

Review

# Austenite-Based Fe-Mn-Al-C Lightweight Steels: Research and Prospective

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**Abstract:** Fe-Mn-Al-C lightweight steels have been investigated intensely in the last a few years. There are basically four types of Fe-Mn-Al-C steels, ferritic, ferrite-based duplex/triplex (ferrite + austenite, ferrite + austenite + martensite), austenite-based duplex (ferrite + austenite), and single-austenitic. Among these steels, austenite-based lightweight steels generally exhibit high strength, good ductility, and outstanding weight reduction effects. Due to the addition of Al and high C content,  $\kappa'$ -carbide and  $\kappa$ -carbide are prone to form in the austenite grain interior and at grain boundaries of lightweight steels, respectively, and play critical roles in controlling the microstructures and mechanical properties of the steels. The microstructural evolution, strengthening mechanisms, and deformation behaviors of these lightweight steels are quite different from those of the mild conventional steels and TRIP/TWIP steels due to their high stacking fault energies. The relationship between the microstructures and mechanical properties has been widely investigated, and several deformation mechanisms have also been proposed for austenite-based lightweight steels. In this paper, the current research works are reviewed and the prospectives of the austenite-based Fe-Mn-Al-C lightweight steels are discussed.

**Keywords:** Fe-Mn-Al-C lightweight steel;  $\kappa'$ / $\kappa$ -carbide; microstructure; mechanical property; deformation mechanism



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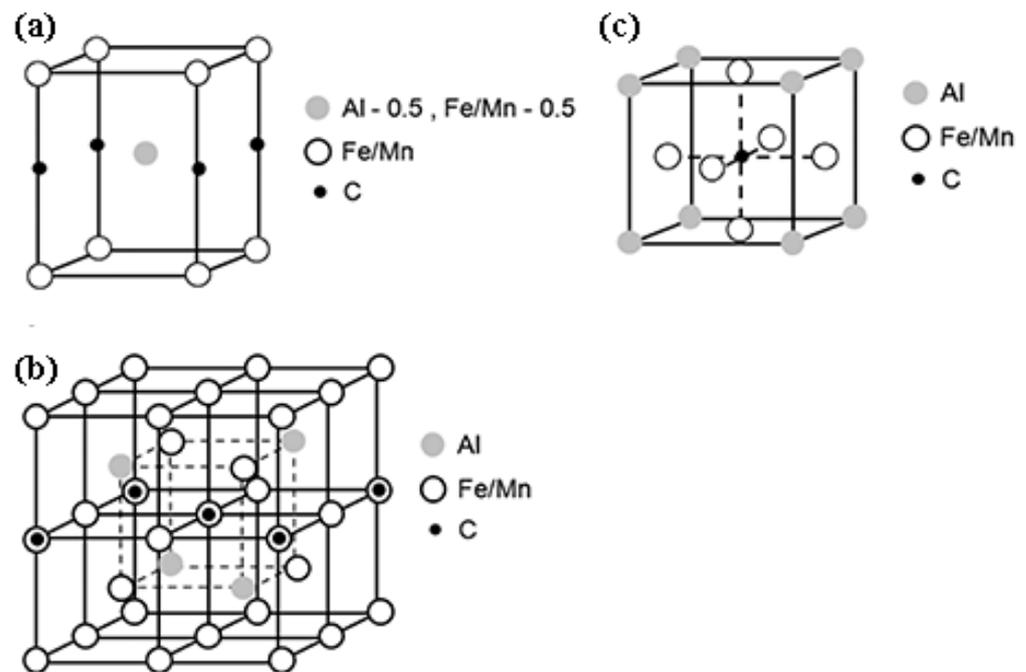
## 1. Introduction

Fe-Mn-Al-C lightweight steels, also known as low-density steels, first developed in the 1950s as substitutes of Fe-Cr-Ni stainless steels, have drawn research interests for their good comprehensive mechanical properties and lightweight effect as structural materials in the last a few years [1]. Fe-Mn-Al-C lightweight steels possess a variety of mechanical properties ranging by tailoring their microstructures, e.g., yield strength of 300–1200 MPa, ultimate tensile strength of 600–1500 MPa and total elongation of 30–100% [1]. In addition, these alloys have been reported to possess good service properties such as fatigue properties [2–7] and oxidation resistance at elevated temperatures [8–12]. These promising properties of Fe-Mn-Al-C steels have attracted considerable interests in several fields, such as transportation, especially in automobile vehicles and power trains as well as military use [1].

Based on the phase constituents of the materials, several types of Fe-Mn-Al-C steels have been investigated, such as ferritic, ferrite-based duplex/triplex (ferrite + austenite, ferrite + austenite + martensite), austenite-based duplex (ferrite + austenite), and austenitic ones.

Fe-Al lightweight steels alloyed with Mn lower than 5% and a very low C content possess a fully ferritic microstructure, which may contain A2-disordered FeAl, B2-ordered FeAl (Figure 1a) and DO<sub>3</sub>-ordered Fe<sub>3</sub>Al (Figure 1b) at room temperature depending on Al content [1,13–20]. The Fe-Al alloys based on FeAl or Fe<sub>3</sub>Al intermetallic compounds always show promising properties for high temperature structural applications due to their resistance to oxidation, sulfidation and carburizing, good resistance to corrosion

in sea water, high resistance to wear, erosion, or cavitation, and high strength-to-weight ratios [19,20].



**Figure 1.** Schematic visualization [18] of the supercell of (a) B2, (b) D0<sub>3</sub>, (c)  $\kappa'$ -carbide. (Reproduced with permission from [18]. Copyright 2013 Elsevier).

Ferrite-based duplex and triplex Fe-Mn-Al-C steels with moderate Mn (Mn: 2–12%) and C contents (C: 0.05–0.5%) possess the microstructures consisting of austenite +  $\delta/\alpha$ -ferrite and austenite +  $\delta/\alpha$ -ferrite + martensite, respectively, in which the fraction of ferrite is higher than 50% [21–37]. In this kind of steels, the transformation-induced plasticity (TRIP) effect is a very important mechanism in enhancing the strength and ductility of the materials.

Austenite-based duplex Fe-Mn-Al-C steels containing a higher Mn content, typically between 8 and 32%, Al up to 12%, and C between 0.3 and 1.2% are characterized by austenite +  $\delta/\alpha$ -ferrite or austenite +  $\alpha$ -ferrite [18,38–84]. Different from the ferrite-duplex one, the fraction of austenite is more than half in the austenite-based duplex steels, and the stability of austenite is quite high due to the high alloying elements.

Full austenite structure at room temperature can be obtained in Fe-Mn-Al-C steels with high Mn and high C contents, which are in the range of 13–40% and 0.6–2.0%, respectively, in spite of the high-Al content [63,64,85–130]. Meanwhile, the fully austenitic microstructure has been also obtained in the medium-Mn Fe-Mn-Al-C steels [46].

The increase in Al content in steels exhibits good lightweight effects but also easily gives a rise to cause the formation of brittle intermetallic compounds, eventually leading to poor ductility [131]. Among these four types of lightweight steels mentioned above, the austenite-based Fe-Mn-Al-C steels show superior weight reduction effect via alloying more Al and possess both high strength and ductility, which are closely associated with their unique microstructure features and deformation mechanisms. The microstructural evolution of austenite-based Fe-Mn-Al-C steels is different from the conventional steels, and they also show distinguished features from TRIP and TWIP steels due to the additions of relatively high alloying elements. The physical metallurgy is quite complex and there are still some theoretical aspects which needs to be clarified in the austenite-based Fe-Mn-Al-C lightweight steels. Furthermore, the detailed research work needs to be carried out for practical applications. In this paper, the recent developments of austenite-based single and

duplex Fe-Mn-Al-C lightweight steels are reviewed and future directions for the research in Fe-Mn-Al-C steels are proposed.

## 2. Phase Constituents in the Austenite-Based Fe-Mn-Al-C Lightweight Steels

The medium-Mn ( $5\% \leq \text{Mn} \leq 12\%$ ) and high-Mn ( $\text{Mn} > 12\%$ ) lightweight steels are characterized by single austenite and austenite-based duplex (austenite + ferrite) matrix due to their high content of austenite stabilizer elements of Mn and C [129]. When the content of ferrite stabilizer element, i.e., Al, is high and the associated  $\text{Ni}_{\text{eq}}/\text{Cr}_{\text{eq}}$  is relatively low, a considerable amount of banded coarse  $\delta$ -ferrite forms during solidification and became inherited during subsequent hot rolling, cold rolling and annealing [44,80,81,129,132–134]. Meanwhile, some fine  $\alpha$ -ferrite grains could form during hot deformation or annealing in the intercritical temperature ( $\gamma \rightarrow \alpha$ ), which is propitious to the microstructure refinement [46,81,111].

Regarding precipitation, the formation of  $\kappa'$ -carbide precipitates depends on heat treatment conditions. The phase has a perovskite crystal structure designated as  $L'1_2$ , and its ideal stoichiometry is  $(\text{Fe}, \text{Mn})_3\text{AlC}$  [1,135]. The crystal structure of the phase is illustrated in Figure 1c. A metastable  $(\text{Fe}, \text{Mn})_3\text{AlC}_x$  ( $x < 1$ ) phase has the same crystal structure as the phase but with an uncompleted occupation of the C atoms [136]. The off-stoichiometric concentration of Al was explained by mismatch-induced strain, which facilitates the occupation of Al sites in the  $\kappa'$ -carbide by Mn atoms [123,125,136].

When the high-Mn austenitic lightweight steels were quenched from high temperature or aged at 450–650 °C, nano-sized  $\kappa'$ -carbide particles formed within austenite grains [90,91,116,125]. This intra-granular  $\kappa'$ -carbide is a metastable  $(\text{Fe}, \text{Mn})_3\text{AlC}_x$  phase which is coherent to the matrix. The steel matrix (austenite) and  $\kappa'$  phase have the cube-on-cube crystallographic orientation relationship from these selected area diffraction patterns (SADPs), i.e.,  $[100]_{\kappa} // [100]_{\gamma}$ ,  $(100)_{\kappa} // (100)_{\gamma}$  [114]. It has been long believed that the formation of intra-granular  $\kappa'$ -carbide is through spinodal decomposition and following ordering reaction [137,138]. Transmission electron microscopy and X-ray diffraction were generally used to provide experimental evidence supporting the spinodal decomposition-ordering mechanism by observing the modulated structure [139], diffuse satellites around the (200) diffraction spots in electron diffraction patterns, and XRD side band peaks around the (200) reflections [138,140]. However, some recent transmission electron microscopy (TEM, FEI Titan Themis, Hillsboro, OR, USA) and atomic probe tomography (APT, FEI Helios Nano-Lab 600i, Hillsboro, OR, USA) results obtained in an Fe-30Mn-9Al-1.2C lightweight steel indicated that the formation of an ordered structure was earlier than chemical partitioning of any solute elements during the early stage of  $\kappa'$ -carbide precipitation [141]. Near-atomic scale characterization of an austenite-based Fe-20Mn-9Al-3Cr-1.2C steel, using high-resolution scanning TEM (HRSTEM, FEI Tecnai G2-20, Hillsboro, OR, USA) and APT also revealed that the initially-formed  $\kappa'$ -carbide (2–3 nm in particle size) are characterized by an ordered  $L'1_2$  structure but without detectable chemical partitioning [114]. However, the increasing Mn content could delay the formation of intra-granular  $\kappa'$ -carbide via suppressing the C occupation of the vacancy at the body-centered site of  $L'1_2$ , which is related to the C ordering process [123,125,136]. Thus, the intra-granular  $\kappa'$ -carbide is more prone to precipitate in medium-Mn lightweight steels.

Meanwhile, the extended aging and relative-low-temperature annealing caused the precipitation of perovskite-structured  $(\text{Fe}, \text{Mn})_3\text{AlC}_x$  carbide at the grain boundaries of austenite [40–42,45,57,58,62,90,91,99,119]. Hereafter, we distinguish the ordered grain boundary  $L'1_2$  phase from the intra-granular  $\kappa'$  phase of the same structure by naming the former as  $\kappa$ -carbide. Such inter-granular  $\kappa$ -carbide grew into the austenite grains in the form of a lamellar structure together with  $\alpha$ -ferrite through the following dominant route: the eutectoid reaction  $\gamma \rightarrow \kappa + \alpha$  [40]. The cellular transformation is a form of continuous reaction which occurs during the transformation of high-temperature austenite into lamellae of austenite,  $\alpha$ -ferrite, and  $\kappa$ -carbide [62]. The formation of coarse second-phase particles

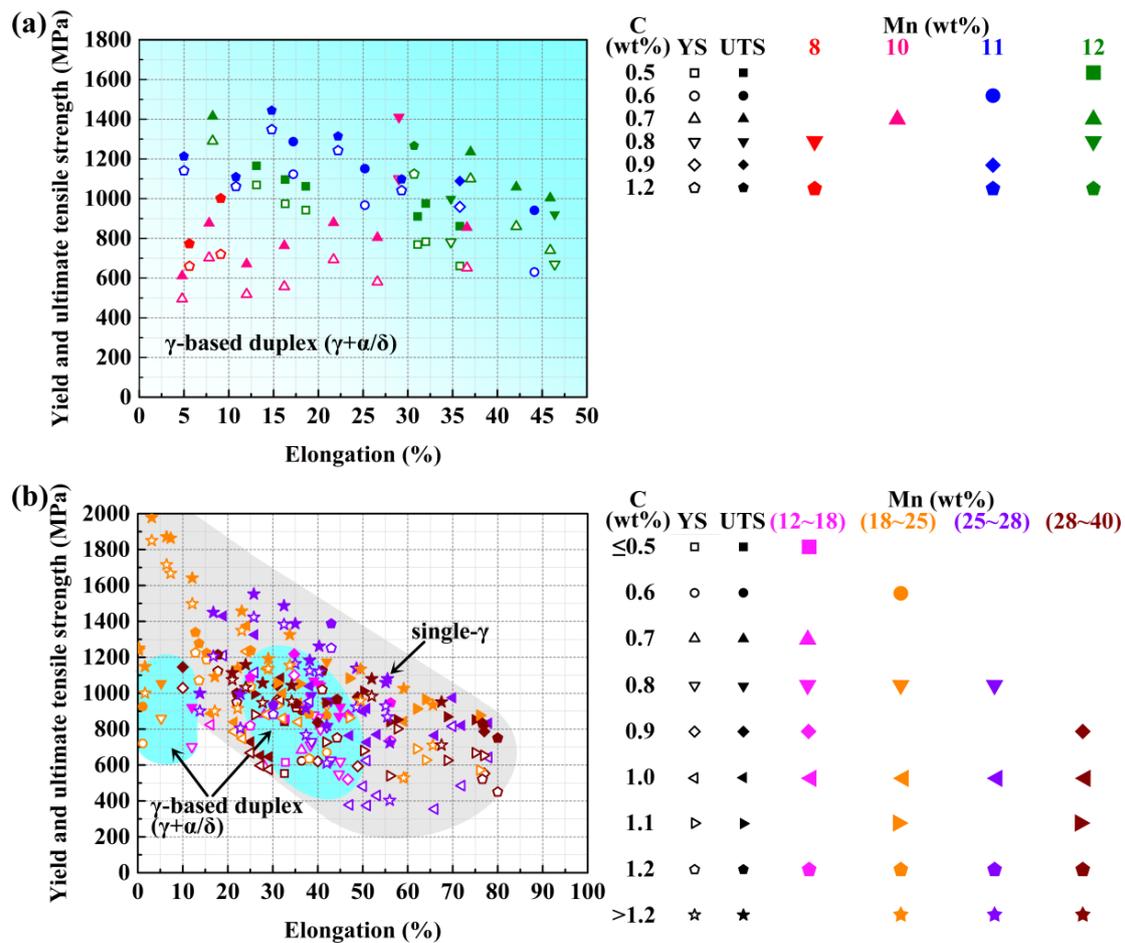
which have a lamellar morphology was also observed after aging for a longer period of time in a solution treated Fe-(11–30)Mn-(7.8–10)Al-(0.8–2.0)C alloys [45,58,90,91,99,119].

In the Fe-27Mn-12Al-0.8C duplex lightweight steel [18], various ordered phases such as DO<sub>3</sub>, B2 and  $\kappa'$ -carbide were formed in the duplex microstructure upon quenching in water after intercritical annealing. Fine DO<sub>3</sub> were evenly distributed through both B2 domains and disordered ferrite matrix. Meanwhile, nano-sized  $\kappa'$ -carbides precipitated in austenite. Similar results were also attained in the Fe-11Mn-10Al-0.9/1.2C, Fe-15Mn-10Al-1.0C and Fe-18Mn-10Al-1.2C steels alloyed with lower Mn contents [46,59,63].

Since there are variations of phase constituents in Fe-Mn-Al-C steels which are controlled by compositions and processing schedules, the mechanical behavior could be tailored in a wide range for these kinds of steels.

### 3. Mechanical Properties

The mechanical properties of the representative medium-Mn [38,40–42,45,46,50,51,53–55,84] and high-Mn [3,18,57,58,63,67,69,78,79,83,85,91,100,103,107,109–111,113,116,122,124,127,129,141–159] lightweight steels are shown in Figure 2a,b. It is clearly indicated that both the medium-Mn and high-Mn steels possess good mechanical properties and show large space to be regulated, yield strength: 375–1850 MPa, ultimate tensile strength: 765–1978 MPa, and total elongation: 1–80%. Generally, both intra-granular  $\kappa'$ -carbide and inter-granular one can effectively improve the yield strength of lightweight steels, regardless of Mn content [40–42,45,46,57,58,84,85,91,122]. However, the coarse inter-granular  $\kappa$ -carbide results in an abrupt loss of elongation, while the fine intra-granular  $\kappa'$ -carbide enhance the strengths of lightweight steels without significantly sacrificing ductility but brings out a relatively high yield ratio (>0.9) [40–42,45,46,91]. Moreover, although the relationship between strength and ductility of duplex and single-austenitic lightweight steels follow the “banana” curve, the mechanical properties of single-austenitic steels with high-Mn content seems to be superior to those of duplex ones. For instance, the austenitic steel (Fe-28Mn-9Al-1.8C) demonstrates ultrahigh strength (yield strength of 1383 MPa and ultimate tensile strength of 1487 MPa) with good elongation of ~32.5% [91]. The increase in Mn content seems to bring more room for improving the strength and ductility of steels, but it also increases the difficulty in fabrication, not to mention the sharp increment in material cost [160]. Achieving high performance lightweight steel and maintaining its economy is also an important subject during its development process.



**Figure 2.** Comparison of room-temperature tensile properties of the present developed (a) medium-Mn [10,38,40–42,45,46,50,51,53–55,84] and (b) high-Mn [3,18,57,58,63,67,69,78,79,83,85,91,100,103,107,109–111,113,116,122,124,127,129,141,143–159] lightweight steels.

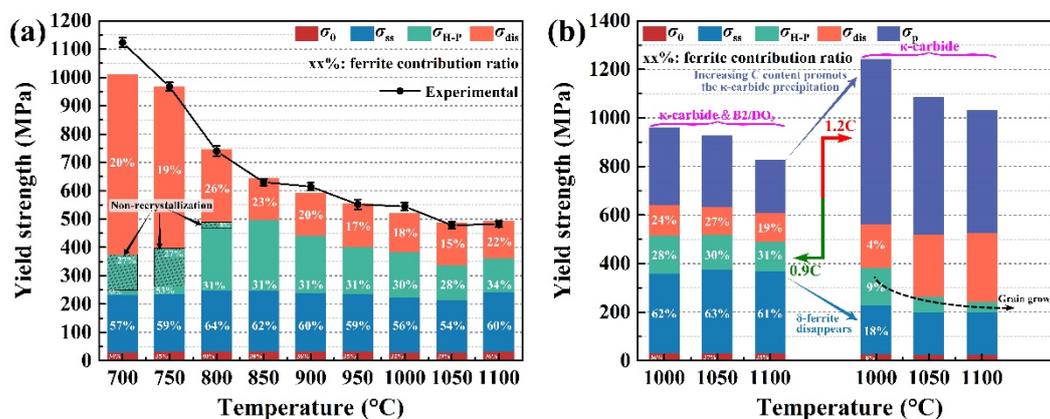
#### 4. Strengthening Mechanisms

Solid solution hardening plays a role in the strengthening of Fe-Mn-Al-C steels due to the high amount of alloying elements C, Al, and Mn in these steels and grain refinement is another strengthening mechanism [46,74,125,141]. The austenite grain size can be refined by thermomechanical processing (TMP) combined with cold working and annealing [42,111,155]. The existence of ferrite can also make the austenite size decrease owing to the prohibition of growth of austenite in both medium-Mn and high-Mn lightweight steels [46,111].

It was reported that the yield strength of these steels increases as the Al concentration increases [74]. The high yield strength of 12Al steel, 952 MPa, is due to fine grain strengthening, precipitation strengthening, the existence of ferrite, and Al solution strengthening. Quantitative investigations in Fe-26Mn-Al-1C indicated that the effect of Al on yield strength of the alloys is not quite significant in the Al range of 3 to 10%. Precipitation hardening is the most significant strengthening mechanism in the alloys containing homogeneously distributed nano-sized  $\kappa'$ -carbides.

Dislocations moving through an austenitic matrix containing intra-granular  $\kappa'$ -carbide can either shear the precipitates or bypass them and consequently result in alloy strengthening [116]. In Fe-Mn-Al-C lightweight steels, it is believed that the operative mechanism (shearing mechanism or Orowan bypassing mechanism) and the corresponding strengthening effect are closely associated with the size of  $\kappa'$ -carbide [116,125]. For a given volume fraction (~20%), Yao et al. calculated the shearing strengthening effect of the Fe-30.4Mn-

8Al-1.2C steel aged at 600 °C for 24 h based on the antiphase boundary (APB) energy, which is about 500 MPa [116]. As the size of  $\kappa'$ -carbide is beyond the critical radius (~6.8 to 13.5 nm), the Orowan looping can in principle be activated [116]. Since the lower Mn content facilitates the formation of intra-granular  $\kappa'$ -carbide [123,125,136], the precipitation hardening is expected to reach the higher value in the medium-Mn steels. Liu et al. studied the strengthening mechanisms of the cold-rolled Fe-11Mn-xAl-yC ( $x = 7/11$ ,  $y = 0.6/0.9/1.2$ ) medium-Mn lightweight steels annealed at 700–1100 °C; see Figure 3 [46]. It is clearly observed that the maximum of the  $\kappa'$ -carbide precipitation strengthening effect of the 1000 °C-annealed Fe-11Mn-10Al-1.2C steel is estimated as 679 MPa. Meanwhile, the existence of high density dislocations in the partially-recrystallized hetero-structured Fe-11Mn-7Al-0.6C steels also results in a considerable increment in yield strength of the materials.

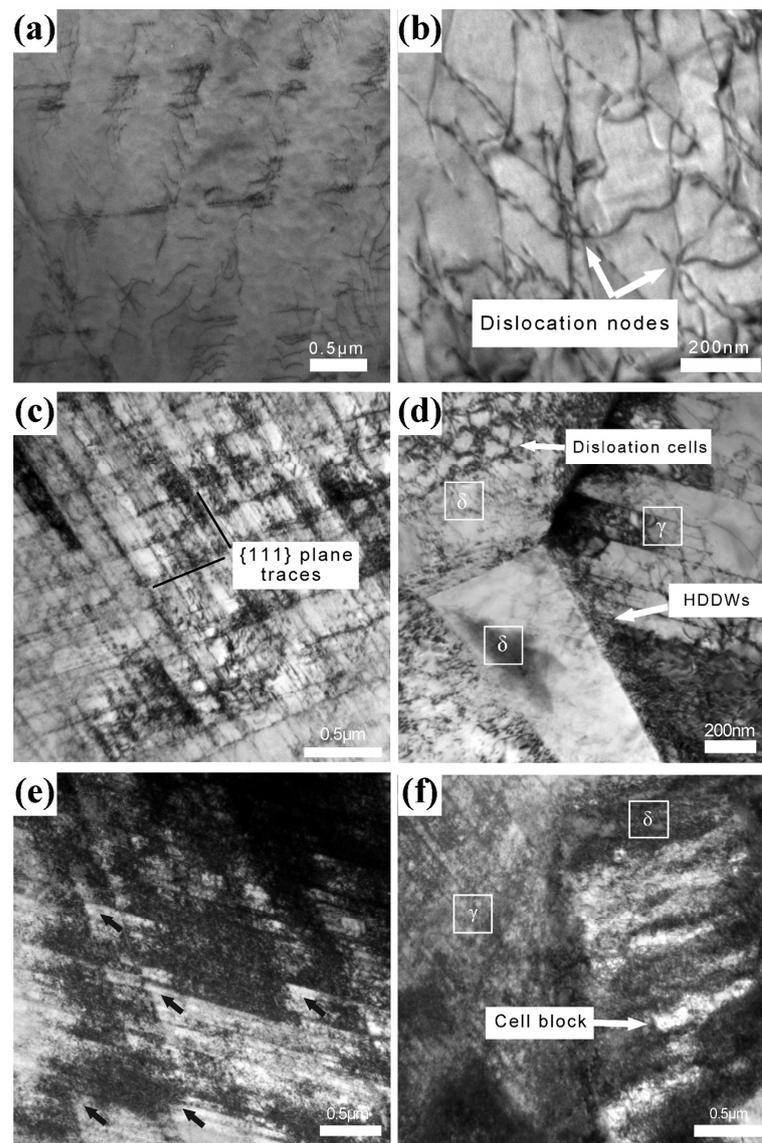


**Figure 3.** The calculated yield strength values for assessing the strengthening mechanisms of the (a) Fe-11Mn-7Al-0.6C steel and (b) Fe-11Mn-10Al-0.9/1.2C steels under various annealing temperatures [46]. (Reprinted with permission from [46]. Copyright 2022 Elsevier).

### 5. Strain Hardening Behaviors and Deformation Mechanisms

Several researchers investigated the strain hardening behaviors of Fe-Mn-Al-C steels. The results in the investigation of tensile deformation of a duplex Fe-20Mn-9Al-0.6C steel [67] revealed that strain hardening in both austenite and ferrite was monotonic during tensile deformation, but the strain hardening exponent of austenite was higher than that of ferrite, indicating the better strain hardenability of austenite. Three low-density Fe-18Mn-10Al-xC steels containing 0.5, 0.8 and 1.2 C (wt%) were utilized to investigate effects of C contents on the microstructural evolution and the corresponding mechanical behaviors during plastic deformation [63]. The differential Crussard–Jaoul (C–J) analysis demonstrated a two-stage strain hardening behavior in both 0.5C and 0.8C steels and a three-stage one in the 1.2C steel. This difference in strain hardening behavior was further understood in terms of microstructural analysis at the different stages of plastic deformation.

The strain hardening behavior in relation with the evolving dislocation substructures during uniaxial tensile deformation for an austenite-ferrite Fe-18.1Mn-9.6Al-0.65C steel was investigated [77]. The steel consisted of austenite and ferrite and possessed a good combination in mechanical properties. The deformation mode of austenite is dominated by planar glide and Taylor lattices and microbands formed as the deformation proceeded, whereas dislocation nodes, dislocation cells, and cell blocks formed due to the occurrence of wavy glide in ferrite, as shown in Figure 4 [77]. Three-stage strain hardening behavior was revealed in this steel, which is similar to the aforementioned steel. Since the increase of Al content in the steels increases the volume fraction of shearable  $\kappa'$ -carbide precipitates, an increased strain softening is activated in glide plane, which results in a decreased density of slip bands and significant decrease of the strain hardening rate [107]. Generally, the Fe-Mn-Al-C steels containing intra-granular  $\kappa'$ -carbides show low strain hardening rate because the nanocrystalline coherent precipitates are easily sheared by gliding dislocations [131].

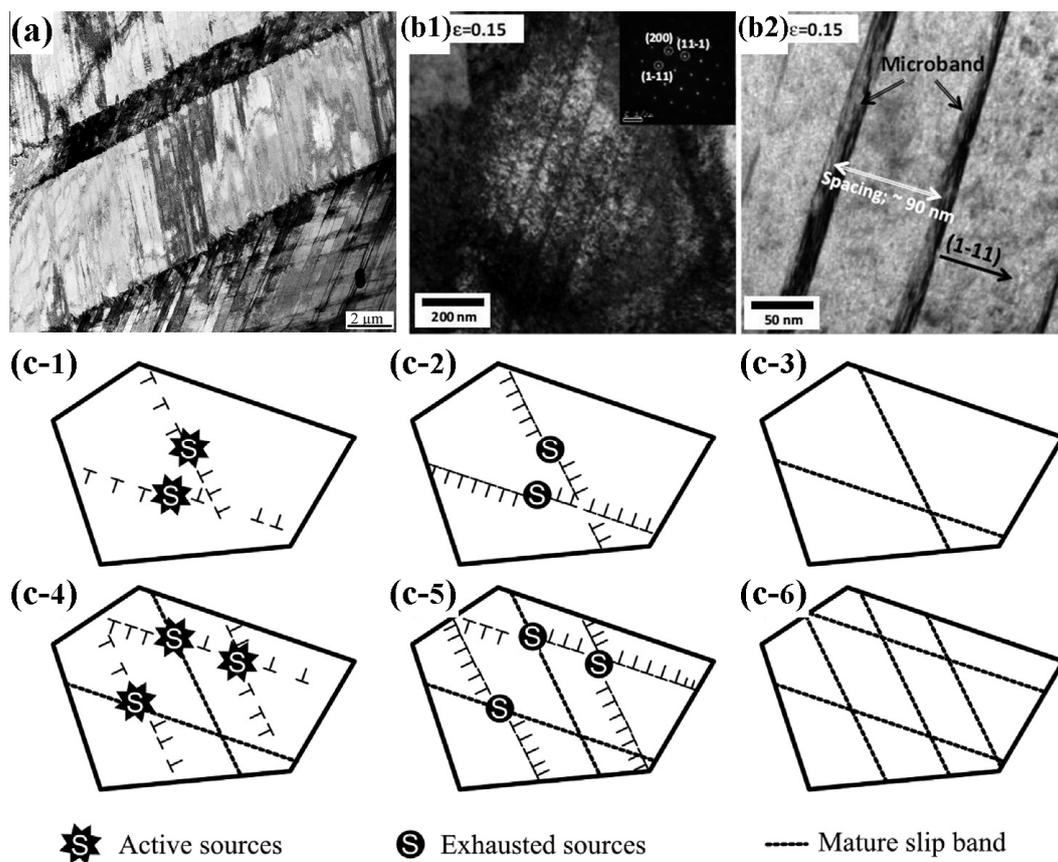


**Figure 4.** Dislocation substructure evolution within (a,c,e) austenite ( $\gamma$ ) and (b,d,f) ferrite ( $\delta$ ) in the Fe-18.1Mn-9.6Al-0.65C steel at different strains. (a,b)  $\epsilon = 0.02$ : (a) planar dislocation arrays, (b) dislocation nodes, (c) Taylor lattices ( $\epsilon = 0.18$ ), (d) dislocation cells ( $\epsilon = 0.10$ ), (e,f)  $\epsilon = 0.36$ : (e) microbands intersections, (f) cell block [77]. (Reproduced with permission from [77]. Copyright 2015 Elsevier).

The deformation mechanisms of austenite can be predicted basing on the thermodynamic calculation of stacking fault energy (SFE) [51,54,64,86,94]. It was reported that TRIP effect appears when the SFE is lower than  $\leq 18$  mJ/m<sup>2</sup>, and TWIP effect is dominant when the SFE is between 18–35 mJ/m<sup>2</sup>. When the SFE is higher than 60 mJ/m<sup>2</sup>, neither TRIP nor TWIP effect appears [161]. Mn, C and Al all increase the SFEs of Fe-Mn-Al-C steels. Generally, for FCC materials, it is difficult for extended partials to form when the material possesses high SFE. Therefore, cross slip of screw dislocations of the extended partials is easy and wavy slip would be dominant, forming cellular structure. However, planar slip features have been found in the FCC materials with high solute element concentrations. In this case, SFE is not the dominant factor influencing the deformation mode.

There are basically three deformation mechanisms reported in high SFE Fe-Mn-Al-C steels, shear band-induced plasticity (SIP) [75], microband-induced plasticity (MBIP) [87,89], and dynamic slip band refinement [97]; see Figure 5. For the SIP mechanism, it was suggested that the enhanced ductility is closely associated with the formation of the

homogeneous shear band accompanied by the dislocation glide sustained by the uniform arrangement of nano-size  $\kappa'$ -carbides coherent to the austenite matrix with defined inter-particle spacing. For the MBIP mechanism, planar slip occurs in austenite when the strain is low, Taylor lattice appear as the deformation proceeds and microbands form afterwards, increasing the strain hardening capacity of the steels. Strain hardening by dynamic slip band refinement in a Fe-30.4Mn-8Al-1.2C high-Mn lightweight steel was investigated, and it was characterized that material deforms mainly by planar dislocation slip causing the formation of slip bands [97]. The deformation mechanism was therefore regarded as dynamic slip band refinement. This slip band refinement-induced plasticity (SRIP) was also verified in Fe-29.8Mn-7.65Al-1.11C steel [122].



**Figure 5.** Three deformation mechanisms reported in high SFE Fe-Mn-Al-C steels. (a) shear band-induced plasticity: uniformly arranged shear bands on {111} planes within the austenitic matrix of a deformed high-Mn steel [75] (Reproduced with permission from [75]. Copyright 2016 Wiley.); (b) microband-induced plasticity: (b1) and (b2) well-developed microbands having distinct boundaries in austenite of the medium-Mn duplex lightweight steel at the true strain of 0.15 [51] (Reproduced with permission from [51]. Copyright 2015 Elsevier). (c) schematic illustration of dynamic slip band refinement: (c-1) activation of sources, (c-2) slip planes filled up with dislocations. (c-3) exhausted sources due to back stresses and fully developed slip bands, (c-4) activation of new sources, (c-5) and (c-6) newly activated sources undergone the same evolution as the previous sources leading to a refinement of the slip band substructure [97] (Reproduced with permission from [97]. Copyright 2016 Elsevier).

Tensile deformation of Fe-27Mn-12Al-0.8C duplex steel was studied in association with ordered phases [18]. In austenite, a single-planar dislocation glide is a dominant mechanism at low strains and multiple planar slip occurs at high strains, whereas short, straight segments of paired dislocations with narrow mechanical antiphase boundaries were formed in ferrite. It was reported that strain hardening of the duplex steel is associated

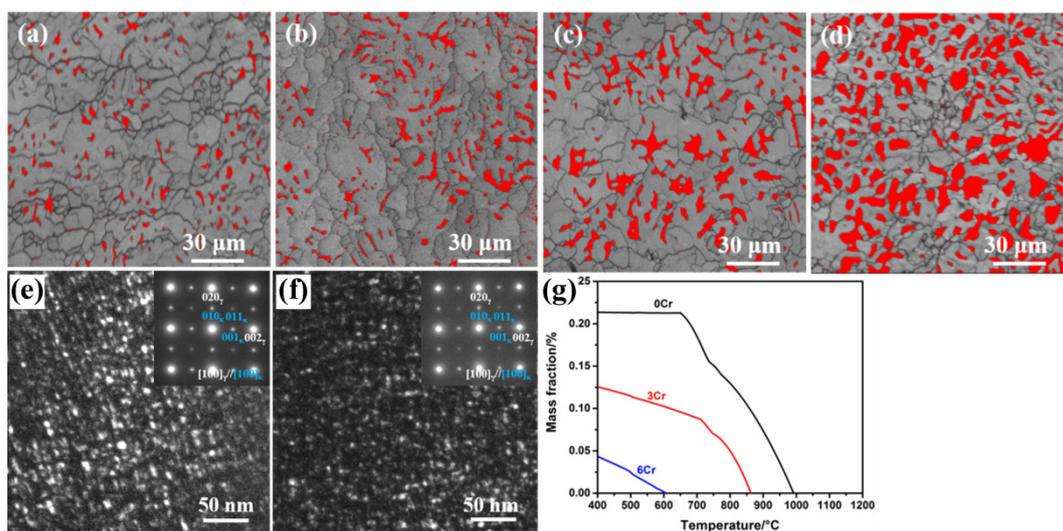
with the combined effect of the shearing of nano-sized ordered phase by superdislocations in ferrite and planar gliding dislocations in austenite.

## 6. Effect of Trace Alloying Elements

Kim et al. investigated the effect of another lightweight element, Si, on deformation mechanisms in light weight steels by atomic-scale analysis [120]. It was found that the addition of Si accelerated the formation kinetics of the  $\kappa'$  precipitates and increased the partitioning coefficient of carbon from 2.4 to 5.3. C-rich  $\kappa'$ -carbides are more resistant to shearing by dislocations due to a higher coherency strain and the formation of Al-C bonding which makes dislocation motion energetically more difficult. Therefore, the energy required for dislocation shearing  $\kappa'$ -carbides in the aged 1% Si steel was higher than the one in the aged Si-free steel.

The effect of Mo addition on the precipitation behavior the  $\kappa'$ -carbide in the austenitic Fe-Mn-Al-C lightweight steels was investigated [162]. First-principle calculations indicated that the substitution of Fe or Mn by Mo in  $\kappa'$ -carbide is energetically unfavorable with respect to the formation energy and it increases strain energy contribution to interfacial energy between austenite matrix and  $\kappa'$ -carbide. TEM observation and nano-indentation experiments showed that Mo delayed the kinetics of  $\kappa'$ -carbide formation and changed the age hardening behavior. This calculation was also verified by APT analysis, showing both are in a good agreement.

Sutou et al. reported the addition of Cr could improve the strength, hardness and cold-workability of Fe-20Mn-Al-C steels with higher C and Al contents [69]. The addition of Cr, which is a ferrite stabilizer, suppresses the formation of coarse inter-granular  $\kappa$ -carbides, and consequently, austenite retains more stability due to an increase in the amount of carbon inside austenite [163]. The increasing content of Cr in the Fe-20Mn-9Al-1.2C lightweight steel increased the volume fraction of ferrite but decreased the volume fraction of intra-granular  $\kappa'$ -carbide. Meanwhile, the increased Cr content also significantly slowed down the growth rate of  $\kappa'$ -carbides during isothermal aging treatment at 600 °C, as shown in Figure 6 [164].



**Figure 6.** Electron backscattered diffraction (EBSD) phase maps of the as-cast Fe-20Mn-9Al-1.2C steels with different Cr content [164]: (a) 0Cr; (b) 3Cr; (c) 6Cr; and (d) 9Cr (red and gray contrasts indicate  $\delta$ -ferrite and  $\gamma$ -austenite phases, respectively). TEM DF images of  $\kappa'$ -carbides of the aged steels with (e) 0Cr and (f) 3Cr contents and (g) the thermodynamic calculation showing the variation in the mass fraction of  $\kappa'$ -carbides different Cr content. (Reproduced with permission from [164]. Copyright 2022 Elsevier).

V has been added in the austenite-based lightweight steel so as to improve the strength and strain hardening rate through dual-precipitation of V-carbides and  $\kappa'$ -carbide and the alloying element V could exert an impact on the precipitation behavior of  $\kappa'$ -carbide [152,156]. First principle analysis showed that the addition of V would increase the nucleation barrier energy of  $\kappa'$ -carbide due to the segregation of V [156], and the precipitated V-carbides could induce the subsequent precipitation of  $\kappa'$ -carbides in the form of band distribution [156].

Adding individual Nb/Ti, and adding compound Nb and Ti in lightweight steels can both refine grains through the precipitation of carbides, and thus enhance the yield strength of materials [1–3]. Wang et al. studied the effect of Ti addition on the mechanical properties and microstructures of Fe-30Mn-10Al-1.57C-2.3Cr-0.3Si-xTi ( $x = 0, 0.3, 0.6,$  and  $0.9$  wt%), and it was revealed that grain refinement effect become extremely obvious with increasing Ti content [100]. However, there is continuing debate as to the effect of trace element addition on the precipitation of  $\kappa'$ -carbides [142]. Li et al. found  $\kappa'$ -carbides densely distributed at the (Ti,Mo,Nb)C/ $\gamma$  interface, which could act as the nucleation site of the  $\kappa'$ -carbide [143], whereas Park et al. reported Nb addition caused the consumption of C solute atoms to form the primary and secondary NbC carbides, thus lowering the precipitation rates of  $\kappa'$ -carbide [144].

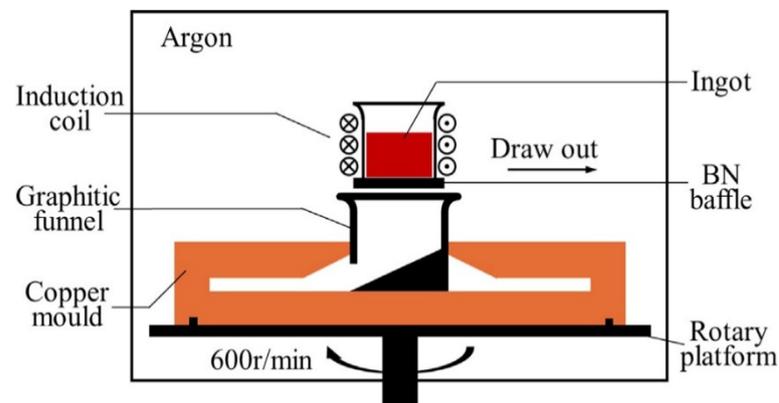
Recently, Cu addition was proposed as a promising method to achieve the high yield strengths of medium-Mn and high-Mn lightweight steels by the co-precipitation of nano-scale Cu-rich and  $\kappa'/\kappa$ -carbide particles [53,165]. Cu, as an austenite stabilizer, not only increases the volume fraction of austenite but also hinders the recrystallization due to the solute drag effect, and it promotes the formation of Cu-rich B2 particles and Cu-segregated interfacial layers [53]. Since the Cu-rich particles promoted the precipitation of nanosized  $\kappa'$ -carbide particles, the yield strength of particle-strengthened Fe-28Mn-9Al-0.8C-5Cu austenitic lightweight steel reaches 808 MPa with total elongation of more than 20% [165].

Sang-Heon Kim, Hansoo Kim, and Nack-J Kim reported a Ni-doped austenitic lightweight steel (Fe-16Mn-10Al-0.86C-5Ni) which possesses ultrahigh specific strength, good ductility and phenomenally high strain hardening owing to the unique duplex microstructure consisting of  $\gamma$ -matrix and evenly dispersed fine B2-intermetallic second phase [131]. The addition of Ni into the Fe-15Mn-10Al-0.8C lightweight steel led to the ordering of  $\alpha$ -ferrite and its transformation to stronger B2 compounds and prevented the formation of lamellar structure of  $\alpha + \kappa$  [166], and the interplay between B2 and  $\kappa$ -carbide precipitation was utilized to control the morphology and distribution of these precipitates. The initial formation of intra-/inter-granular  $\kappa'/\kappa$ -carbide particles within the hot-rolled Fe-21Mn-10Al-1C-5Ni steel is expected to increase the chemical driving force and correspondingly reduce the critical energy barrier for B2 nucleation, consequently facilitating the formation of a large fraction of B2 nanoparticles with size of 20–500 nm within austenite grains [149]. The investigation on the Fe-30Mn-10Al-0.9C-0.5Si-1.5Mo-1.5/3Ni steels demonstrates the reverse partitioning of Al from  $\kappa'$ -carbide to the  $\gamma$ -matrix through Ni addition, indicating that the affinity of Ni-Al is higher than that of C-Al [148].

## 7. Fabrication of Fe-Mn-Al-C Lightweight Steels

### 7.1. Fabrication Methods

Adding Al to steels could effectively reduce their mass density [75,167]. However, the excessive Al content can produce massive  $\text{Al}_2\text{O}_3$  and AlN inclusions which cause severe nozzle clogging and surface cracks in the continuous casting of slabs, which are great challenges for industrial production of lightweight steels [168–171]. Recently, a near-net shape approach to fabricate the lightweight steels by a near-rapid solidification process was proposed, which was conducted by the centrifugal casting (Figure 7) [50,70,84,143,165,172,173]. It was reported that such route could reduce the energy consumption during the rolling deformation and promote the near-rapidly solidified material possessed features of ultra-fine microstructure, low segregation, high solid solution, and possibly non-equilibrium or metastable phases [50,174]. It was revealed that near-rapidly solidified lightweight steels showed satisfactory mechanical properties.



**Figure 7.** Schematic of centrifugal casting used for near-rapid solidification [84]. (Reprinted with permission from [84]. Copyright 2019 Elsevier).

### 7.2. Hot Deformation Behavior

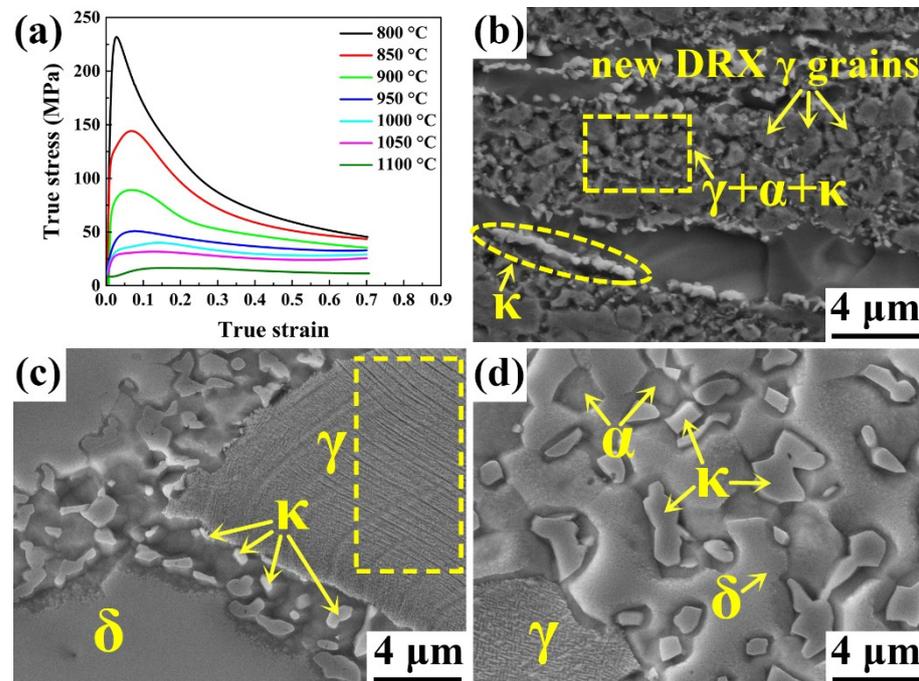
As the TMP is a devoted part of steel production, hot deformation behaviors of high-Mn and medium-Mn Fe-Mn-Al-C lightweight steels have been investigated by several researchers [39,175–183]. The hot deformation and dynamic recrystallization (DRX) behavior of Fe-27Mn-11.5Al-0.95C steel was investigated by compressive testing in the temperature range of 900–1150 °C and strain rate of 0.01–10 s<sup>-1</sup>. Typical DRX behavior was observed and a DRX kinetics model of the steel was established [175].

The high temperature behavior of the duplex low-density Fe-18Mn-8Al-0.8C steel was investigated in the temperature range of 600–1000 °C, and a 3D processing map was developed considering the effect of strain [176]. The dynamic transformation from austenite to ferrite was found to occur in the safe efficiency domain. Therefore, the microstructure factor must be considered in the high-Mn, high-Al alloys with relatively lower Mn concentrations. Continuous dynamic recrystallization of Fe-17.5Mn-8.3Al-0.74C-0.14Si steel with a duplex microstructure was investigated [177]. The formation of progressive sub-boundaries and its effect on the materials' ductility were explored.

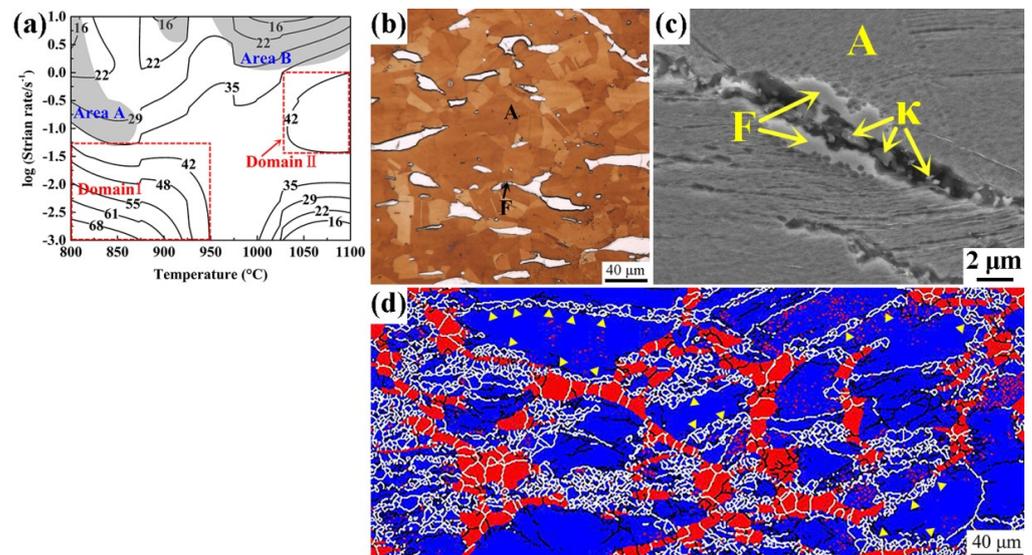
The hot deformation behaviors of the Fe-26Mn-8/10Al-1C steels were investigated by the 3D processing map at temperatures of 850–1050 °C and strain rates of 0.001–10 s<sup>-1</sup> and the effect of Al was considered [178]. The constitutive equations of the steels were established. The steel alloyed with more Al (i.e., 10Al steel) has a higher flow stress and a higher Z value (i.e., Zener–Hollomon parameter), indicating the increasing Al content suppressing the nucleation and growth of DRX. The number of unstable zone extends from one to two with increasing Al content and the instability region at each strain rate also increased.

As the increasing amount of C and Al together with the decreasing Mn content could facilitate the precipitation of both intra-/inter-granular  $\kappa'$ / $\kappa$ -carbides, the medium-Mn lightweight steels usually undergo hot working in the ferrite + austenite +  $\kappa$  phase region, thus resulting in complicated flow behaviors [39,43,44,180–183]. The study on the hot deformation behavior of Fe-11Mn-10Al-0.9C steel reveals the occurrence of dynamic precipitation of intra-/inter-granular  $\kappa'$ / $\kappa$ -carbides as well as discontinuous/continuous dynamic recrystallization (DDRX/CDRX) of austenite and ferrite, and a significant softening is observed, as shown in Figure 8 [181]. As the formation of inter-granular  $\kappa$ -carbide particles is detrimental to the hot workability of the medium-Mn lightweight steels [39,182], processing maps were developed by employing dynamic materials model (DMM) to determine the optimal hot deformation condition of the Fe-11Mn-10Al-0.9C steel [182]. According to the processing map, the best process window of the Fe-11Mn-10Al-0.9C steel at large strains (0.7) was identified as deformation temperature of 950–1100 °C and strain rate of 0.01–1.0 s<sup>-1</sup>; see Figure 9a. In this domain, the original coarse grains were refined, indicating that the high efficiency was dissipated by DRX (Figure 9b). Meanwhile, two unstable regions re-

sulted from inter-granular  $\kappa$ -carbides (Figure 9c) and necklace structure (Figure 9d) should be avoided during hot working.



**Figure 8.** (a) Flow curves of the Fe-11Mn-10Al-0.9C medium-Mn duplex lightweight steel deformed at the strain rate of  $0.001 \text{ s}^{-1}$  and (b–d) microstructural evolution showing the dynamic precipitation of GB  $\kappa$ -carbides at the deformation temperature of  $800 \text{ }^\circ\text{C}$  under various strain rates of (b)  $10 \text{ s}^{-1}$ , (c)  $0.01 \text{ s}^{-1}$ , (d)  $0.001 \text{ s}^{-1}$  [181]. (Reproduced with permission from [181]. Copyright 2020 Elsevier).



**Figure 9.** (a) The processing map the Fe-11Mn-10Al-0.9C medium-Mn duplex lightweight steel at true strain of 0.7, (b) optical image of the specimen deformed at  $1100 \text{ }^\circ\text{C}$  and  $0.1 \text{ s}^{-1}$  corresponding to Domain II showing a fine and uniform DRX microstructure, (c) scanning electron micrograph of the specimen deformed at  $800 \text{ }^\circ\text{C}$  and  $0.001 \text{ s}^{-1}$  corresponding to the instable area A showing the micro-crack induced by  $\kappa$ -carbide, and (d) the EBSD phase map of the specimen deformed at  $1000 \text{ }^\circ\text{C}$  and  $1 \text{ s}^{-1}$  corresponding to the instable area B showing the necklace structure (A: austenite, F: ferrite,  $\kappa$ :  $\kappa$ -carbide) [182]. (Reproduced with permission from [182]. Copyright 2019 Springer Nature.).

## 8. Future Research Aspects

### 8.1. Alloy Design

The variations of concentrations in Mn, Al, and C alter the phase constituents in lightweight Fe-Mn-Al-C steels and a wide range of tensile properties could be achieved. Although Al acts as an alloying element both for weight saving and microstructure modification, the limitation of additions of Al should be explored to utilize its lightweight function. The major alloying elements, Mn, Al, and C, need to be utilized properly to enhance the properties of the steels. The role of microalloying elements, such as Nb, V and Ti, in the lightweight Fe-Mn-Al-C steels should be further understood and the research of other alloying elements such as Cu, Ni, and Cr are also needed to be involved so as to optimize the overall properties and thus expand the application fields of the lightweight alloys.

### 8.2. Microstructure Design

As different phases possess different features, the coordination is needed during deformation. The stress/strain partitioning among different phases should be investigated and simulation methods might be helpful to assist the microstructure design by investigating the deformation behavior of the steels. The assessment of contributions of different strengthening mechanisms needs to be investigated for different lightweight Fe-Mn-Al-C steels.

Meanwhile, the hardening mechanisms such as the precipitation of  $\kappa$  or B2 ( $\text{DO}_3$ ) should be considered and the conditions to control of the formation of these precipitates still need to be investigated. Measures to effectively utilize these ordered intermetallics as the second phase should be explored further.

### 8.3. Comprehensive Properties

The research work on the comprehensive properties of lightweight Fe-Mn-Al-C steels is still limited. For the applications of the steels in automobile industry, investigations on the impact toughness of the materials are necessary and the formability of the materials such as hole expansion ratio and bending properties need to be evaluated and the research on high strain rate deformation is also necessary. Systematic research on the weldability and coatability of Fe-Mn-Al-C lightweight steels is also needed. To extend the applications of the steels, service properties such as fatigue, hydrogen cracking resistance, and oxidation resistance should be further investigated.

### 8.4. Fabrication Method Development

With high Mn and Al concentrations, fabrication processes such as steel making, casting, and hot rolling are facing challenges. The difficulties in fabricating the materials should be overcome in the industrial production lines. Meanwhile, new fabrication methods need to be explored. For example, 3D printing could be a promising technology for producing the newly designed Fe-Mn-Al-C steels.

## 9. Summary

- (1) In austenitic Fe-Mn-Al-C steels, the microstructure in solution treated condition is a single  $\gamma$  phase or the one with nano-sized  $\kappa'$ -carbide, depending on the compositions and heat treatment schedules. Solid solution hardening, grain size refinement, and precipitation hardening are the basic strengthening mechanisms. Measures to effectively utilize nano-sized precipitates should be explored to increase the strengths and strain hardening rates of the steels.
- (2) Austenite-based duplex Fe-Mn-Al-C steels generally possess high strength with moderate ductility. The strain partitioning between the dual phases should be investigated to guarantee the accommodation of deformation in two phases with different properties so as to achieve higher performance.

- (3) Mn and Al, the major alloying elements of the present alloy, play the opposite role on the phase stabilization of alloys as the austenite stabilizer and ferrite stabilizer, respectively. It is essential to examine the phase constituents of the steels treated with different schedules and their effects on the mechanical properties. The effects of trace elements on the mechanical properties still need to be investigated to modify the microstructures and tailor properties of the materials.
- (4) Mechanical behaviors of the Fe-Mn-Al-C steels at different temperatures and strain rates need to be investigated to meet the requirements for actual applications. Other properties such as fatigue, formability, weldability, and coatability need to be evaluated further to fulfill the requirements in the different practical uses.
- (5) The fabrications of the lightweight Fe-Mn-Al-C steels need to be explored in the conventional production line. New methods to fabricate the Fe-Mn-Al-C steels with relatively high Al concentrations need to be developed.

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