, which is demonstrated by the analogous nanoindentation curve of each phase in figure 23(b) and by the slight difference in nano-hardness in figure 23(c). This improvement in deformation compatibility intrinsically decreases the strain partitioning among phases and thus inhibits interface crack initiation, thence improving HE resistance.

5.3. Perspectives of HE resistant design

From the overview of recent works on HE of MMS, it is found that the retained austenite plays a critical role. Despite a pursuit of TRIP effect, DIM transformed from austenite is usually the dominant cause of HE [125–127]. Nevertheless, improved HE resistance can be obtained by either utilizing austenite as the H trapping site [109, 128], or breaking its percolation [117]. Single precipitation provides minor help for improving HE resistance [123], while co-precipitation...
can enhance HE resistance through better deformation compatibility [130]. Moreover, pre-straining or alternative ways to increase the dislocation density in ferrite can enhance the reversible H absorption [116, 117, 124]. Therefore, two pathways for HE resistant MMS design are suggested: (a) to reduce the TRIP effect and obtain a relatively small fraction of stable austenite as strong H trapping site and crack blunting site; (b) to obtain a better deformation compatibility either by precipitation or increasing dislocation density, and to increase H solubility or decrease H diffusion rate by increasing defects density, thus to ultimately prevent grain boundary or phase boundary cracking.

6. Summary

Here, we systematically reviewed the state-of-the-art developments of MMS. For the alloy design, the influence and selection strategy of different alloying elements, such as C, Mn, Al, Si, etc., have been discussed. Several typical thermomechanical processing routes, including IA, Q&P, chemical patterning, WR, WS, D&P, are summarized. The microstructures and mechanical properties of the MMS produced by different processing routes are compared and discussed, which provides guidance for the development of new high-performance MMS in the future. The fracture and toughening mechanisms of MMS are reviewed, including the methodology for the quantitative measurement of fracture toughness, the fracture modes, and the potential toughness enhancing strategies. It is noted that the research on the damage, fracture and toughening mechanisms of MMS is still inadequate compared to other steel grades such as the HSLA and TWIP steels. Further in-depth investigations, especially in-situ characterization, are needed in the future. Finally, the HE mechanisms of MMS are reviewed, based on which strategies for development of novel MMS with good HE resistance are suggested.
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Conflict of interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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