# A Review on Influence of alloying elements and process parameter on the structure properties correlation of Mn based Maraging Steel

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Abstract- Maraging steels are a type of steel that was designed with high strength to provide exceptional performance. Despite their high cost, these steels had a high amount of toughness with high strength, which was one of the intended needs of these steels. In order to reduce the cost and improve the ductility of maraging steels without sacrificing strength, the high content of costly Ni element is partially substituted by inexpensive Mn to achieve metastable austenite. This will result in a reduction in vehicle weight, which will boost efficiency therefore overall society will be benefit. Given the information that Fe-Mn alloys can exhibit TRIP phenomena at a specific composition regime in terms of manganese content (15 wt.%) and that TWIP is indutable at a greater level of manganese (28 wt.% and above). In the present paper, several aspects will be reviewed related to Mn based maraging steels including a brief details of its development, role of alloying elements in maraging steel, Mn embrittlement in Fe-Ni-Mn and Fe-Mn alloys, reverted austenite, influence of process and composition on microstructure and mechanical properties.

## 1. INTRODUCTION

To be effective in their applications, steels must have a good mechanical balance (tensile strength and elongation). They are seen as the important materials of the future, particularly in the automotive sector, for light weight methods and the associated reductions in fuel and vehicle emissions. Due to the enormous commercial demand, industry have made significant advancements in the creation of advanced high strength steel (AHSS) grades and related processing technology.

Advanced high-performance steels with high strength and ductility are in high demand for lowering weight and improving energy efficiency of structural components, such as those used in bridges, ships, and vehicles, etc [1]. The automotive industry is under increasing pressure to meet rising demands for performance, environmental responsibility, and fuel efficiency at cost-effective prices. Steel, aluminium, magnesium, plastics, and polymer composites are just a few of the material businesses that are working to

meet the increasing demands of the automotive industry. Advanced plastics and polymer composites have improved the design, usability, and safety of automobiles for decades while also lowering vehicle weight and providing superior value to customers [2]. New regulations, shifting customer expectations, and recent technology advancements are motivating the automobile industry to continue expanding its usage of advanced plastics and polymer composites to satisfy now-a-days challenges and opportunities. Composite materials provide a way to significantly reduce a vehicle's weight while maintaining strength standards. Given how much environmental issues are affecting both the chemical and the automotive industries today, engineered plastics are quickly replacing them as the preferred material. More lightweight plastics and polymer-based components are being used by automakers to maintain maximum fuel efficiency [2]. According to national statistics released by the UK Department of Energy and Climate Change in March 2013, automobiles account for 24% of the country's carbon emissions [3]. In terms of improving fuel efficiency, reducing carbon emissions, and improving overall driving performance of the automotive vehicles, lighter materials are used in the construction of automobiles. The American Iron and Steel Institute claims that the Life Cycle Assessment (LCA) on automotive materials demonstrates that steel, which currently constitutes about 60% of the average vehicle, is the most environmentally friendly automotive material when compared to other competing automotive materials [4].



Figure 1 2010 Light Vehicle Material Content adapted from Steelworks online [4]

The most common type of HSS, high-strength low-alloy (HSLA) steels have a microstructure of fine ferrite grains reinforced by carbides and/or nitrides of Ti, V, or Nb. Their strength typically ranges between 700 and 800 MPa. These types of steels have been used extensively in automotive applications because they can be produced using straightforward processing pathways. More complex alloy systems have been developed, such as Dual Phase (DP) steels, Transformation Induced Plasticity (TRIP) steels, Twinning Induced Plasticity (TWIP) steels, and Martensitic (MART) steels, in order to further improve the combining properties of strength and ductility by added other alloying elements. Global tensile strength-elongation profile of those steel families is displayed in Figure 2 [5]. Dual phase steels are mostly composed of soft ferrite with islands of hard martensite distributed throughout. Their strength is determined by the amount of martensite present, as well as the morphology and distribution of the martensite [6-9]. TRIP steels are multiphase grades that use specifically planned alloying additions and processing routes to retain a certain proportion of austenite inside a ferritic matrix down to room temperature. When this metastable austenite is subjected to plastic deformation, it changes from austenite to martensite, which alters the volume and shape of the resulting martensite and matrix to account for the transformation misfit and increases the ductility [10-12]. The typical microstructure of TWIP steels at room temperature is completely austenitic. At room temperature, TWIP steels typically exhibit a completely austenitic microstructure. A superior balance of strength and ductility is achieved thanks to the development of mechanical twins of austenite during deformation, which causes high strain hardening and inhibits necking [13-15]. Maraging steels are the most extensively researched subgroup of the martensitic (MS) steels group, which is another subgroup of HSS. Because the martensite matrix is severely strained after quenching and is further strengthened by precipitation during ageing, maraging steels are noted for their extremely high strength. These precipitates utilise the Orowan mechanism to serve as very effective barriers against dislocation motion.
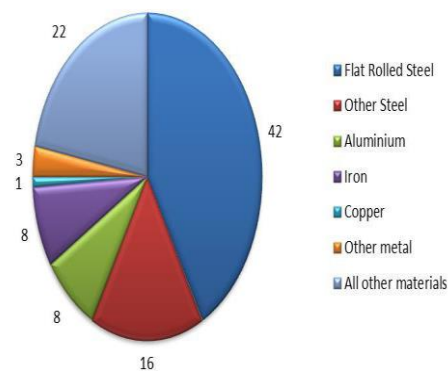
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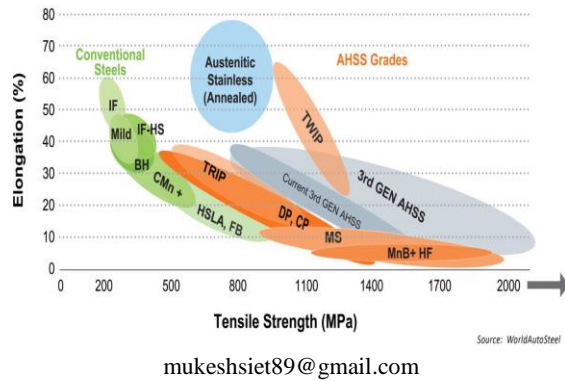*Corresponding author. E-mail addresses: mukeshsiet89@gmail.com*

Figure 2 Global tensile strength-elongation profiles for various kinds of steels [5].

It is widely acknowledged that deformation-induced martensitic transformation improves the mechanical characteristics of TRIP steels. To keep austenite at room temperature, a high degree of carbon content is traditionally added; however the poor weldability caused by high carbon addition limits its applicability in the automotive industry. As a result, Mn is utilised in place of carbon as another efficient austenite stabiliser. A low-carbon TRIP steel with significant austenite content (0.1 wt% C, 8 wt% Mn) was initially reported to have exceptional toughness by Goldshtein et al. in the 1960s [16]. Since then many researches on medium Mn steels have been carried out [17-20]. In a study by Luo et al. [21], it was shown that a 5Mn steel with more than 30 vol.% austenite had a total elongation (TE) of 31-44% and an ultimate tensile strength (UTS) of 1-1.5 GPa. On the other hand, it is generally accepted that the superior mechanical properties in maraging steels are attributed to the formation of strengthening nano-precipitates upon aging. In fact, it is anticipated that martensite will partially revert to austenite as the ageing process typically takes place at the two-phase field. Evidence suggests that austenite reversion can have a considerable impact on the mechanical characteristics [22-24]. The most traditional type of maraging steel is nickel-based (18 wt% Ni) with significant additions of Co (8 wt%), Ti, Cr, and Mo. Due to the dramatic decline in Co availability since the 1960s, numerous researchers have concentrated on creating cheaper alternatives to costly Co and Ni-containing maraging steels. Studies on Mn-based maraging steels, among others, have made significant strides toward minimising the strength loss brought on by the absence of Co and Ni.

A plan to combine the two strengthening mechanisms in the form of Fe-Mn alloys was suggested based on the discussion above. According to this strategy, a microstructure including hard martensite strengthened by nano-precipitates and ductile austenite in which transformation-induced plasticity can occur should be attained. Low carbon steels with 0.01 wt% carbon, a Mn content of 9–12 wt%, and minimal additions of Ni, Ti, and Mo (1-2 wt.%) were developed by Raabe et al. [22]. Those steels showed a simultaneous rise in strength and ductility, defying conventional wisdom that suggests that a gain in strength is accompanied by a fall in ductility (Figure 1.2). A 12 weight percent Mn maraging TRIP steel showed UTS of ~1300 MPa and a 21% tensile elongation.

This outstanding accomplishment creates a new strategy for creating ultra-high steels with comparatively low alloying costs. However, there are still several unanswered problems regarding this novel type of steel, such as the nature and composition of precipitates, which are under discussion. It is unclear how austenite reversion and precipitation are caused by their respective production methods. Given that the reverted austenite in these steels has a different chemical composition, shape, and grain size than the retained austenite does, the deformation mechanism of reverted austenite in these steels is supposed to be different from that of retained austenite. It is also worthwhile to research the strengthening mechanism of these kinds of steels. As a result, a thorough investigation of these steels is required, and the results of the investigation can be used to propose the ideal microstructure and related processing routes.

## 2. THE DEVELOPMENT OF MN TRIP STEELS
The development of advanced high strength steel (AHSS) for an automotive application has been driven by two principal motives. The first motive is to reduce fuel consumption and automobile emissions through mass reduction. The second factor is the growing demand for vehicle comfort, safety, and speed. In this regard, AHSSs with a high strength-to-weight ratio are significant materials for car manufacturers [5, 25, 26]. The first-generation automotive-grade AHSSs (Figure 1.2), such as DP steels, TRIP steels, and complex-phase (CP) steels, have been thoroughly studied in the past [5, 26]. These steels feature multi-phase

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microstructures made up of varying proportions of bainite, martensite, and carbon-rich austenite. Twinning-induced plasticity (TWIP) steels, which are the second-generation AHSSs, are steels. These high Mn (> 15 wt%) austenitic-based steels produce a combination of strength and ductility superior to that of the first-generation AHSSs [14, 15], but their use in the automotive industry was constrained by high alloying additions (such as Mn, Al, and C) and complex processing technology [27]. Third-generation AHSSs with medium Mn concentrations (5-12 wt%) have recently been conceived and produced. Adjusted intercritical tempering yielded different proportions of metastable austenite with fine grain size in various steels [20, 28, 29, 30]. Medium Mn steels with sub-micro austenite grains have been reported to have high ultimate tensile strength and outstanding total elongation [31, 26, 30]. This is because austenite plasticity is dependent not only on mobile dislocations, but also on two other plastic modes, namely strain-induced martensitic transition and mechanical twinning. This feature is important for the design of medium Mn steels with austenite. Recent research on steels with 5-7 wt% Mn found that intercritical tempering can be improved to yield elongation of up to 45%. But as of now, this steel grade's yield strength barely exceeds 1 GPa, which is evidently insufficient to meet demand for the many infiltration applications of hole-expanded components in the automobile sector [30].

## 3. THE DEVELOPMENT OF MARAGING STEELS

Maraging steels are an ultrahigh-strength martensitic steel that is hardened by intermetallic precipitates rather than universal carbide precipitates. As a result, the carbon content of maraging steels is kept as low as commercially feasible [32]. The term 'maraging' refers to the aging of martensite. The martensite with low carbon content is very soft yet severely dislocated, and the supersaturated martensite is formed by rapid cooling from the austenite phase field. This microstructural characteristic is thought to improve hardenability, toughness, and formability prior to aging [33], as well as providing an excellent nucleation condition for precipitation. Intermetallic precipitates form as a result of the subsequent aging treatment. The nature of precipitates largely depends on the alloying compositions and aging parameters. On the other hand, if a proper balance of strength and toughness is required, excessive softening of the martensite matrix by recovery during aging should be avoided.

The development of maraging steels began in 1962 with the publication of a report by Decker et al. [34] on 18 wt.%. Ni steels with low carbon content (0.01-0.03 wt.%) but considerable amounts of Co (8 wt.%) and Mo (5 wt.%) respectively. Tensile strengths of up to 2068 MPa, 12% elongation, 60% area reduction, low ductile-brittle transition temperature (DBTT), and notch tensile strengths of around 3034 MPa were attained, according to previous research. These alloys also had outstanding stress-corrosion resistance and weldability [34]. The ensuing studies resulted in the development of well-known maraging systems, such as 18Ni (200 (ksi)), 18Ni (250), and 18Ni (300) alloys. However, the expensive alloying additions of maraging steels limited their applicability in the most important industries. The dramatic decline in cobalt availability from 1978 to 1980 encouraged the exploration of alternatives to Co-containing maraging steels. Intensive attempts were made to reduce the concentration of costly alloying elements like Co and Ni. This led to the emergence of a new class of Co-free maraging steels that contain Fe, Ni, Mo, and Ti. These Co-free steels often displayed inferior mechanical qualities than Co-containing steels, although their mechanical properties appeared to be adequate for normal maraging steel applications [35, 36]. Furthermore, because the equilibrium phase diagrams of Fe-Ni and Fe-Mn systems are similar, partial substitution of Ni by Mn was used in the development of lean maraging steels. After 1 hour of aging at 550 °C, a Fe-20.8Ni-2.13Mn-0.8Ti maraging steel with a yield strength (YS: 1371 MPa) and total elongation (TE: 18.8%) was reported to have a remarkable combination of strength and ductility [37]. However, steels with high Mn concentrations have been observed to suffer from embrittlement after short periods of aging.

The main strength of maraging steels is generally attributed to the dense precipitation of fine intermetallic compounds in a soft but heavily dislocated martensite matrix. When aging at high temperatures or for a long time, reversed austenite may occasionally form. Reverted austenite's impact on the mechanical characteristics is controversial. The formation of metastable reverted austenite following rapid cooling from intercritical tempering is generally regarded as a method of enhancing the toughness of

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quenching and partitioning (Q&P) steels for low-temperature use (holding in the temperature range where the austenite and ferrite phases coexist) [30, 31, 38]. Maraging steels are well-known for their exceptional combination of ultrahigh strength and acceptable ductility. Due to the low carbon content, they have good machinability and weldability when solution-treated. They also possess a good strength-to-weight ratio and dimensional stability with aging. The creation of maraging steels has been the subject of various studies due to its array of desirable features, which make it a class of material that has great potential. More recently, a newly developed 17 wt.% Mn maraging TRIP steel with minor additions of Ni, W, Mo, Al, Cu and B was reported to have an excellent combination of strength and ductility (UTS (ultimate tensile strength): 1.4 GPa, TE (total elongation): 27% due to precipitation strengthening mechanism and TRIP effect of austenite). But there are still some unanswered problems regarding these new minerals, such as the type of precipitates and how retained/reverted austenite forms and contributes to strengthening [108].

### 3.1. Role of Alloying Elements in the Design of Mn based Maraging steels

To achieve ultra-high strength steel, we have to obtain a very fine (nano scale) precipitate microstructure in the austenite phase. This is accomplished through grain refining and other mechanisms of strengthening. Grain refining by alloying element has a significant influence in strength improvement. Each alloying element has its unique influence on phase stabilisation; some tend to stabilise austenite while others are ferrite stabilisers, but they all give forth various mechanical qualities as a whole. The microstructure and phases have delicate on mechanical properties. As a result, each alloying component has a unique impact on the resulting alloy's strength and behaviour. As a result, the mechanical properties are determined by the alloying elements, and variations in these mechanical properties can be produced by altering the shape, size, distribution, and volume percent of the alloying elements.

### 3.1.1 Carbon

Carbon plays a key role in the formation of retained austenite, which is stable at room temperature, in multiphase steels by enriching austenite. The stabilized retained austenite contributes to the good mechanical properties of transformation induced plasticity (TRIP) steels. Carbon also aids in the precipitation strengthening of steels. It produces carbides of other alloying elements (W, Mo, V, and Nb) that precipitate in the nature of precipitates and are finely distributed in the bulk. However, increasing the carbon content above 0.25% may induce weldability concerns [39] and further generates martensite, which is not ideal for uniform elongation. For precipitation strengthening, carbon levels below 0.1 weight percent are typically sufficient.

### 3.1.2 Silicon

With an improvement in the activity coefficient of C in austenite and ferrite, silicon has a significant impact on the strengthening of the overall solid solution. However, Si has an affinity for reducing C solubility in the ferrite phase. Cementite formation is inhibited, while carbon partitioning from bainite is stimulated in the residual austenite. There are a few observations [40-41] in the recent literature that demonstrate Si has no solubility in $Fe_3C$, therefore cementite nucleation and growth can occur only after Si diffusion is complete. Because this process takes time, the rate of cementite development is reduced. Si additions of less than 0.001 wt% have no discernible effect on the alloy. However, increases more than 0.5 wt% cause surface quality and galvanising issues [42].

### 3.1.3 Aluminium

Aluminum is similar to silicon in that it inhibits cementite production and improves the partitioning of carbon to residual austenite. It has no impact on the plating qualities of the alloy and not only works as a potent deoxidizer but also aids in increasing C activity coefficient and, eventually, C solubility in ferrite. Contrary to Si formation, bainite formation is encouraged, which shortens the typical industrial processing time. However, unlike Si, Al does not contribute to solid solution strengthening; it also raises the martensite start and finish ($M_s$ and $M_f$) times above room temperature [43]. Girault et al. [44] compared Al alloyed TRIP steels to Si alloyed TRIP steels and determined that Al was similarly good despite the significant drop in strength to ductility balance. Amounts of Al larger than 3 wt% increase surface defects and also decrease overall elongation and stretch flangeability, despite the fact that it has been demonstrated that amounts of Al less than 0.50 wt% do not significantly affect the properties of the alloy [42].

### 3.1.4 Copper, Nickel

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In general, copper is regarded as an impurity element that is trapped in the bulk of steel, resulting in hot shortness. At normal temperature, copper has no solubility in steel and will not form carbide. The strength, hardness, hardenability, and toughness of solid solutions are generally enhanced by nickel addition. In comparison to the ε-phase, both copper and nickel operate as γ-phase stabilisers. Cu and Ni atoms form clusters in the austenite phase, forming Cu-Cu and Ni-Ni bonds rather than Cu-Fe and Ni-Fe bonds. [45] Showed that the inclusion of Ni promotes the development of the [(Fe, Mn) Al] phase with Al. Additionally, Kim et al. [47] provided further confirmation that Ni addition to Fe-10Al-15Mn-0.8C-5Ni steel promotes precipitation of the intermediate [(Fe, Mn) Al] phase in the austenite matrix. On the mechanical behaviour of TWIP steel, Lee et al. [47] looked into the impact of Cu addition. They have also noted that the addition of 2 weight percent Cu to Fe-Mn-C-Al TWIP steel can enhance γ from 22.8 to 26.0 mJ/$m^2$. The simultaneous addition of Cu impairs the kinetics of twin formation and changes the type of abnormalities on the stress-strain curve. By changing the serration and increasing the number of Portevin-Le Chatelier (PLC) bands, the dynamic stain aging effects of Cu are quite similar to those of N for Fe-Mn-C-Al based TWIP steels.

### 3.1.5 Tungsten

For the purpose of hardening and strengthening solid solutions, tungsten addition is necessary. For the most part, its addition to steel works well as a ferrite former to improve the mechanical qualities. To a certain extent, tungsten dissolves in the gamma iron phase and is still present in the solution as the gamma phase transitions to the alpha phase. Steel is more easily hardenable when tungsten is present in dissolved form. A few specific ratios of tungsten, chromium, and vanadium tend to lessen the propensity of steel to fracture and deform during heat treatment [48]. When tungsten is added to steel, tungsten carbide is also formed, and when tungsten is added along with other elements that can form carbides, such as Mo, Cr, and V, complex carbides are more likely to form, increasing the amount of undissolved carbides in the steel matrix. For the same carbon concentration, it has been discovered that tungsten produces a larger carbide volume than the other alloying elements [49]. Since tungsten is potent carbide forming, the carbides that are produced are exceedingly hard, increasing the

steel's hardness and yield strength but lowering its elongation value [50]. The primary function is to reduce the rate of grain growth during the heat treatment process by distributing fine and very fine carbide precipitation in the microstructure, which is capable of preserving finer grain steels. [51].

### 3.1.6 Titanium

In the presence of carbon and/or nitrogen, Ti, like V, Nb, and Mo, forms composite carbides or carbonitrides. TiCN, like NbCN, is significantly less soluble in austenite than VCN, according to solubility products. As a result, TiC forms at significantly greater temperatures than VC. This is the main factor that makes titanium carbides more efficient for grain refining than VC. While TiC generated at lower temperatures is more effective at strengthening precipitation, strain-induced TiC is significant in the refinement of grains. The number of precipitates generated at compositions less than 0.02 wt% is insufficient to significantly contribute to precipitation strengthening. Ineffective for either grain refining or precipitation strengthening, compositions greater than 0.4 wt% tend to enhance coarsening of the precipitates [42].

### 3.1.7 Molybdenum

As the slope of the $A_3$ line increases, greater freedom is provided during the heat treatment processes, which is where molybdenum plays a crucial role. It influences the strengthening of solid solutions. When combined with Nb and C, Mo is also recognized to play a significant role in the creation of fine precipitates [52, 42]. When the steel is microalloyed with Ti, Mo, Al, and C, JFE steels [42] have strengths up to 780MPa and excellent ductility is possible. The TiAl(Mo)C composite carbides are finely dispersed in ferrite, which accounts for the strength. Park et al. [83] obtained similar findings when Ti was replaced with Nb. In contrast, additions of less than 0.1 wt% had no substantial effect on precipitation strengthening. However, it has been determined that the saturation point is 0.8 wt%, after which more Mo just increases manufacturing costs without improving the value of the steel [42].

### 3.1.8 Manganese

Mn has a minimal impact on solution strengthening in austenite and a substantial impact in ferrite. By lowering the temperature at which austenite decomposes during cooling and aiding ferrite grain refinement and size reduction, Mn aids in enhancing

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the strength and toughness after rolling. When Mn is added to austenite, the rate of work hardening increases, and at normal concentrations, it slows the rate of work hardening in ferrite. Steels containing Mn are stiffer when rolling and forging because Mn tends to increase resistance to deformation [54, 55].

Mn improves the response of steel to quenching, which also affects the transformation temperature, and Mn is weak carbide former. Both of these properties are desirable during steel heat treatment. Another advantage of Mn, like Ni, is its ability to stabilise austenite. Because Mn is not as potent as Ni in terms of its capacity to stabilise austenite, more Mn is required to have the same result. But Mn has the clear advantage of being very cost-effective. By combining Mn with nitrogen, which may also stabilise austenite, the effect of Mn for generating austenite is increased. Depending on the steel grade and the finished product, Mn also increases hardenability rate, which can be used for significant advantage for improving mechanical properties. Mn has a less significant impact than C on the increase of hardenability and tensile strength. Like no other alloying element, Mn has the capacity to reduce the critical cooling rate needed during the hardening process, increasing the hardenability of steels more effectively. However, embrittlement in steels is a common occurrence when C and Mn are both present at significant concentrations in the alloy's chemistry. Mn somewhat boosts ferrite strength when hardness penetration increases during the quenching process as the critical quenching speed is slowed down. This increases the stability of the steel throughout the quenching process. Steels containing Mn can thus be quenched in oil rather than water with minimal crack susceptibility due to the lower thermal shock of quenching [54- 58].

Manganese's main function in TRIP steel is to stabilise austenite and to lower the temperature at which cementite begins to precipitate. Along with increasing carbon partitioning to retained austenite, it prevents cementite from forming [39]. Additionally, the presence of Mn in ferrite reduces the carbon activity coefficient, enhancing the solubility of carbon in ferrite. Mn plays a critical role in lowering the temperature at which austenite transforms into ferrite, which prevents cementite from precipitating at ferrite grain boundaries and refines the resulting pearlitic structure. Steel's strength and hardness are influenced by the pearlite's volume fraction and grain size. When

compared to C, N, P, and Si, Mn is one of the alloying elements that contribute to the solution hardening of ferrite, but its impact is minimal. The austenite changes into structures with great strength, similar to bainite and martensite, if the cooling is accelerated through quenching. Mn actively enhances deep hardening for heat treatable steels by suppressing austenite transformation. The transformation temperature and eutectoid C concentration are both decreased by Mn. All of this explains why the Mn steels, where transformation is meant to be controlled, cover such a wide range. Mn has a propensity to lower 'Ms' temperature; therefore steels with high Mn levels have residual austenite. It is an important alloy element in flat rolled sheets, and its importance in lowering the transformation temperature is why its addition promotes finer grain size in rolled and normalised conditions. As grain size reduces, yield strength increases, which improves impact properties. An additional advantage of increasing pearlite content with increased Mn concentration and fixed C content is that. As a result, the strength is ultimately increased without sacrificing weldability. Even Mn can produce carbide alike to cementite, which does not undergo secondary hardening during tempering. The susceptibility to temper embrittlement is enhanced by the presence of Mn in concentrations greater than 0.30%. The crucial temperature range must therefore be carefully observed while tempering [43, 54, 55, 59].

### 3.2 Effect of Mn on the precipitation in Fe-Ni-Mn-Ti-Al steels

As mentioned in Section 3.1.8, efforts to partially replace Ni with a lean element like Mn, which offers similar effects upon the austenite-martensite transformation, have resulted in the development of Fe-Ni-Mn-Ti-Al alloys. A significant acceleration of age hardening is associated with an increase in Mn content at the expense of Ni content. An Fe-20Ni-1.8Mn-1.5Ti-0.59Al alloy aged at 550 °C showed a considerable rise in hardness by about 200 VHN within 5 s, and the yield strength increased by around 215 MPa to above 900 MPa [60]. Before plate-shaped $(Ni,Fe)_3Ti$ and spheroidal $(Ni,Fe)_3(Al,Mn)$ precipitates formed after ageing for about 60 seconds, co-clustering of the Ti+Al and Mn+Fe was first noticed [61]. Ti has the fastest diffusion rate among these four alloying elements, followed by Mn and Al, and Ni has the slowest. As a result, Ti-rich clusters normally acted as nuclei for the formation of η-Ni3Ti

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precipitates [61]. The (Ni,Fe)$_3$(Al,Mn) precipitates, on the other hand, were primarily observed in two sites: uniformly within the matrix or the periphery of plate-shaped (Ni,Fe)$_3$Ti particles, which revealed that Al and Mn were segregated at the interface between the (Ni,Fe)$_3$Ti phase and surrounding matrix (Figure 3). This segregation of Al+Mn atoms was explained by the high thermodynamic affinity between Al and Mn [62, 63, 64] and the propensity of Mn atoms in steels to segregate on the interfaces [65, 66].
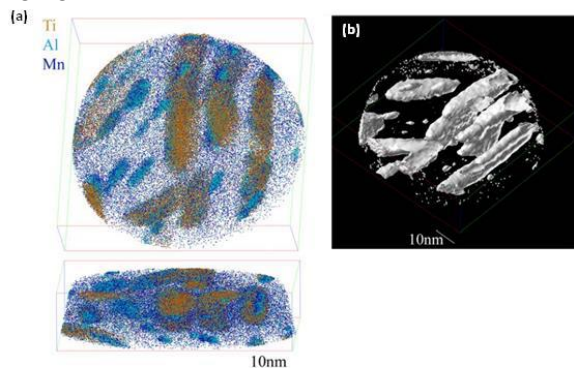


Figure 3 (a) Three dimensional atom probe tomographic reconstructions of rod-shaped precipitates after treated for 3600 seconds and (b) corresponding Ni+Ti isoconcentration surfaces [61].

## 4. THE Mn EMBRITTLEMENT IN Fe-Ni-Mn AND Fe-Mn ALLOYS

As was noted in Section 3.1, efforts have been made to create less expensive substitutes for the typical Ni-based maraging steels. A series of Fe-Ni-Mn steels were the focus of experimental experiments, which demonstrated that these steels exhibited a notable maraging strengthening due to the creation of fine f.c.t. θ-NiMn phase [67, 68]. Some Fe-Ni-Mn alloys have been discovered to produce highly fine and thermally stable Ni$_3$Ti precipitates when Ti is added [37]. Additionally taken into account as potential precipitation-strengthening contributors were other phases including MnNi, fcc Ni$_3$Mn, and ordering [69, 70]. Thus, this grade of steels was once considered to be a promising substitute for traditional maraging steels.

Unfortunately, severe grain-boundary embrittlement was found in Fe-Ni-Mn steels at the start of the aging treatment [65, 71, 72, 73]. Many studies have been conducted to determine the cause of grain boundary embrittlement, and certain theories have been offered.

It is well acknowledged that grain boundaries serve not only as barriers to dislocation motion but also as zones for interface segregation. Raabe et al. [74] outlined the following possible behaviours for the solute segregation in more detail: (i) the segregation may strengthen the grain boundary by increasing coherence and preferential bonding at the interface; (ii) the reverse may weaken the grain boundary by increasing incoherence and unfavourably directed bonding at the interface;(iii) Phase transformation may take place at the grain boundaries (grain boundary phase transformation); (iv) At the decorated interface, one or more second phases may form (grain boundary precipitation); (v) Discontinuous precipitation could possibly be encouraged.

Squires and Wilson [65] proposed that the intergranular fracture seen in a Fe-12Ni-6Mn alloy was caused by embrittling components such as Mn and P segregating to prior austenite grain boundaries. A subsequent investigation on a comparable alloy showed that Ni and N were also discovered near the fractured grain boundaries [71]. On the fracture surfaces of a Fe-8Mn-7Ni alloy in the embrittled condition, Heo and Lee [73] observed the segregation and desegregation of Mn at the previous austenite grain boundaries. They noted that the embrittlement and de-embrittlement of the alloy were directly related to the segregation and desegregation of Mn, but they also concluded that given the low enrichment level of Mn, such severe embrittlement was not likely to be primarily due to the Mn segregation alone [62].

Suto and Murakami, on the other hand, stated that the small segregation of Mn and Ni at grain boundaries would not lead to the transition to brittle state [72]. Regarding the relationship between Mn segregation at grain boundaries and mechanical behaviours, Raabe and his colleagues [74] have a different perspective. Due to the low mutual misorientation between adjacent lath boundaries, they thought that embrittlement was connected to the susceptible crack penetration [74]. Unfortunately, a lot of lath boundaries were present because these were the main interfaces in quenched martensite. However, because there was no fluctuation in the bulk grain free energy, the main thermodynamic reason behind solute segregation at interfaces was to reduce the interface free energy. Materials with reduced grain boundary energy were thought to have a finer and more stable grain size because the interfacial free energy was reduced, which in turn decreased the

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driving force for grain growth. As a result, they suggested that, assuming the particular kind of segregation does not result in grain boundary embrittlement, solute segregation at grain borders helps to both strength and toughness by optimising grain size.

On the other hand, other research hypothesised that the intergranular embrittlement in the Fe-Ni-Mn system was caused by the interaction of mobile dislocation with precipitates near grain boundaries. These researchers maintained that the primary cause of grain boundary failure in Fe-Ni-Mn maraging steels was precipitation on the grain boundary. When ageing a Fe-10Ni-5Mn alloy at 753 K for a brief period of time, Lee et al. [75] discovered θ-NiMn intermetallic precipitates at the grain boundaries, and this sort of precipitate was previously thought to be the cause of the embrittlement. The association between intergranular failure and precipitation reactions at grain boundaries was further validated in investigations later done by Nedjad et al. [76, 77] on Fe-10Ni-7Mn maraging steels. They also highlighted how the development of coarse precipitates at grain boundaries led to the establishment of soft, solute-depleted precipitate-free zones. The microcracks were then initiated at negligible macroscopic strains due to the large strain localization caused by the weakening of the grain boundary. Alternately, Mun et al. [69] observation of austenite particles precipitating at grain boundaries in Fe-7Ni-8Mn steel led them to hypothesis that the austenite-ferrite interface enhanced intergranular fracture. Wilson and his colleagues [78, 79] countered this claim, claiming that while precipitates and reverted austenite were certainly the main causes of Mn segregation to earlier austenite grain boundaries, the segregation of Mn itself still acted as the main cause of grain boundary embrittlement. Binary Fe-Mn alloys containing 4-10 wt% Mn and having a soft but extensively dislocated lath martensite after quenching have microstructural properties similar to Fe-Ni-Mn maraging steels. Brittleness was discovered in Fe-Mn alloys by mechanical behaviour studies [79, 80].

## 5. AUSTENITE REVERSION IN Mn BASED MARAGING STEELS

Engineering steels, as previously mentioned, suffer from the fact that an increase in strength typically results in a decrease in ductility. Their use as structural materials is generally constrained by this inverse strength-ductility connection. Dual-phase microstructures, including hard martensite with ductile ferrite or austenite that can coordinate during deformation, are created as a solution to this issue. The inclusion of metastable austenite has also been shown to boost strength further due to deformation-induced martensitic transition under external load. This kind of steel is currently widely used in the automotive industry's body frame industry. Additionally, dual phase steels have favourable properties for a variety of applications, including the automotive industry, structural materials, etc., such as a relatively high strength to weight ratio, low yield-to-ultimate strength ratio, high initial work hardening rate, and good formability [81]. There are two common processing methods for stabilising the austenite phase in steels. One method, known as retained austenite, is being fully researched for TRIP steels [82, 83, 39, 84]. Another method, known as reverted austenite, develops austenite as a result of partial reversion from martensite during aging of two-phase structures at low temperature [85, 86]. When compared to retained austenite, reverted austenite appears to work with martensite more favourably. The increase of mechanical characteristics in maraging steels revealed by the researchers is due to the reverted austenite, which forms thin compliant interlayers with both the prior austenite grain boundary and the martensite lath boundaries. Traditionally, intermetallic precipitates in maraging steels have been thought to contribute to strength enhancement during aging [87, 88]. Subsequent research has shown that the reversed austenite formed in some maraging steels exhibits excellent mechanical properties [85, 89]. It is proposed that [90]:

1. Austenite has the ability to act as a sink for impurities like N and P, which reduces embrittlement when heat treatment is applied.
2. In order to increase toughness and lower the Ductile-to-Brittle transition temperature, the ductile phase behaves as a compliance layer or mechanical buffer that prevents crack propagation along the {100}-planes of the martensite laths (DBTT).
3. Deformation may result in the transformation of austenite to martensite, which also increases toughness.

*5.1 Formation mechanism of reverted austenite*

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As a result of the dissolving of cementite or other precipitates, granular reverted austenite is formed, and this diffusion-controlled process results in the enrichment of nearby areas with austenite stabilizing elements [86, 91]. According to Sinha et al. [92], Ni diffusion at dislocations or other defects in the parent matrix is the primary cause of the enrichment of austenite stabilising elements. This finding is supported by Schnitzer et al. [86], F. Qian, and W. M. Rainforth [93], who showed that the activation energy needed to form reverted austenite (234–250 kJ/mol) is within the range of the activation energy for Ni diffusion in pure Fe (246 kJ/mol) [94]. However, the formation mechanism of lath like reverted austenite is still arguable [95]. The growth of lath-like austenite has been proven to be a function of diffusion, according to Plichata and Aaronson [96], whereas other researchers have hypothesised that the formation of lath-like reversed austenite is a shear-dominated process with diffusion as a supporting factor [95, 97]. Shear deformation-oriented phase transformation mimics bainitic transformation, also known as displacive phase transformation [98]. In some maraging steels, austenite reversion has also been reported to be accompanied by shear motion [95, 99]. Lee et al. [100] revealed the mechanism of austenite reversion in Fe-3Si-13Cr-7Ni (wt.%) martensitic stainless steel and medium Mn steels. His research highlighted the effect of heating rate on the mechanism of austenite formation. Below critical heating rates, reverted austenite of globular shape (diameter 200–250 nm), with a low dislocation density and a Mn enrichment, has been formed, with diffusion predominating in the formation process. Above the critical heating rates, the diffusionless shear mechanism dominates the formation process, resulting in lath-shaped (200-300 nm width and 400-700 nm length) austenite with a high dislocation density. Because of the low manganese content and particle size, the resulting austenite has poor mechanical stability. Unfortunately, shear driven mechanisms for bainitic and $\alpha' \rightarrow \gamma$ transformations [101] are not entirely acknowledged. For a deeper understanding, more research on the mechanism of $\alpha' \rightarrow \gamma$ transformation is required.

In retained austenite steels, austenite reversion occurs between the nucleation of fresh reverted austenite grains and the growth of already-existing retained austenite grains. The literature has revealed two mechanisms connected to the $\alpha' \rightarrow \gamma$ transformation: abnormal $\alpha' \rightarrow \gamma$ transformation and normal $\alpha' \rightarrow \gamma$ transformation [102]. The abnormal $\alpha' \rightarrow \gamma$ transformation implies that coalescence grain growth is caused by the reverted austenite nucleate at the prior austenite grain with a similar orientation. The process of reconstructing prior austenite grains (Figure 4) is known as the austenite grain memory effect (also known as austenite recrystallization) in contrast to the normal $\alpha' \rightarrow \gamma$ transformation, which prevents the growth of coalescence grains since the new grains have a different orientation.
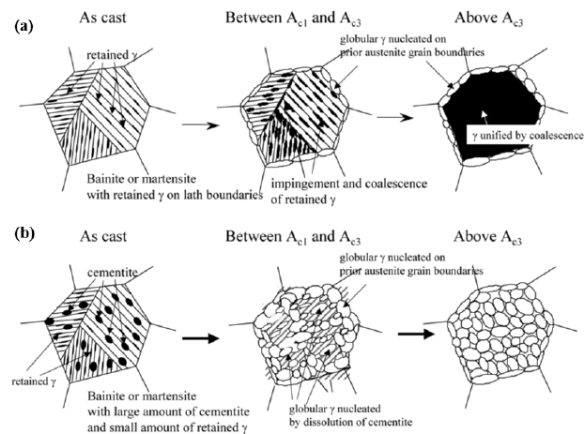


Figure 4 Schematic diagram of (a) abnormal $\alpha' \rightarrow \gamma$ transformation and (b) normal $\alpha' \rightarrow \gamma$ transformation [102]

## 6. MICROSTRUCTURE EVOLUTION OF Mn BASED MARAGING STEELS

Several scientists studied the microstructure on the basis of compositions of alloying elements and heat treatment processes. F.Qian et al. [103] the influences on mechanical properties and the microstructural evolution in a group of Mn-based maraging steels (7–12 wt% Mn) aged at 460–500 °C for varied times up to 10,080 min are thoroughly investigated It has been proven that by lowering the temperatures of $A_{e3}$ and $M_s$, the addition of Mn refines both the earlier austenite grains and martensite packets. The increased ductility and strength of the 12% Mn alloy in the SHT condition can be attributed in part to this grain refining as shown in figure 5. The EBSD observation of Raabe et al. [22] that a small amount of austenite was retained after quenching in a 12 weight percent Mn alloy is in conflict with this result. Three other explanations for

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this variance exist in addition to the EBSD resolution restriction. (i) Raabe might have employed a slower cooling rate, allowing some austenite to retain as it was and not transform into α'-martensite. (ii) Experiment and model simulation both showed that the martensite-start temperature (Ms) is dependent on the austenite grain size that existed before [38, 109, 110]. According to reports, one of the causes for the existence of retained austenite is the ultrafine prior austenite grain. Astenite is more stable during cooling because of the small grain size's suppression of the martensitic transition [38, 110]. (iii) The temperature of Ms rose as a result of the addition of Al in the current investigation.
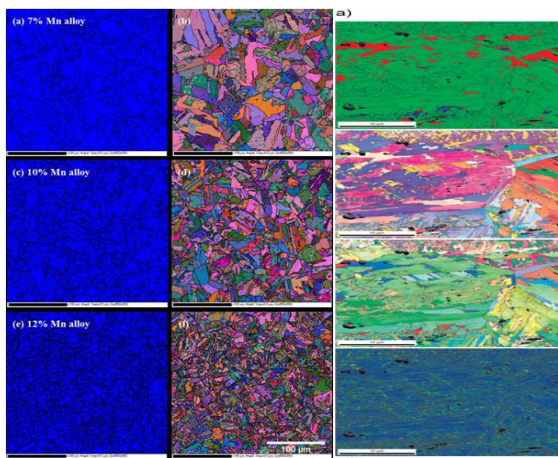


Figure 5 Microstructure of the three alloys in the SHT state by EBSD analyses. Left column: phase maps showing martensite phase (blue) with grain boundaries (black lines, >15°). Right column: inverse pole maps of the same area, showing the orientation distribution of the martensite.(Left) [103], High resolution EBSD maps taken on the 12 wt% Mn sample in the as-quenched plus aged state (Right) [22]

H. Luo et al. [31] studied the intercritical annealing of 5 wt% Mn steel at 650 C for various times up to 144 hours results in microstructures made up of globular γ/α grains and acicular γ/α grains, with the latter being predominately present after the long annealing process. Longer annealing times at 650 C produce thicker γ laths with an enrichment of Mn and a larger volume fraction of austenite, which greatly enhances elongation and reduces yield stress while having no impact on ultimate tensile stress. The impact of copper addition on steel with the composition Fe-12%Mn-0.7%C-1.0%Al twinning-induced plasticity is investigated [47], the strain rate for the FeMn-CAl and FeMnCAl-2Cu TWIP steels was 20%. In contrast to the abundance of mechanical twins that formed in the Cu-free FeMnCAl sample, just a few mechanical twins were seen in the FeMnCAl- 2Cu sample. Prior to the tensile tests, no mechanical twins were seen, indicating that the Cu content is clearly important for the twin formation kinetics.

Jingwei Zhao et al. [50] researched the effects of adding tungsten (W) to four different types of steel using specimens made by changing the W additions (0, 0.1, 0.5, and 1 wt pct). It was observed that W addition had no discernible impact on the amount of precipitation. The precipitates in W-containing steels were all highly concentrated in W, and their W concentration increased as the W content rose. With an increase in W concentration, austenite precipitates and grains both shrank in size. When the W level was equal to or less than 0.5 pct, the addition of W encouraged the creation of allotriomorphic ferrite, which subsequently encouraged the development of acicular ferrite (AF) in microalloyed forging steels. For Fe-9Mn-0.05C (wt.%) steel, the effects of the initial microstructure of α′ martensite on microstructural evolution during intercritical annealing for Hot rolling (HR) and Cold rolling (CR) were examined by J. Han et al. [104]. Results found that the HR Fe-9Mn-0.05C (wt.%) specimen exhibited a completely α′ martensitic microstructure absence of precipitates. After intercritical annealing at 620 °C for 600 s, the HR specimen had a mixed microstructure of lath-shaped ferrite ($α_L$) and retained austenite ($γ_L$). The volume proportion of $γ_L$ was 30%. The CR Fe-9Mn-0.05C (wt.%) specimen exhibited a fully deformed α′ martensite microstructure free of precipitates. After intercritical annealing at 620 °C for 600 s, the CR specimen showed a mixed microstructure of globular-shaped ferrite ($α_G$) and retained austenite ($γ_G$). The percentage of $γ_G$ volume was 37%.

J. Han et al. [58] used composition of steel Fe-7.22Mn-0.093C-0.49Si-0.013Al-0.005P-0.007S in wt.%, cold-rolled (CR) and hot-rolled (HR) specimens were intercritically annealed in a tube furnace at 640 C for 30 min in a vacuum environment, and afterwards air-cooled to room temperature. Results were found that the hot-rolled and annealed (HRA) specimen exhibited a lath-shaped microstructure of retained austenite ($γ_R$) and ferrite (α), whereas the cold-rolled

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and annealed (CRA) specimen had a globular morphology. Although the variation in microstructural morphology had no effect on the H permeability, it had a considerable impact on the HE behaviour. 11 weight percent manganese alloys was examined by Yen et al. [30], the initial deformation-induced microstructure of austenite can be dominated by parallel ε martensite lamellae or mechanical twinning, which cannot effectively serve as nucleation sites for strain-induced martensite, and it is observed when the austenite is reduced to a fine scale of around 300 nm. As a result, stress-assisted α′ martensite nucleation occurs independently, enhancing transformation-induced plasticity in ultrafine-grained austenite. The microstructural development and texture generation of 17Mn-0.06C steel after hot rolling, cold rolling to 45% of thickness reduction, and annealing at 700 C for various durations were assessed by Pérez Escobar D et al. [83]. Due to the presence of copper and brass components in significant amounts on the samples, transformation texture analysis demonstrated that austenite reversion does not occur simultaneously with recrystallization and it was discovered that ε martensite forms during processing and changes instantly to α martensite, drastically altering the mechanical behaviour of the steel.

With a chemical composition of 0.3C-4.5Mn-1.5Si (wt.%), medium manganese steel was quenched and partitioned using two distinct partitioning temperatures in J. Hidalgo et al. [57] study of how microstructure development affects fracture modes. It was discovered that the microstructure at the partitioning temperature of 400 °C displays intergranular fracture at low plastic strain, immediately after Mn-rich regions where fresh martensite predominates. At previous austenite grain boundaries, elongated thin precipitates make fracture initiation and growth easier. Following partitioning at 500 °C, the redistribution of carbon causes the precipitation of carbides in the carbon-enriched austenite, the creation of pearlite, and the formation of spheroidal carbides at previously formed austenite grain boundaries. P. Gong and colleagues investigated a novel idea for altering the strain hardening rate of medium Mn steel with 8 wt% Mn by varying the intercritical annealing time after hot rolling [107]. It was discovered that the maximum strain hardening rate and the stability of the austenite phase exhibited linear and inverse square root relationships, respectively, with respect to the intercritical annealing duration. Continuous Mn enrichment with longer intercritical annealing times was implicated for the alteration in austenite stability. The most likely outcome of the martensitic transition during deformation was likewise discovered to be twinned martensite. In an experiment

by M.K. Chourasia et al. [108], steel with the following composition was isochronally aged at temperatures between 500 and 650 C at intervals of 50 C: 0.1C-17Mn-5.5Ni-1.8Al-2.7W-4.6Mo-2.3Cu-0.002B and concluded that the experimental steel features a multiphase microstructure, with retained austenite, α′ -martensite, and ε-martensite making up the matrix. With rising temperatures, α′ -martensite content decreases. An increase in ageing temperature favours austenite's athermal conversion to ε-martensite. Ni3Al and M2C (M ~Mo, W) precipitate as a result of ageing, however the precipitate structures seen in the aged samples seem complex and require further investigation.

## 7. ACTING STRENGTHENING MECHANISMS OF Mn BASED MARGAING STEELS

The research findings on medium Mn steels (often with 5-7 wt% Mn) regarding the tensile properties of Mn-based maraging steels are summarised in Table 1. As demonstrated in Table 1, excellent tensile properties have been found in different Mn-based steels subjected to different processing pathways. However, considering the steels' low carbon content (0.03 wt.%), the tensile strength of 9–12% Mn steels is still impressive when compared to steels with the same amount of carbon content in Table 1. It is proposed that the high UTS is due to the combined effect of precipitation strengthening of the Ni2TiAl intermetallic phase and reverted austenite work hardening. The ductility of the 10% and 12% Mn alloys, on the other hand, is only average. The comparatively lower austenite fraction compared to other alloys is one potential explanation. Another possibility is related to reverted austenite's high mechanical stability.

The strengthening mechanism for steels is complex, but it can be summarised as follows:

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$$\sigma_y = \sigma_B + \sigma_{SS} + \sigma_{GR} + \sigma_{PH} + \sigma_{dislocation} \qquad (1)$$

Where $\sigma_y$ is the total yield strength, $\sigma_B$ is the yield strength of base steel, $\sigma_{SS}$ is the solid-solution strengthening, $\sigma_{GR}$ is the grain refinement strengthening, $\sigma_{PH}$ is the precipitation strengthening and $\sigma_{dislocation}$ is the dislocation strengthening.

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Table 1 Summary of the tensile properties of Mn-based steels.

| Materials (wt.%) | Heat treatment | $V_\gamma$ (vol.%) | YS (MPa) | UTS (MPa) | TE (%) | Ref. |
|---|---|---|---|---|---|---|
| Fe-0.015C-10Mn-2Ni-1.2Mo-1Al-0.8Ti | 500 °C / 5760 min, WQ | 10.4 | 860 | 1062 | 17 | [103] |
| Fe-0.022C-11.8Mn-2Ni-1.2Mo-1Al-0.8Ti | 500 °C / 5760 min, WQ | 24.6 | 870 | 1120 | 18.4 | [103] |
| Fe-0.007C-8.86Mn-2.0Ni-1.07Mo-1.04Ti | 450 °C / 48 h | - | 984 | ~1000 | ~15 | [22][30] |
| Fe-0.007C-8.86Mn-2.0Ni-1.07Mo-1.04Ti | 450 °C / 48 h [a] | - | 978 | 1011 | 12.7 | [22][30] |
| Fe-0.01C-11.9Mn-2.06Ni-1.12Mo-1.09Ti | 450 °C / 48 h | - | 1142 | ~1300 | 21 | [22][30] |
| Fe-0.01C-11.9Mn-2.06Ni-1.12Mo-1.09Ti | 650 °C / 144 h, AC | 32.7 | ~500 | ~980 | ~45 | [31] |
| Fe-0.11C-5.7Mn | 560 °C / 1 h, AC | 11 | 955 | 981 | 12 | [28] |
| Fe-0.11C-5.7Mn | 520 °C / 4 h, AC | 10 | 934 | 956 | 19.5 | [28] |
| Fe-0.11C-5.7Mn | 600 °C / 1 h, AC | 23 | 874 | 914 | 18.5 | [28] |
| Fe-0.11C-5.7Mn | 640 °C / 1 h, AC | 30 | 798 | 1145 | 30.5 | [28] |
| Fe-0.11C-5.7Mn | 600 °C / 16 h, AC | 29 | 727 | 879 | 34 | [28] |
| Fe-0.11C-5.7Mn | 640 °C / 16 h, AC | 30 | 87 | 174 | 18.5 | [28] |
| Fe-0.04C-5Mn | 590 °C / 16 h [b] | 75.5 | 952 | 1049 | 39.5 | [17] |
| Fe-0.045C-5Mn-3Ni | 590 °C / 4 h [b] | 76.4 | 1104 | 1249 | 39.9 | [17] |
| Fe-0.2C-4.72Mn | 650 °C / 1 min, AC | 4.54 | ~830 | ~970 | ~20 | [31] |
| Fe-0.01C-7.09Mn | 575°C / 1 week, WQ | 26 | 766 | 800 | 33 | [27] |
| Fe-0.01C-7.09Mn | 625°C / 1 week, WQ | 40 | 954 | 503 | 22 | [27] |
| Fe-0.12C-4.6Mn-1.1Al-0.55Si | 720 °C / 2 min, 10°C/s [c] | 28 | 766 | 1204 | 12.9 | [20] |
| Fe-0.05C-9Mn | 620 °C / 600 s, FC [d] | 37 | 1110 | 1193 | 25 | [104] |

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| Fe-0.1C-5.25Mn-2.09Si | 675 °C / 1 h, FC | - | - | 1029 | 27 | [18] |
|---|---|---|---|---|---|---|
| Fe-0.01C-5Mn | 625 °C / 3 h, WQ | ~4 | ~500 | ~800 | ~10 | [105] |
| Fe-0.4C-5Mn | 675 °C / 3 h, FC | ~15 | ~350 | ~1550 | ~10 | [105] |
| Fe0.15C-7Mn-0.2Si | 600 °C / 4 h, AC [c] | 11.8 | 827 | 958 | 20 | [106] |
| Fe-8Mn-2.5Al-1.5Si-0.4C-0.02Nb-0.03V | 660 °C/ 10 min [c] | 30 | 772 | 1216 | 27 | [107] |
| 0.1C-17Mn-5.5Ni-1.8Al-2.7W-4.6Mo-2.3Cu-0.002B | 550 °C/ 1 h | - | 640 | 1408 | 25 | [108] |
| 0.1C-17Mn-5.5Ni-1.8Al-2.7W-4.6Mo-2.3Cu-0.002B | 600 °C/ 1 h | - | 624 | 1326 | 25.54 | [108] |

Where, AC: air cooling; WQ: water quenching; FC: furnace cooling.

[a] Cold rolled before annealing

[b] four-step thermal cycling 2BT treatment, detail see [17]

[c] single-step heat treatment of intercritical annealing

[d] Cold rolled after hot rolling

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*7.1 The effect of reverted austenite on mechanical behavior*

It has been widely investigated how the presence of reverted austenite improves the mechanical properties of maraging steels. In a work on 18%Ni maraging steels, Vylezhnev et al. [111] found that ductility may be improved without compromising tensile strength. Similar improvements were noted by Nakagawa et al. [24] and Raabe et al. [22] in maraging steels containing Cu and 9 weight percent Mn, respectively. The improvements are attributable to an increased fraction of reversed austenite following ageing treatment. The ultimate tensile strength and total elongation were reported to have increased by 25% (from 810 MPa to 1 GPa) and 150% (from 6% to 15%), respectively, by Raabe et al. [22]. Strength and ductility can be combined in a way that is advantageous if there is an ideal volume fraction of the reverted austenite phase in the parent matrix. Chemical composition, aging temperature, and treatment duration can all affect this proportion [22, 31]. The ultimate set of attributes of maraging steels are mostly determined by stability in addition to the volume fraction of austenite. In order to achieve deformation-induced martensitic transformation in TRIP and quenching and partitioning (Q&P) steels, austenite must maintain critical temperature stability; otherwise, it may lose ductility by changing at lower strain values or resist the martensitic transformation, which would prevent the TRIP effect. As a result, austenite stability needs to be adjusted in accordance with specifications. Chemical composition, temperature, size of retained austenite, shape, adjacent microstructure, and crystallographic orientation relationship are all factors that might affect austenite stability [112–115].

## 8. CONCLUSION

This study contributes a substantial body of new knowledge on a number of topics related to Mn-based maraging steels, such as a brief history of their development, the function of alloying elements in maraging steel, Mn embrittlement in Fe-Ni-Mn and Fe-Mn alloys, reverted austenite, and the effects of process and composition on microstructure and mechanical characteristics. Due to their high UTS and elongation, Co and Ni were initially used in maraging steels. However, because they are expensive, many researchers have concentrated on developing alternatives. The similarity of the equilibrium phase diagrams of the Fe-Ni and Fe-Mn systems led to the partial substitution of Mn for Ni with Fe in the creation of lean maraging steels.

Based on the discussion in this study, there are some suggestions:

- With the use of atom probe microscopy, precipitates at the atomic scale of magnitude should be examined for detailed information on composition. Combining HRTM and atom probe microscopy (APM) should be used to examine the structure of the various forms of precipitates.
- Further research on deep drawability should be done in order to validate the adaptability of the steels for various structural applications, including those in the automobile sectors.

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