Design of Fully Austenitic Medium Manganese Steels

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Abstract. Due to their higher ferrite potential compared to high Mn twinning-induced plasticity (TWIP) steels, medium Mn steels usually exhibit austenitic-ferritic microstructures, which makes them suitable for third-generation advanced high-strength steel applications. Nevertheless, the strain hardening characteristics of medium Mn steels are inferior to those of fully austenitic high Mn steels. The present work introduces alloy design strategies to obtain fully austenitic medium Mn steels capable of the TWIP effect. To achieve a fully austenitic microstructure, the martensite start temperature is reduced by raising the C concentration to above 1 mass-%, which in turn facilitates the formation of cementite. The formation of cementite during cooling from austenitization temperature is counteracted by alloying with Al. Microstructural examination of slowly-cooled Fe–Mn–Al–C and Fe–Mn–C steels indicated that Al changes the morphology of intergranular cementite from plate-shaped to equiaxed.

1. Introduction
Steels containing raised Mn contents receive an ever-increasing attention among the scientific community and steelmakers. The superior strength and ductility combination of such steels is of particular interest to the automotive industry, where enhanced passenger safety and reduced weight without substantial increase in manufacturing and materials costs drives the industry [1]. Steels with raised Mn levels can be classified as super high Mn steels if containing 22–30 mass-% Mn, high Mn twinning-induced plasticity (TWIP) steels with 12–22 mass-% Mn, and medium Mn transformation-induced plasticity (TRIP) steels with 4–8 mass-% Mn [2].

The occurrence of the TWIP effect in high Mn steels is responsible for their exceptional work hardening characteristics. High Mn steels can work harden to flow stresses in the range of 600–1100 MPa and show tensile elongations in the range of 60–95 %, indicative of their high formability [3]. The presence of high Mn concentrations in TWIP steels impairs their weldability and galvanizability and increases production costs, which are responsible for their limited industrialization [4]. Therefore, many researchers have attempted to decrease the Mn concentration of TWIP steels without significant sacrifice of the mechanical properties [4], [6,7]. Tsuchiyama et al. [7] reported a tensile strength of 1070 MPa and a total elongation of 33 % for an Fe–5Mn–1.2Si–0.1C steel (concentrations in mass-%). Zhang et al. [8] achieved similar tensile properties with an Fe–7Mn–0.23Si–0.14C steel. Tensile strengths and total elongations were lower for an Fe–5.3Mn–1.53Si–0.56Ni–0.01C–0.12(Nb+V+Ti) steel with a reduced austenite stability [9]. The preceding examples of medium Mn steels contain low C concentrations leading to the formation of austenitic-ferritic microstructures. As long as the carbide formation is suppressed and the fraction of austenite and ferrite is fixed, an increase in the C concentration can raise the stability of austenite and lead to enhanced strength-ductility combinations [10]. In order to retain a high fraction of austenite at room temperature (RT), these steels are generally intercritically annealed to
enrich Mn (and C, if present) in the austenite [5,6,9]. When these steels are deformed, the occurrence of deformation-induced martensitic transformation results in high strain hardening rates at the expense of ductility. The stability of austenite and the type of deformation-induced processes can be controlled by adjusting the stacking fault energy (SFE) [11,12]. When the SFE is less than 15 mJm\(^{-2}\), martensitic transformation and the TRIP effect are expected in the austenite. The TRIP and TWIP effects occur simultaneously in the SFE range 15–20 mJm\(^{-2}\). The TWIP effect is dominant in the SFE range 20–35 mJm\(^{-2}\) [2]. Since C and Mn both increase the SFE [13], it should be possible to substitute at least some of the Mn in high Mn steels with C and still retain the TWIP effect. Seo et al. [14] have demonstrated that twinning is indeed a dominant deformation mechanism during RT deformation of an Fe–7.0Mn–1.2C medium Mn high C steel. The present work discusses important considerations in the design of medium Mn high C steels with fully austenitic microstructures and adequate mechanical stabilities at RT.

2. Experimental procedure

Chemical compositions of the medium Mn high C steels studied in the present work are given in table 1. The ingots were produced using a vacuum induction melting and casting facility. The cast ingots were homogenized at 1250 °C for 2 hours under vacuum. The average cooling rate was 8 °C/s between 1250 °C and 650 °C and 4 °C/s between 650 °C and 400 °C. To eliminate carbides possibly formed during N\(_2\) gas cooling to RT, small specimens taken from homogenized ingots were austenitized at 1150 °C for 30 min and then quenched in water. Ferromagnetic phase fractions after different mechanical and thermal treatments were quantified by magnetic flux density measurements in a magnetic field of nearly 3.5 kOe. These measurements were done using a Metis MSAT device equipped with a Lakeshore 480 fluxmeter. Specimens with an approximate length of 10 mm and cross sections of approximately 3.5 × 3.5 mm\(^2\) were used to apply cooling rates of 0.5 °C/s and 0.1 °C/s in a Bähr-DIL805 dilatometer.

| Steel ID | C  | Mn | Al | Fe       |
|----------|----|----|----|----------|
| 8Mn2Al   | 1.44 | 8.0 | 1.9 | Balance |
| 8Mn3Al   | 1.47 | 7.8 | 3.3 | Balance |

3. Results and Discussion

After a discussion of alloy design strategies to obtain fully austenitic medium Mn steels exhibiting TWIP effect, the effect of Al addition to medium Mn high C steels on the microstructure after various thermal and mechanical treatments is discussed.

3.1. Design of fully austenitic medium Mn steels

In order to ensure a fully austenitic microstructure in medium Mn steels, the formation of martensite, pearlite, and cementite must be inhibited by proper alloy design and thermal processing as outlined in the following.

3.1.1. Inhibition of martensite. To obtain a fully austenitic microstructure at RT, the martensite start (\(M_s\)) temperature must be reduced to below RT. Empirical relationships could be consulted to estimate \(M_s\) temperature from the chemical composition, for instance [15]:

\[
M_s = 539 - 423C - 30.4Mn - 7.5Si + 30Al
\]  

(1)

where \(M_s\) temperature is in °C and elemental concentrations are in mass-%. According to equation (1), the addition of C is very effective in reducing the \(M_s\) temperature and suppressing martensite formation. Carbon also has a substantial strengthening effect [16]. Experiences with stainless steels has demonstrated that the stability of austenite with an \(M_s\) temperature equal to RT is too low for an appropriate formability at RT [17,18]. In such cases, maximum tensile elongation is obtained at temperatures higher than RT, where the formation of martensite during tensile deformation is almost entirely suppressed. In other words, to achieve steels with appropriate tensile elongation at RT, the \(M_d\)
temperature (the highest temperature associated with deformation-induced martensite formation) must be near RT [19] in which case the respective \( M_s \) temperature would be subzero or non-existent. Figure 1 shows the variation of \( M_s \) temperature for Fe–8Mn–2Al steels alloyed with up to 1.5 mass-% C. The \( M_s \) temperature was calculated based on equation (1). The hatched area in figure 1 indicates the anticipated temperature range for \( M_s \) assuming that it lies somewhere between 100 °C and 300 °C above \( M_s \) temperature [17,18]. Although relationships such as the one given in equation (1) provide good approximations of \( M_s \) for low-alloy steels containing common alloying elements, their applicability to high-alloy steels containing less common alloying elements such as Al is associated with a raised uncertainty. For instance, although Al is regarded in equation (1) as an \( M_s \)-raiser and it is therefore expected to raise the \( M_d \) temperature, its addition to austenitic stainless steels has been found to decrease the \( M_d \) temperature [20]. Based on figure 1, it is expected that the medium Mn high C steels in the present study (table 1) meet the condition of subzero \( M_s \) temperature and \( M_d \) temperatures in the vicinity of RT.

3.1.2. Inhibition of pearlite. Obtaining fully austenitic microstructures at RT requires that the formation of pearlite during cooling from austenite range is suppressed. This necessitates the addition of hardenability elements to decelerate the kinetics of pearlite formation. Mn has an obvious hardenability effect in steels [21] and its being an austenite stabilizer ensures that it does not impede the formation of austenite at the solution annealing temperature [22]. The microstructures in figure 2 demonstrate the effect of Mn addition to high C steels on the formation of pearlite during continuous cooling from 1150 °C at a rate of 0.5 °C/s. The microstructure of the binary Fe–1.4C steel consists of pro-eutectoid cementite and pearlite only. The reduced fraction of pearlite in the Fe–1.6C–3Mn steel indicates the decelerated pearlite formation upon alloying with 3 mass-%Mn. Increasing the Mn content to 6 mass-% (Fe–1.7C–6Mn steel) results in the full suppression of pearlite. The presence of Mn in medium Mn steels such as the ones dealt with in the present work (table 1) ensures their adequate resistance against the formation of pearlite.

![Figure 1. Calculated dependence of \( M_s \) temperature on C content in Fe–8Mn–2Al–xC steels.](image)

![Figure 2. Microstructure of Fe–(1.4 - 1.7)C–(0 - 6)Mn steels after cooling from 1150 °C at a rate of 0.5 °C/s. Pikral was used as etchant. P and θ denote pearlite and cementite, respectively. Matrix with a bright contrast in (b) and (c) denotes austenite, possibly coexisting with bainite and martensite.](image)
3.1.3. Inhibition of cementite. As discussed in 3.1.1, raised C concentrations are needed in order to achieve subzero $M_s$ temperatures in medium Mn steels. As figure 2 implies, the presence of high C concentrations in steels increases the risk of cementite formation during cooling from the austenite range. Figure 2 also shows that the formation of cementite is not inhibited by the presence of Mn although pearlite formation is clearly suppressed. Therefore, the addition of alloying elements such as Al and Si, which have a recognized decelerating effect on the precipitation of cementite [23,24], becomes necessary. The low solubility of Si in cementite might be responsible for the suppression of cementite during the austempering of Si-added steels [22]. Similar to Si, Al is also insoluble in cementite and therefore decelerates its formation [25]. Accordingly, as the Al content is increased, the C content of retained austenite in the Fe–C–Mn–Al TRIP-assisted steels increases as well [22].

Figure 3. Microstructure of 8Mn2Al and 8Mn3Al steels after water quenching from 1150 °C. Nital was used as etchant. G.B. denotes grain boundaries. The matrix is austenite.

Figure 4. Microstructure of 8Mn2Al and 8Mn3Al steels after homogenization. Nital was used as etchant. G.B. and $\theta$ denote grain boundaries and cementite, respectively. The matrix is austenite.
In accordance with the preceding principles, the medium Mn high C steels used in the present study contain Al. The high affinity of Al for oxygen favors the formation of an oxide layer on the surface of steels. This oxide layer can in turn protect the bulk alloy from further oxidation and act as a barrier against the outward diffusion of C during high temperature solution annealing of high C steels. Furthermore, Al addition reduces the specific weight and can provide an additional tool to adjust the stability of austenite.

![Microstructure of 8Mn2Al and 8Mn3Al steels after cooling from 1150 °C at a rate of 0.1 °C/s. Nital was used as etchant. P, G.B. and θ denote pearlite, grain boundaries and cementite, respectively. The matrix is austenite.](image)

### 3.2. Mechanical and thermal stability of austenite

The microstructure of specimens after austenitization at 1150 °C and quenching in water are shown in figure 3. The microstructure of both steels is austenitic. With light optical microscopy, no obvious sign of cementite formation could be detected even at grain boundaries. The particles with an approximate diameter of 2 μm which are uniformly distributed in the microstructure are either inclusions or carbides of strong carbide-forming elements which might be present in the steels. Magnetic measurements using water-quenched specimens indicated near-zero magnetizations, thereby confirming the absence of pearlite and martensite. Magnetizations were also close to zero for the homogenized specimens which had been cooled at an average rate of 8 °C/s between 1250 °C and 650 °C and 4 °C/s between 650 °C and 400 °C. Small cementite particles were however visible at grain boundaries in this condition (figure 4). Cementite formation was more obvious in the 8Mn2Al than in the 8Mn3Al steel indicating the efficiency of Al in opposing cementite formation. When the cooling rate from 1150 °C was reduced to 0.1 °C/s, the fraction of cementite in both steels was increased dramatically (figure 5). In addition to cementite, some pearlite was also observed in the microstructure of both steels. The fraction of pearlite was lower in the 8Mn2Al steel than in the 8Mn3Al steel. In other words, an increase in the concentration of Al opposes the cementite formation but the simultaneous destabilization of austenite with respect to ferrite favors the formation of pearlite. A comparison of the microstructures in figure 5 with those of the Fe–1.7C–6Mn steel cooled at a higher cooling rate of 0.5 °C/s (figure 2) indicates the effectiveness of Al in inhibiting the cementite formation. Furthermore, Al addition changes the morphology of intergranular cementite from plate-shaped to equiaxed (figure 2 vs figure 5).

To compare the thermal stability of austenite in the water-quenched specimens, they were further cooled in liquid nitrogen. Magnetic measurements confirmed the absence of martensite, namely M, temperature cannot be reached at temperatures as low as -196 °C. These results qualitatively agree with the expectations based on figure 1. To assess the mechanical stability of austenite in the water-quenched specimens, they were given a cold rolling reduction of nearly 50 % at RT and the ferromagnetic phase...
fraction was subsequently determined. The martensite fractions induced by cold rolling were 0.75 vol.% and 1.35 vol.% for the 8Mn3Al and 8Mn2Al steels, respectively. The formation of only small fractions of martensite by cold rolling at RT implies the high mechanical stability of austenite, in spite of the low substitutional content of alloys compared to high Mn steels. In other words, the high stability of austenite is mainly due to C, as if some of the Mn in high Mn steels is replaced by C.

The schematic time-temperature-transformation (TTT) diagram in figure 6 summarizes the effects of Mn and Mn+Al addition on the kinetics of phase transformations in high C steels. Although Mn addition significantly suppresses the kinetics of pearlite formation, mainly due to the destabilization of the ferritic constituent of pearlite, it is not effective against the precipitation of cementite. Therefore, obtaining fully austenitic microstructures in high C steels containing Mn as the only substitutional alloying element would not be possible unless very high cooling rates are applied. In order to reduce the critical cooling rate associated with a cementite-free microstructure, use should be made of alloying elements such as Al and Si. In the schematic of figure 6, the effect of Al on the kinetics of bainite formation was extracted from Ref. [26].

4. Conclusions
1. The kinetics of pearlite formation in high C steels is decelerated by the addition of Mn.
2. Medium Mn steels can become fully austenitic at RT by raising the C content which reduces the Ms temperature. To suppress the formation of cementite in high C steels, however, high cooling rates are necessary.
3. The addition of Al decelerates the kinetics of cementite formation in medium Mn steels and facilitates obtaining a fully austenitic microstructure at lower cooling rates compared to the ternary Fe–Mn–C steels. Furthermore, Al addition changes the morphology of intragranular cementite from plate-shaped to equiaxed.
4. Both of the Al-added medium Mn high C steels studied in the present work exhibited a high resistance against the formation of martensite by thermal and mechanical treatments. Both steels remained fully austenitic after cooling to -196 °C and formed only less than 2 vol.% martensite after applying a cold rolling reduction of nearly 50 % at RT.
5. References

[1] Ghassemieh E 2011 Materials in Automotive Application, State of the Art and Prospects *New Trends and Developments in Automotive Industry* (Croatia) pp 365–94

[2] Lee S and De Cooman B C 2014 Effect of the Intercritical Annealing Temperature on the Mechanical Properties of 10 Pct Mn Multi-phase Steel *Metall. Mater. Trans. A* 45 5009–16

[3] De Cooman B C, Chen L, Kim H S, Estrin Y, Kim S K and Voswinckel H 2009 State-of-the-Science of High Manganese TWIP Steels for Automotive Applications *Microstructure and Texture in Steels* (Springer, London) pp 165–83

[4] Lee C-Y, Jeong J, Han J, Lee S-J, Lee S and Lee Y-K 2015 Coupled strengthening in a medium manganese lightweight steel with an inhomogeneously grained structure of austenite *Acta Mater.* 84 1–8

[5] Lee S and De Cooman B C 2015 Tensile Behavior of Intercritically Annealed Ultra-Fine Grained 8% Mn Multi-Phase Steel *Steel Res. Int.* 86 1170–8

[6] De Cooman B C, Lee S J, Shin S, Seo E J and Speer J G 2017 Combined Intercritical Annealing and Q&amp;P Processing of Medium Mn Steel *Metall. Mater. Trans. A* 48 39–45

[7] Tsuchiyama T, Inoue T, Tobata J, Akama D and Takaki S 2016 Microstructure and mechanical properties of a medium manganese steel treated with interrupted quenching and intercritical annealing *Scr. Mater.* 122 36–9

[8] Zhang Y, Wang L, Findley K O and Speer J G 2017 Influence of Temperature and Grain Size on Austenite Stability in Medium Manganese Steels *Metall. Mater. Trans. A* 48 2140–9

[9] Zou Y, Xu Y B, Hu Z P, Gu X L, Peng F, Tan X D, Chen S Q, Han D T, Misra R D K and Wang G D 2016 Austenite stability and its effect on the toughness of a high strength ultra-low carbon medium manganese steel plate *Mater. Sci. Eng. A* 675 153–63

[10] Furukawa T, Huang H and Matsumura O 1994 Effects of carbon content on mechanical properties of 5%Mn steels exhibiting transformation induced plasticity *Mater. Sci. Technol.* 10 964–70

[11] De Cooman B C, Estrin Y and Kim S K 2018 Twinning-induced plasticity (TWIP) steels *Acta Mater.* 142 283–362

[12] Allain S, Chateau J-P, Bouaziz O, Migot S and Guelton N 2004 Correlations between the calculated stacking fault energy and the plasticity mechanisms in Fe–Mn–C alloys *Mater. Sci. Eng. A* 387–389 158–62

[13] Saeed-Akbari A, Mosecker L, Schwedt A and Bleck W 2012 Characterization and Prediction of Flow Behavior in High-Manganese Twinning Induced Plasticity Steels: Part I. Mechanism Maps and Work-Hardening Behavior *Metall. Mater. Trans. A* 43 1688–704

[14] Seo E J, Kim J K, Cho L, Mola J, Oh C Y and De Cooman B C 2017 Micro-ductility of medium Mn austenitic steel: Perfect dislocation plasticity and deformation twinning *Acta Mater.* 135 112–23

[15] Mahieu J, De Cooman B C and Maki J 2002 Phase transformation and mechanical properties of si-free CMnAl transformation-induced plasticity-aided steel *Metall. Mater. Trans. A* 33 2573–80

[16] Bouaziz O, Zurob H, Chehab B, Embury J D, Allain S and Huang M 2011 Effect of chemical composition on work hardening of Fe-Mn-C TWIP steels *Mater. Sci. Technol.* 27 707–9

[17] Huang Q, Volkova O, Biermann H and Mola J 2017 Tensile elongation of lean-alloy austenitic stainless steels: transformation-induced plasticity versus planar glide *Mater. Sci. Technol.* 33 1224–30

[18] Mola J, Ullrich C, Kuang B, Rahimi R, Huang Q, Rafaja D and Ritzenhoff R 2017 Austenitic Nickel- and Manganese-Free Fe-15Cr-1Mo-0.4N-0.3C Steel: Tensile Behavior and Deformation-Induced Processes between 298 K and 503 K (25 °C and 230 °C) *Metall. Mater. Trans. A* 48 1033–52

[19] Mola J 2017 Considerations in the Design of Formable Austenitic Stainless Steels Based on Deformation-Induced Processes *Austenitic Stainless Steels - New Aspects* (Croatia) pp 7–28
[20] Rahimi R, Ullrich C, Klemm V, Rafaja D, De Cooman B C, Biermann H and Mola J 2016 Influence of Al on the temperature dependence of strain hardening behavior and glide planarity in Fe–Cr–Ni–Mn–C austenitic stainless steels Mater. Sci. Eng. A 649 301–12

[21] Calcagnotto M, Ponge D and Raabe D 2012 On the Effect of Manganese on Grain Size Stability and Hardenability in Ultrafine-Grained Ferrite-Martensite Dual-Phase Steels Metall. Mater. Trans. A 43 37–46

[22] Kuziak R, Kawalla R and Waengler S 2008 Advanced high strength steels for automotive industry Arch. Civ. Mech. Eng. 8 103–17

[23] Samek L, Moor E D, Penning J and De Cooman B C 2006 Influence of alloying elements on the kinetics of strain-induced martensitic nucleation in low-alloy, multiphase high-strength steels Metall. Mater. Trans. A 37 109–24

[24] Kozeschnik E and Bhadeshia H K D H 2008 Influence of silicon on cementite precipitation in steels Mater. Sci. Technol. 24 343–7

[25] Liu S, Zhang F, Yang Z, Wang M and Zheng C 2016 Effects of Al and Mn on the formation and properties of nanostructured pearlite in high-carbon steels Mater. Des. 93 73–80

[26] Mertens A, Jacques P j., Sietsma J and Delannay F 2008 Relative Influences of Aluminium and Silicon on the Kinetics of Bainite Formation from Intercritical Austenite Steel Res. Int. 79 954–9

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