

# TRIP Steels: Factors influencing their formation, Mechanical Properties and Microstructure- A Review

Victor Alcántara Alza<sup>1</sup>, Víctor Pelaez Chavez<sup>2</sup>

<sup>1</sup>(Mechanics and Energy, Faculty of Engineering/ National University of Trujillo, Perú)

<sup>2</sup>(Mechanics and Energy, / Faculty of Engineering / National University of Trujillo, Perú)

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## Abstract:

TRIP steels (transformation induced by plastic deformation) are new generation steels within the advanced AHSS steels. They are characterized by having a good combination of mechanical properties, such as high strength, high ductility and a high rate of strain hardening apart from a moderate yield point. Its microstructure is composed of ferrite, bainite and retained austenite. The first is the soft matrix, the second provides strength and toughness, and the third is austenite, which during deformation become into martensite will give it great mechanical strength. This combination of properties makes it very suitable for use in applications in the automotive and aeronautical industries. The purpose of this article is to investigate the factors that most influence their formation, mechanical properties and microstructure of TRIP steels, based on the specialized and updated literature that exists to date. Depth review is made on the methods of producing these steels based on the heat treatment routes used, taking into account the kinetic and thermodynamic aspects of their formation. It be report on the theoretical models that evaluate the parameters of phases evolution at each stage of their formation. Not all steels can be subjected to the TRIP effect; ; Certain requirements must be met, ranging from the elements that make up the alloy to the thermomechanical parameters of the process. Of all the factors that influence the TRIP effect, it is worth highlighting the stabilization of the retained austenite, as well as the rate and time of deformation in shaping. Finally, it is reported that despite existing theoretical models, the TRIP effect is still not fully understood.

**Key Word:** Aadvanced steels; Trip steels; Retained Austenite; Intercritical Annealing; Austempering .

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## I. Introduction

One of the most important achievements of recent decades in the search for new and better materials for the automotive sector, was the development of steels with a TRIP effect (Transformation Induced by Plasticity), which gave rise to a new type of steel called TRIP steels, which belong to the category of advanced high-strength steels (AHSS), and are widely used in the automotive industry [1–10]. These steels exhibit excellent combinations of high mechanical strength and ductility after cold forming, making them particularly suitable for applications in the manufacture of sheets and plates in automotive industry [11]. This excellent combination of properties (high strength and ductility) is achieved because these steels have a multiphase microstructure that contains ferrite, bainite and retained austenite, which provides them with high performance during cold forming operations. [12, 13]. In these forming processes that use TRIP steels, the retained austenite is transformed into martensite under the action of the applied stresses and strains. This mechanically induced transformation of the retained austenite into martensite is the so-called TRIP effect. Due to the TRIP effect, it is possible to produce, by forming, a steel with a combination of high mechanical strength and high ductility [14]. This proper balance between strength and ductility comes from the strain-induced transformation of metastable retained austenite to martensite during plastic deformation [15]. With These new steels with a TRIP effect is possible to reduce the thickness of the plates used in car bodies maintaining their high properties, which brings about a reduction the weight of the bodies and improving the efficiency of fuel consumption and reducing the amount of emissions into the environment. To meet these requirements, this new category of advanced AHSS steels, called TRIP steels have been developed [16].

Mechanical properties of the recently developed TRIP steels range from maximum tensile strength (UTS) of 1000-1500 MPa and elongation at failure the order: 20-30% [17] overcoming other types advanced AHSS steels. On the other hand, specialized literature reports that retained austenite, only with optimal carbon content (>0.5-0.6% and <1.8%) can provide the TRIP effect and increase elongation [18]. In low alloy steels, the TRIP effect can be achieved with small amounts of retained austenite transformed into martensite during

deformation, as long as the grain size is small and the phase distribution is homogeneous to produce a strong hardening and therefore excellent mechanical strength and elongation [19]. The particularity of TRIP steels is their elongation and mechanical resistance increase as the retained austenite transforms into martensite during plastic deformation [20].

TRIP steels generally arise from heat treatment programs, which consist of an isothermal intercritical annealing, with a subsequent isothermal treatment in the bainitic region. The purpose of the treatment is to stabilize the austenite by increasing its carbon concentration. The carbon enrichment is due to phase separation during intercritical annealing and suppression of carbide formation during bainitic transformation. The microstructure after heat treatment consists of ferrite, bainite, retained austenite, and sometimes martensite [21]. The optimization of TRIP steel properties is based on maximizing the amount of retained austenite with adequate stability [22].

During intercritical annealing and subsequent transformation to the bainitic phase, the remaining austenite is enriched with carbon due to carbon partitioning. In this way, the austenite is stabilized against martensitic transformation during subsequent cooling to room temperature. Austenite stabilization must be well balanced to prevent martensite formation during cooling and allow continued martensite formation during deformation. The most important aspect in the thermal treatment process of TRIP steels is, above all, the production of an adequate and relatively high carbon content in the austenite [16]. An appropriate composition of these alloys consists of a content that falls within the range (0.1–0.2% by weight) and small additions of silicon, aluminum and manganese (each in the range of 0–2% by weight) [23]. Silicon strongly slows down or completely prevents cementite precipitation which normally occurs during bainite formation, and leads to a high carbon content in the solid solution. Aluminum which has a similar effect but weaker with respect to inhibition of cementite formation, strongly increases the bainite formation Kinetics [24]. In some cases phosphorus is added, which also inhibits the formation of cementite, due to its strong solid solution hardening effect on ferrite. Manganese is added to prevent or slow pearlite formation allowing a lower critical cooling rate and to stabilize austenite at the bainite transformation temperature.

The chemical composition, especially the carbon concentration, the size and the morphology of the austenite determine its stability against plastic deformation. For example, when the carbon concentration is less than 0.06%, the austenite transforms to martensite in the first stages of deformation, without contributing to the increase in ductility, on the other hand, when the carbon concentration in the austenite is greater than 1.8 % by weight no transformation occurs and therefore there is also no contribution to elongation increase [25]. It has also been found that austenite units larger than 1  $\mu\text{m}$  transform to martensite at low strains, while submicron-sized austenite units are more stable. Finally, it has been found that block austenite is less stable than austenite particles found between bainitic ferrite sheets [26, 27]. Although retained austenite is the phase with the least presence in the TRIP microstructure, it has an important effect on the mechanical properties of TRIP steels. Si, Al, P additions alone or in combination suppress carbide formation during isothermal bainite transformation. This allows the remaining austenite to become enriched in C, resulting in its stabilization at room temperature. Retained austenite transforms to martensite during deformation of TRIP steels, resulting in a higher work hardening rate at higher strains. This causes the superior balance between strength and ductility of TRIP steels. [11].

During plastic deformation, austenite transforms into martensite, which improves remarkably the material properties and causes hardening. Stress spreads evenly across the entire workpiece, preventing local build-up of stress and strains [28].

There is no doubt that Mechanical Properties of TRIP steels are due to the transformation of the retained austenite into martensite during deformation and therefore, appear to be dominated by the volume fraction and carbon content of the retained austenite. [29, 30, 31, 32, 33, 34]; However, the optimization of all parameters is necessary to obtain the appropriate technological and mechanical properties. The parameters mainly include the chemical composition, the primary microstructure and the heat treatment procedure, which ensure adequate volume fractions of the individual phases and stabilize the austenite for the TRIP effect. The heat treatment is carried out under intercritical conditions. of temperatures between A1 and A3. In certain cases, a resulting microstructure has been reported that typically contains 50% ferrite, 30-35% bainite, and 15-20% retained austenite. During processing, retained austenite provides high plasticity of the material. After cold deformation or final shaping, austenite transforms into martensite and, together with bainite, forms a hardened microstructural component. [35].

The main objective of any TRIP steel process route is to obtain this required minimum amount of retained austenite in the final microstructure at room temperature. This is possible only if the highest temperature at which austenite transforms to martensite on cooling, i.e. the initial temperature of martensite or  $M_s$ , is somehow lowered to a value below room temperature. On the other hand, if the significant influence of carbon on  $M_s$  reduction and the fact that carbon is the cheapest source of austenite stabilization, these are aspects that must be taken into account.[36].

Process routes are generally designed to ensure carbon enrichment and consequent retention of austenite at room temperature. To date, a great deal of effort has been devoted to optimizing thermomechanical processing parameters to improve the mechanical properties of TRIP steels [37-41]

The objective of this research work is to investigate the contributions provided by the specialized literature regarding the advance and development of the main factors that influence the formation of TRIP steels, as well as their mechanisms that determine each stage of their formation to achieve a total understanding of the phenomenon in its entirety and in this way theoretical models closer to reality can be proposed.

## **II. Methods to Produce TRIP Steels**

There are various methods to produce TRIP steels, whether cold or hot rolled, but the most common method to produce a TRIP steel is to start with a hot rolled steel, followed by cold rolling in order to deform the microstructure and diffuse the potential energy needed to achieve recrystallization efficiently. The steel is reheated to the intercritical temperature where the steel recrystallizes by growing small grains of austenite and ferrite. Depending on the amount of initial strain and holding temperature, this recrystallization into ferrite and austenite can be very rapid. Long waiting times in this period are avoided to prevent the growth of harmful grains. Methods are specified by their heat treatment Routes.

### **2.1. Thermal Treatment Routes.**

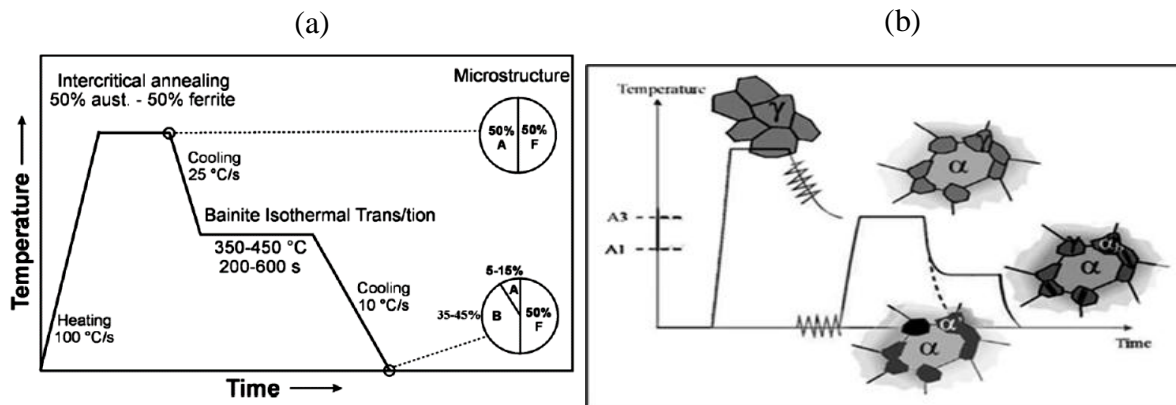
TRIP steel obtains the triphasic microstructure following a specific heat treatment procedure presented in Figure 1a. The steel sample is first heated to a temperature of 700-900°C where intercritical annealing (**IA**) is carried out. The intercritical annealing temperature and time (nearly 3 minutes) are chosen such that a ~50% ferrite and 50% austenite microstructure is formed. Cooling after intercritical annealing is performed at an intermediate temperature above the martensite starting temperature  $M_s$ , which allows bainite transformation to occur during isothermal holding to produce bainite (**IBT**). The isothermal transformation of bainite takes place at temperatures of ~350-450°C and times of 200-600 seconds, which leads to the transformation of a part of austenite to bainite. The amount of austenite transformed into bainite depends on the temperature and isothermal transformation time of the bainite. Therefore, after the isothermal transformation of bainite, the steel obtains a microstructure of 50% ferrite, 35-45% bainite and 15-5% austenite. In this step, the remaining austenite is further enriched with carbon, which shifts the starting temperature of the martensite below room temperature. Finally, the steel sample is quenched at room temperature without the austenite transforming into martensite and the desirable microstructure of 50% ferrite, 35-45% bainite and 15-5% austenite is retained until it is formed.

Figure 1b) shows a representative graphic scheme of the way in which the phases evolve until their final structure. Figure 2 shows a more detailed scheme of the formation of the phases from the austenitic grain structure  $\gamma$ . This figure clearly shows the evolution of the microstructure, which presents intermediate phases, as is the case of the proeutectoid ferrite  $\alpha_T$ , which will later transform into the second stage of the IA. Figure 3 shows the image of a typical microstructure of a TRIP steel, both optical and electronically. [42]. The phases are clearly observed in the SEM image of figure 3b).

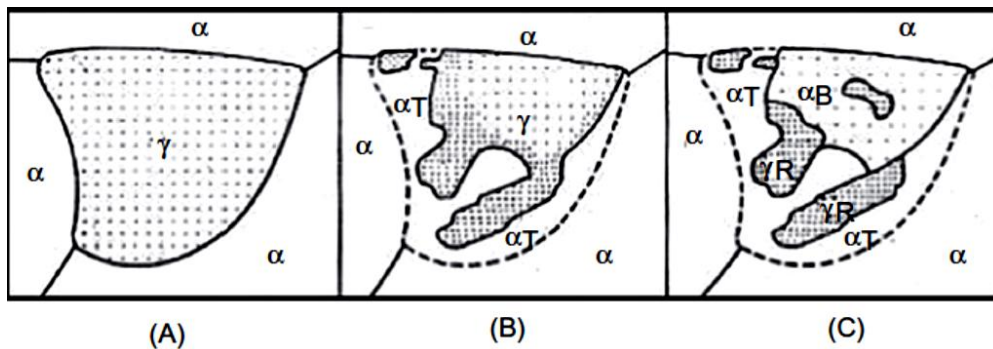
TRIP steels can also be produced in the hot-rolled state. After finishing rolling at the intercritical temperature of around 780-860 °C [44, 45] or even higher than  $A_{c3}$  in some cases [46]. In this process, the winding step is equivalent to the IBT step in the previous cold rolling treatment route. This is followed by air cooling which results in the desired microstructure. TRIP-assisted hot-rolled steels up to 780 Mpa have been commercially produced. After finishing rolling at intercritical temperature around 780-860°C. The winding step is equivalent to the IBT step in the previous cold rolling treatment route. This is followed by air cooling which results in the desired microstructure. TRIP-assisted hot-rolled steels. These hot-rolled TRIP-assisted steels, of up to 780 Mpa have been produced by this procedure [47]. However, TRIP steels with a mechanical strength ~ 1 GPa are currently found, generally manufactured following the cold deformation route. The choice of one process or another is subject to the design and service conditions where the respective steel is going to be used.

In addition to the routes mentioned above, another process route called the "cool and partition" route has been developed to produce a higher fraction of retained austenite by partitioning carbon from supersaturated martensite. This process involves an initial full/partial austenitization step, followed by "tempering" and then "partitioning" treatments.

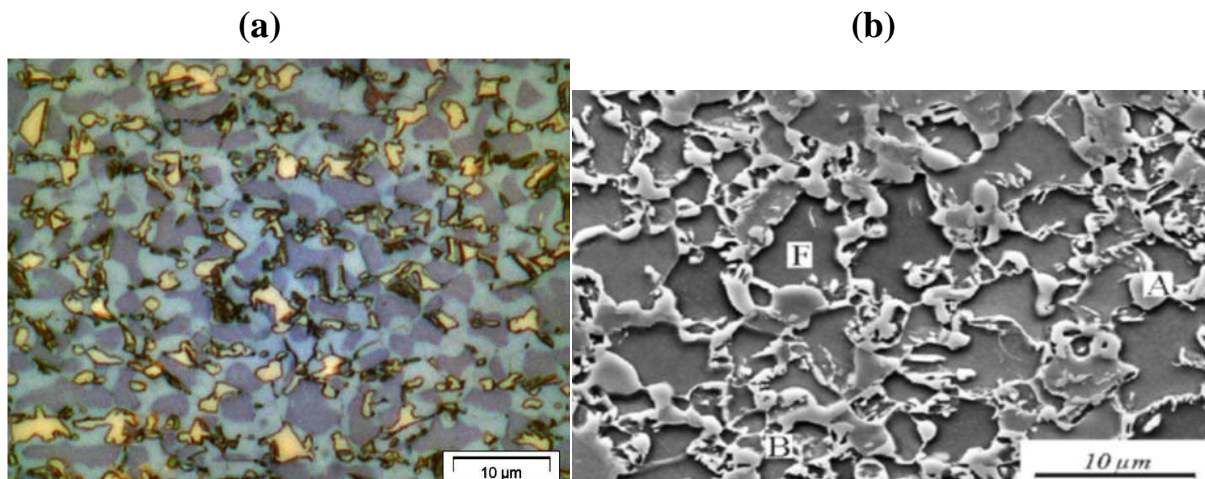
Though the conventional TRIP aided steels with polygonal ferrite as matrix as well as the TRIP aided steel with bainitic ferrite matrix can be produced by conventional processing routes like HR and cold rolling as described above, the commercial viability of the other varieties of TRIP aided steels is yet to be established.



**Figure 1:** a) Schematic representation of the heat treatment performed for TRIP steels (A: austenite, B: bainite, F: ferrite) [42]; b) Schematic representation of the thermo-mechanical treatments applied to TRIP steels that complements scheme 1a), with the two treatment stages showing their phases [43]



**Figure 2.** Microstructural changes during two-step heat treatment: (A) IA stage, (B) Cooling from IA temperature to IBT temperature, (C) IBT stage. ( $\alpha$ : ferrite;  $\alpha_B$ : bainite;  $\gamma$ : austenite;  $\alpha_T$ : pro-eutectoid ferrite formed on cooling from temperature IA to temperature IBT;  $\gamma_R$ : retained austenite) [48].



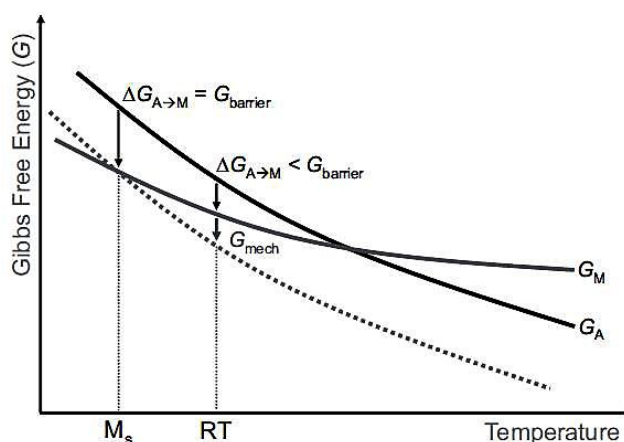
**Figure 3.** a) Optical Level Microphotograph showing the triphasic microstructure of a TRIP steel [42] b) SEM microphotography of typical TRIP steel, where the phases are clearly observed. Ferrite (F), bainite (B) and austenite (A). [43]

### III. Martensitic Transformation of TRIP Steels

#### 3.1. Thermodynamic aspects

The martensite transformation is driven by energy minimization between the phases, like any other phase transformation. At high temperatures, austenite  $\gamma$ , which is a face-centered structure (FCC), is stable because it has the lowest free energy; while at lower temperatures, body centered cubic (BCC) ferrite has lower

free energy and is thus the stable phase. The free energy difference between the FCC and BCC structures is the driving force for the martensitic transformation  $\gamma \rightarrow \alpha$ . Figure 4 schematically shows the free energy of the austenite and martensite phases as a function of temperature.



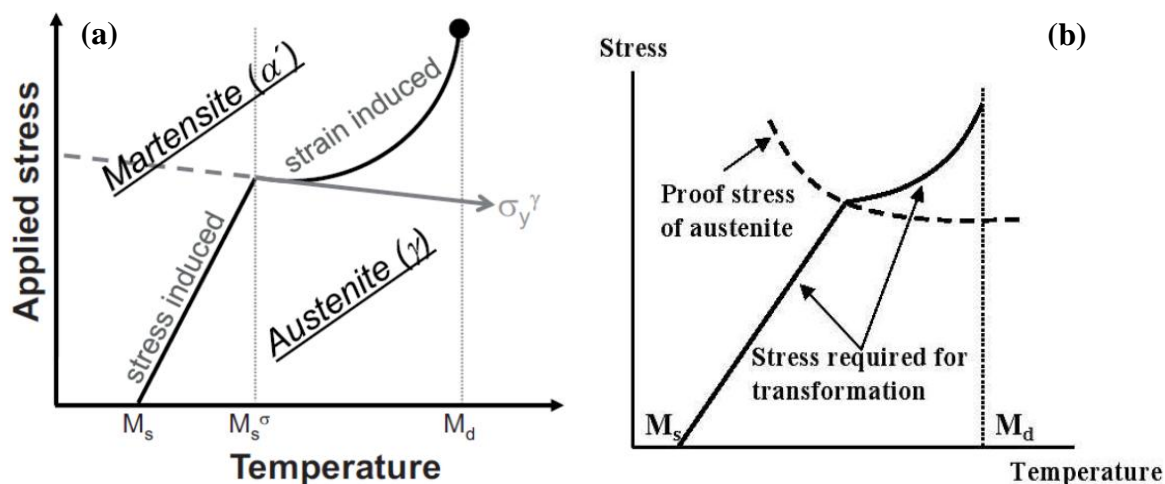
**Figure 4.** Schematic representation of free energy of the austenite and martensite phases as a function of temperature

The transformation of austenite into martensite begins at the temperature  $M_s$ , at which the free energy difference between austenite and martensite  $\Delta G_{A \rightarrow M}$  is sufficient to overcome the transformation energy barrier. The temperature  $M_s$  depends on the chemical composition of the steel. The  $M_s$  temperature for the nominal composition of a low alloy TRIP steel is about 350 °C. On the other hand, the  $M_s$  temperature for austenite in TRIP steels is well below room temperature since the retained austenite has been enriched in carbon during the bainitic support process.

At temperatures above  $M_s$ , the martensitic transformation can be triggered by the application of external mechanical loads. Here, the mechanical energy term ( $G_{mech}$ ) is added to the free energy difference ( $\Delta G_{A \rightarrow M}$ ) to overcome the transformation barrier ( $\Delta G_{barrier}$ ) as shown in Figure 4. It should be noted that the martensitic transformation in carbon steels is irreversible in contrast to shape memory alloys where the strain-assisted transformations are crystallographically reversible. The reverse reaction of martensite to austenite in carbon steels can only be done by reheating; however, this reverse reaction is intervened by tempering.

Strain-induced martensitic transformation can be interpreted in two ways: 1) stress-induced transformations 2) strain-induced transformation. The interrelationships between the applied stress, the plastic deformation, the test temperature and the transformation of the martensite, could give rise to different nucleation mechanisms. These interrelationships were extensively studied by Bolling and Richman [49], and set a temperature  $M_s^{\sigma}$  (which is normally above  $M_s$ ) to distinguish between Stress Induced Transformation and Strain Induced Transformation.

Martensitic transformation in austenitic stainless steels involves shear band intersections,  $\epsilon$ -martensite (HCP phase that grows by overlapping stacking faults), and mechanical twinning. When the amount of  $\alpha'$ -martensite increases with the degree of strain, the amount of  $\epsilon$ -martensite goes through a maximum and then falls off until it disappears. This suggests the  $\gamma \rightarrow \epsilon \rightarrow \alpha'$  transformation sequence. Austenite ( $\gamma$ ),  $\alpha'$ -martensite,  $\epsilon$ -martensite have FCC, BCC, and HCP structures, respectively



**Figure 5.** a) Schematic representation of the interrelationship between stress, temperature, and martensite nucleation.

The critical stress required for stress- or strain-induced martensitic transformation is shown as a function of temperature. b) Stress requirement for martensitic transformation above  $M_s$  increases at higher temperatures; formation of martensite is not possible beyond  $M_d$ . [50]

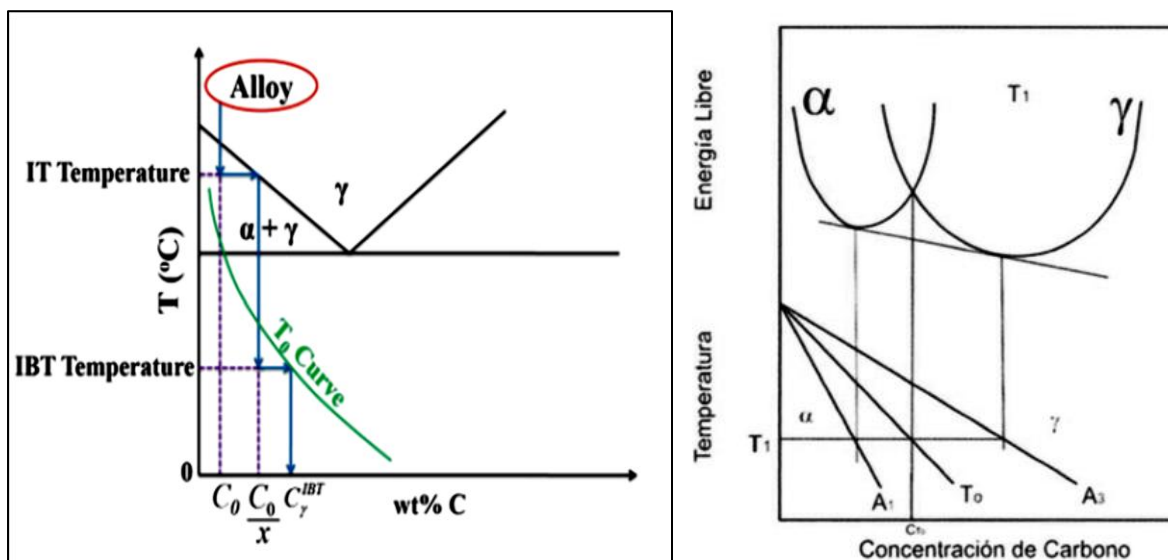
Figure 5 compares the critical stress required for induced transformation as a function of temperature. The yield strength of austenite and the temperature  $M_s$  are also indicated. Below the temperature  $M_s^g$ , the martensitic transformation occurs below the yield stress ( $\sigma_y$ ) of the austenite, and causes plastic flow of the sample. The transformation is thought to be predominantly initiated by nucleation at the same pre-existing sites responsible for spontaneous transformation on cooling, but assisted by elastic stresses. [51], therefore, this transformation is defined as stress-induced transformation. Above  $M_s^g$ , the transformation starts after the applied stress reaches or exceeds the elastic limit of austenite [52]. New nucleation sites can be introduced by plastic deformation and this mechanism is defined as strain-induced nucleation since plastic deformation of austenite precedes transformation. The plastic deformation of the austenite could continue until fracture as represented by the black dot ( $\bullet$ ) in figure 5a). This point also defines the maximum temperature  $M_d$ , at which the chemical driving force is so small that it is virtually impossible to nucleate martensite, even by external mechanical loading.

### 3.2. Phases Calculation

The conventional two-step heat treatment to produce TRIP steels has already been explained above. But the question arises how to determine the amounts of austenite-martensite phases in this process. For this purpose, Figure 6(a) is schematically presented, where it can be seen that the  $T_0$  curve has been superimposed on the Fe-C diagram. The  $T_0$  curve is defined as the locus of the points where the free energies of austenite and ferrite of the same composition are equal and represents the upper limit above which the bainite transformation is thermodynamically impossible [53, 54, 55].

The way in which the  $T_0$  curve is obtained can be shown under thermodynamic aspects in Figure 6(b). This graph represents the free energy of the austenite and martensite phases as a function of temperature, showing that at  $T_0$  both phases have the same free energy ( $G_\gamma = G_{\alpha'}$ ). Therefore, in the temperature range below  $T_0$ , the formation of martensite is thermodynamically possible. However, in the temperature range above  $M_s$  and below  $T_0$ , martensite cannot be obtained because the free energy difference has not yet been determined for reached the critical value for the transformation start.

In modern times, the determination and quantification of the phases are carried out by the techniques of X-ray diffraction (XRD), neutron diffraction and electron backscatter diffraction (EBSD). The quantitative analysis of the phases is determined according to the main diffraction peaks of the  $\gamma$ ,  $\alpha'$  and  $\epsilon$  phases.



**Figure 6.** a) Schematic diagram showing the  $T_0$  curve superimposed on Fe–C phase diagram, b) thermodynamic plot shows Origin of the  $T_0$  curve, explained by a phase diagram considering the free energy of ferrite and austenite

The phase diagram and  $T_0$  curve for a given composition can be calculated using analytical methods or thermodynamics-based software such as Thermo-Calc. This graph allows to obtain the termination point of the decomposition of austenite into ferrite and bainite at the respective treatment temperatures, that is, IT and IBT respectively. The fraction of ferrite, bainite and austenite remaining after the IBT treatment can be calculated by mass balance in the two transformation stages. The carbon enrichment of the austenite after the IBT treatment, given by  $C_0^{IBT}$  determines the amount of martensite that can form on cooling due to room temperature ( $T_r$ ) and the amount of austenite retained in the final microstructure.

### 3.3. Stability of Transformation Mechanics

The martensitic transformation of austenite during deformation can occur due to stress or strain. The rapid transformation at first is due to the nucleation site introduced by the deformation. However, excessive tension can slow the transformation and eventually stop it. This last phenomenon is known as mechanical stabilization of the transformation, a characteristic feature of displacement transformations. Rather, the effect of stress is purely thermodynamic [56]. Strain-induced martensitic transformation is retarded under single-strain conditions relative to uniaxial loading [57]. This is due to the lower stress and strain applied under plane strain conditions. In contrast, under biaxial stretching, the transformation is promoted relative to the uniaxial test, although the results vary with the orientation of the sample relative to the rolling direction. [58]. This phenomenon may be related to the texture of the austenite grains [59].

### 3.4. Theoretical models of TRIP effect

In the specialized literature, there are various micromechanical models to explain the TRIP effect; for which the authors have considered various hypotheses or premises that support the phenomenon; as in the whole process the temperature does not change or other considerations. Some of these models have been experimentally validated, others are going that way. In this article, only the most outstanding models will be referred to, since the presentation of their complete development is beyond the objectives of this study.

Most models are based on the Greenwood and Johnson model; such as the Leblond models [60, 61], Taleb [62, 63] y Fischer [64–66] which are the most relevant. Taleb set out to investigate some discrepancies from a reevaluation of the micromechanical model as originally used by Leblond et al. [61]. A more complete formulation of the phenomenon was proposed taking into account the elasticity in both phases, which resulted in a better agreement with the experimental results. Leblond's model [60, 61] took into account the interaction between classical plasticity and the TRIP effect, but has not yet been experimentally validated [63]. Fischer et al. [65, 66] investigated the orientation effect of shape memory alloys and quantified the orientation effect based on the Greenwood-Johnson model.

Phase transformation kinetics is the key model to describe the transformation of martensite, which was modeled by Olson and Cohe [67], Stringfellow et al. [68], Sumigoto et al. [69–71], Ángel et al. [72], Ludwigsøn y Berger [73]. Ángel et al. y Ludwigsøn y Berger, studied the behaviors of plasticity induced by the

transformation of stainless steel and the kinetics of phase transformation of martensite; Olson and Cohen also modeled the phase transformation kinetics of martensite and Stringfellow et al. They developed a model based on the intersection of the cut bands. Etc.

Currently, new numerical analytical models and others based on computational tools are still being developed to describe the TRIP effect, but as it was said in other passages that this phenomenon is not yet fully understood, so that a single model that cover all cases in which the TRIP effect occurs.

Under the criterion already established and corroborated that in steels containing austenite in the MS-Md temperature range, the formation of martensite during plastic deformation helps maintain strain hardening. Olson and Cohen [67] proposed an empirical-type model for the strain-induced transformation that is summarized in equation (1)

$$F^{\alpha'} = 1 - \exp[-\beta\{1 - \alpha\varepsilon\}^n] \quad (1)$$

Where  $F^{\alpha'}$  is the volumetric fraction of martensite obtained at one strain. The model contains two temperature-dependent parameters,  $\alpha$  and  $\beta$ . The first depends on the stacking failure energy of the steel, which in turn is affected by temperature. defines the probability that a cut band intersection will create a nucleation site, and thus can be thought of as depending on the chemical driving force and thus on temperature. However, the exponent  $n$  in the model comes from the fit to experimental data.

In summary, strain-induced transformation of retained austenite has been considered as the main factor influencing mechanical performance. This is despite the fact that these steels possess multiphase microstructures, of which 10-30% it's just retained austenite. A moderately stable austenite seems to be the key to optimizing ductility. Therefore, not only the austenite fraction, but also its carbon content, seems to be an outstanding microstructural feature that controls the properties of TRIP steels.

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### 3.5. Stability of Retained Austenite

Metastable retained austenite is a crucial microstructural component that improves the mechanical properties of advanced steels through the effect of transformation-induced plasticity (TRIP) [74]. Retained austenite is the essential component of high-strength multiphase steels, increasing strength without compromising elongation. The effectiveness of the TRIP effect is related to the quantity and stability of the AR [75]. The resistance of RA to martensitic transformation during deformation, which is known as its mechanical stability, and is a very important and determining factor of TRIP steels for advanced automotive applications [76]. Gradual transformation of metastable retained austenite to strain-induced martensite during deformation provides an excellent balance between strength and ductility [77].

The stability of the retained austenite depends on numerous internal and external factors [78]. Internal factors are related to the structural characteristics of the alloy [79], as the chemical composition (mainly the contents of C and Mn) [80], austenitic grain size [81], its morphology [82] and a type of surrounding structural components. (ferrite, bainite or martensite) [83]. RA can be stabilized through diffusion of carbon or manganese, leading to lowering of the temperature of Ms [84]. When RA has a low carbon content, it transforms into martensite even with small strains, resulting in a reduced strain hardening rate. On the other hand, a high carbon content results in overstabilization of austenite, which causes reduced ductility. [80].

External factors that affect the stability of the  $\gamma$  phase are associated with the processing and service conditions of steel products, such as deformation temperature [85], strain rate [86] and the state of tension [87]. During forming operations, the temperature of the steel sheets increases locally due to friction with the tools and adiabatic heating of the deformed sheet.

The stability of the retained austenite includes thermodynamic and mechanical stability. The Ms temperature is used to characterize the thermodynamic stability of the retained austenite against transformation under cooling, while the Ms temperature is used to assess the mechanical stability against stress-assisted transformation. At temperature  $M_s^g$ , the stress reaches the slip yield stress in the austenitic phase [88–90].

If the stability of the austenite is low, then the austenite will transform very early during the forming operation, with no beneficial effect on formability. On the other hand, if the stability of the austenite is too high, the austenite may not be transformed at all during the forming operation. Therefore, an optimal level of stability is required to maximize formability. Austenite stability depends on the following factors: (a) chemical composition, (b) austenite particle size, (c) matrix strength, and (d) stress state. These parameters are graphed in Figure 7.



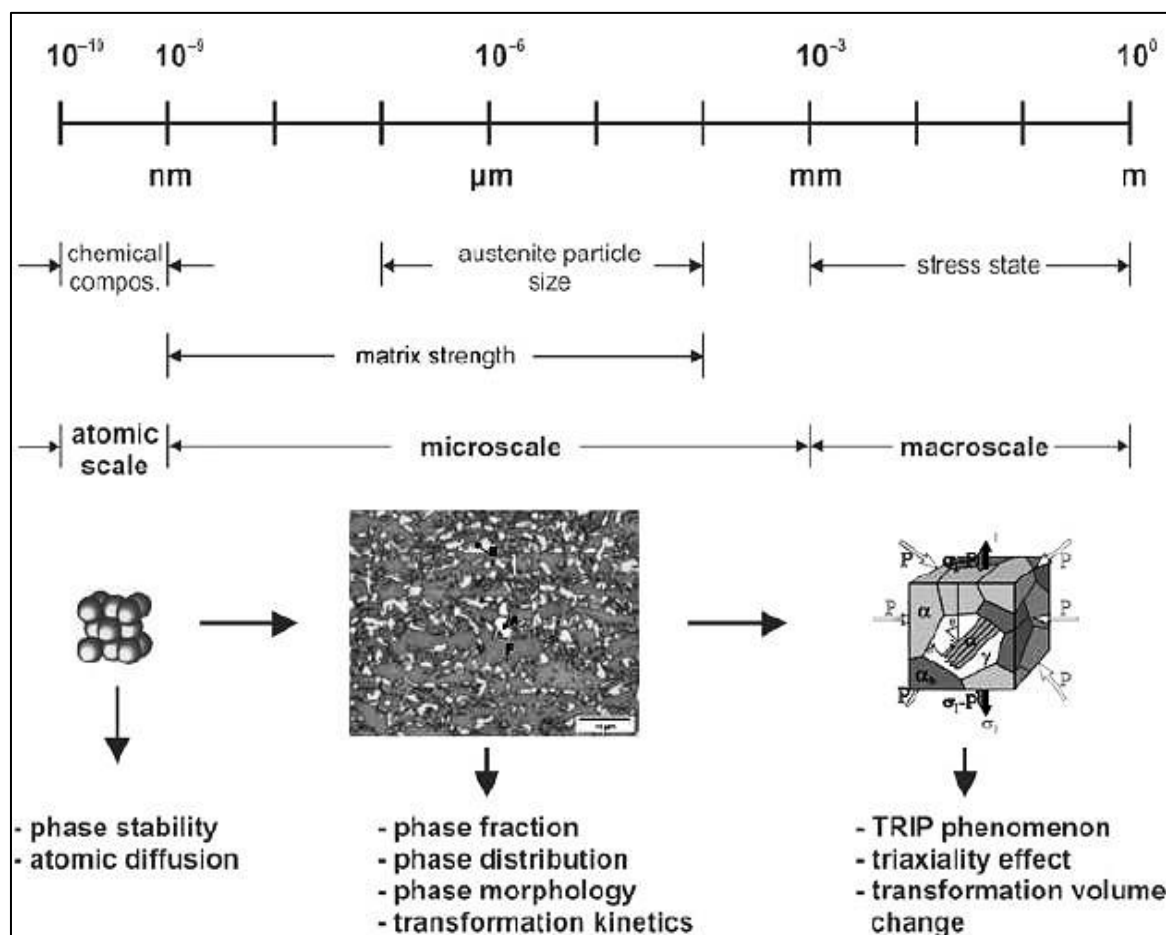


Figure 7. Parameters and scale levels that affect the stability of austenite

### 3.6. Stability Evaluation of Retained Austenite

In steels with TRIP effect, the formation of carbides is restricted by the addition of suitable alloying elements such as Si/Al/P. Therefore, most of the carbon is more likely to split to austenite and thus increase its stability. [91].

There are several experimental methods to determine the stability of retained austenite. One of them is the evaluation of the carbon content in the retained austenite, it is responsible for the MS temperature and, therefore, for the MD temperatures, the latter characterizing the stability of the austenite under deformation. The carbon content of austenite ( $C_{\gamma}$  wt%) can be determined from measurements of the austenite lattice parameter,  $\alpha_{\gamma}$ , for example, using equation (2) proposed by Sugimoto, K. et al., [92]

$$C_{\gamma} = (\alpha_{\gamma} - 0.35476) / 0.0467 \quad (2)$$

### IV. Effect of Alloying Elements on TRIP Steels

As already mentioned above, in Steel with TRIP effect, efforts are made to have a structure with ferrite, bainite and retained austenite for forming. Generally, this microstructure is obtained with steels in the composition range of 0.15%-0.20% C, 1.2%-1.5% Mn and 1%-1.5% Si. Silicon is needed to inhibit the undesirable precipitation of cementite ( $Fe_3C$ ). Since  $Fe_3C$  is not formed, the retained austenite is enriched in carbon during its isothermal transformation of bainitic. In the range of 0.5% to 1.5%, the addition of silicon results in ease of forming, due to the increasing amount of retained austenite. In hot-rolled steels, silicon also has an important favorable effect: promoting the formation of some equiaxed ferrite at high temperature. Manganese is added to suppress high-temperature decomposition of austenite, and also inhibits pearlite formation and stabilizes retained austenite necessary for the TRIP effect in forming.

The heat treatment described above will produce a structure consisting of ferrite, bainite, and austenite, with some martensite or cementite depending on process conditions. Ferrite will appear as equiaxed grains that formed during the intercritical temperature. Austenite that is kept at room temperature appears in small equiaxed or "blocky" grains at ferrite or bainite grain boundaries [43, 44]. The characteristic microstructure of these steels

has been shown in figure 3, which summarizes the influence of the main alloying elements to reach this structure. In table 1, a brief summary of the influence of each alloying element is made as a preamble to then go on to detail the influence of each one of them with greater scope.

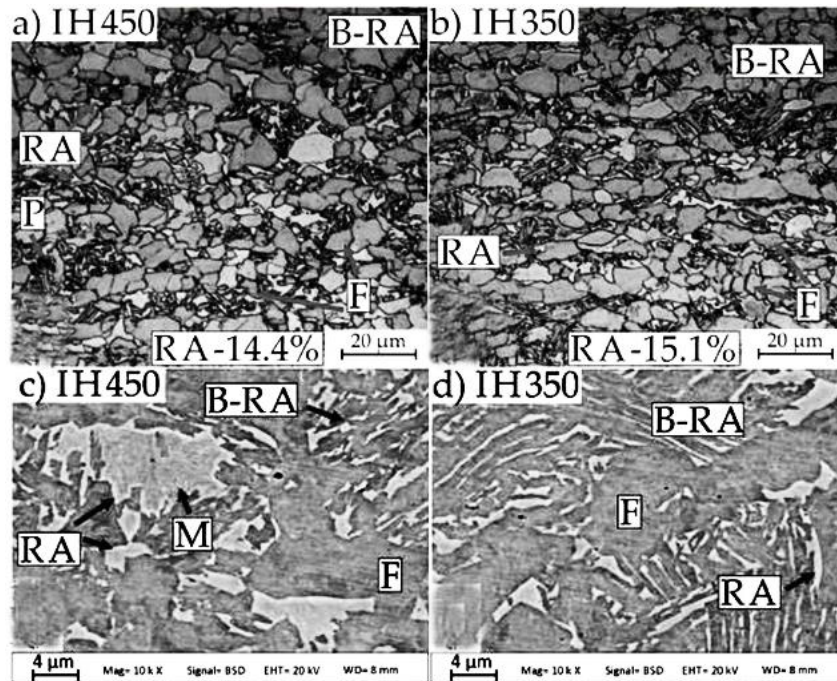
#### **4.1. Carbon effect.**

Carbon plays a key role in the composition of TRIP-assisted steels. It strengthens the austenite through interstitial hardening and improves its stability. The carbon diffuses and is enriched in austenite during the bainitic transformation. TRIP steels use higher amounts of carbon with sufficient content to stabilize retained austenite below room temperature. The stress or strain level at which the retained austenite begins to transform to martensite can be controlled by adjusting the carbon content. At lower carbon levels, the retained austenite begins to transform almost immediately after deformation, increasing the work hardening rate and formability during the stamping process. At higher carbon contents, the retained austenite is more stable and begins to transform only at strain levels beyond that produced during formation. At these carbon levels, the retained austenite persists in the part until the end, transforming into martensite during subsequent deformation. Solutes other than carbon add for the following reasons:

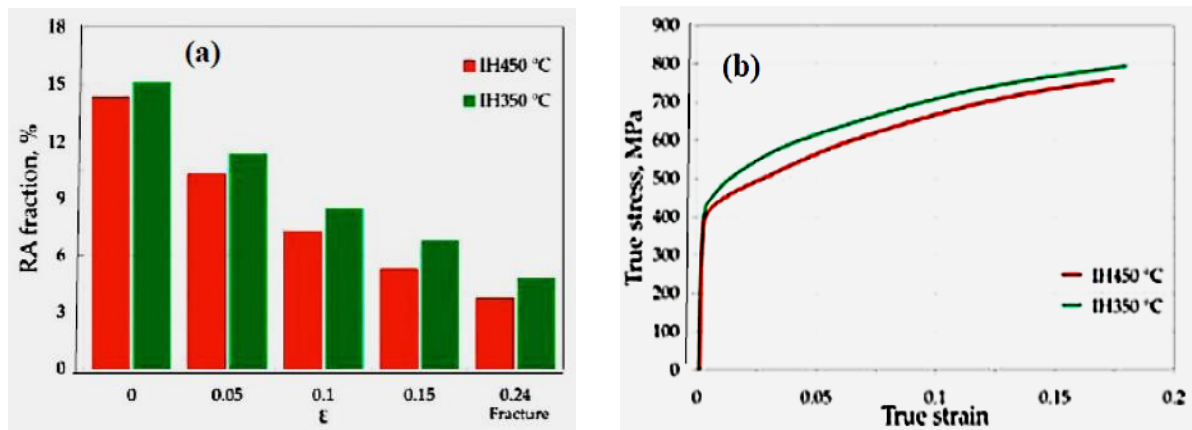
- optimize the fraction of retained austenite,
- control the precipitation of cementite,
- increase the hardness of the ferrite,
- increase hardenability to avoid pearlite formation before bainite reaction.

#### **4.2. Manganese effect.**

Manganese Mn, being an austenite stabilizer, expands the austenite loop and lowers the temperature of AC1, Ae3 and Ms, (martensite initiation temperature). The Ms temperature could decrease around 30 °C by adding only 1% Mn [93]. Mn plays an active role in increasing the volume fraction of retained austenite and can also induce significant solid solution hardening [94]. The typical content of Mn in a TRIP Mn steel is in the range: 0.2% - 2.5% (in percentage of mass) [95, 96] and generally not more than 2.5%. Mn content (more than 2.5%) is not favorable as it can lead to banding in the microstructure and excessively stabilized retained austenite [97], causing the TRIP effect to weaken or disappear. As the Mn content increases further, a transformation from austenite  $\gamma$  to martensite "ε" can occur, which has a hexagonal iron-shaped structure. [98, 99], which is characteristic of twinning induction giving rise to TWIP steels. But the TRIP effect can also be achieved with these medium and high manganese contents, by other procedures and/or by adding retarding elements in the event that the TWIP effect occurs. This was recently reported for a TRIP steel with 10% by weight Mn, which is a third generation AHSS. The martensite phase "ε" was observed to be nucleated in the intersection of bands. [100]. These third generation AHSS steels are called medium-Mn-TRIP Steel. Adam Grajcar et al. [101] studying the mechanical behavior and stability of retained austenite in a medium Mn (1.55%), low carbon (0.24%) and low alloy. Firstly, they performed an IA intercritical anneal at temperatures of 350°C and 450°C, and as a result established the need to obtain RA grains of various sizes (from 4 to 1  $\mu$ m) to guarantee a TRIP effect continuous throughout the strain range. The TRIP effect was achieved; However, grains smaller than 1  $\mu$ m turned out to be too stable or too difficult to transform, leading to a lack of contribution to the TRIP effect. In addition, the microstructure shown in Figure 8 was discussed. The RA fraction shown in Figure 9a) and tensile mechanical properties plotted in Figure 9b).).



**Figure 8.** Optical microphotography LM (a,b) and SEM (c,d) in the intercritical annealing of a low carbon medium manganese TRIP effect steel at temperatures of 450°C and 350°C (IH450 and IH350 simple ). The change in RA morphology due to temperature change is observe. [101]



**Figure 9.** a) The RA fraction as a function of strain for both isothermal treatment variants, b) True stress-true strain curves (uniform deformation range) for both isothermal temperature variants.[101]

#### 4.3. Silicon Effect

Silicon inhibits cementite formation in bainite during isothermal holding and splits carbon from bainite to austenite, resulting in increased retained austenite content. When the silicon content increases to 1.0% or more, the amount of retained austenite increases remarkably, leading to a good strength-elongation balance. Studies by Woo Chang Jeong [102] in the 0.33% C-1.03% Si-1.7% Mn steel showed a tensile strength of 1030 MPa and a total elongation of 34.5% when annealed at 780 °C for 5 min followed by holding isothermal at 400 °C for 5 min. In this case, the amount of retained austenite was around 25%. The variation in the combination of elongation tensile strength with annealing temperature and silicon content correlated well with retained austenite content, reflecting the TRIP effect of retained austenite. The most retained austenite was obtained in the annealed steel just above the AC1 temperature and the temperature giving the most retained austenite decreased with decreasing silicon content.

For TRIP steels. Si is a very important alloying element because it is a solid solution strengthening element that can inhibit Fe<sub>3</sub>C carbide precipitation. Previous investigations on TRIP steels with various Mn contents suggest that Si is also a useful alloying element, but in medium Mn content steels. In this regard, Yan L et al, [103] presented an investigation on the equilibrium phase fractions, the microstructure and the tensile

properties of 0.2C–5Mn–1.5Al steel (% by mass) with two percentages of Si (0% and 0.5%) after intercritical annealing (IA). The research results showed that the addition of 0.5% Si has no significant effect on the equilibrium phase fraction, but influences the optimal temperature of IA. Different intercritical annealing (IA) temperatures result in the retained austenite having different morphologies. In the 730 °C (0.5 Si) samples showed two morphologies: polygonal and lath, while the 760 °C (0.0 Si) sample only has retained polygonal austenite. The optimum tensile properties of the 0.5% Si steel are better than those of 0% Si steel due to the solution hardening effect of the Si and the higher stability of the retained austenite. Figure 10 shows the variation of the microstructure due to changes in AI temperature and Si concentration percentages.

#### 4.4. Effect of Molybdenum (Mo), Niobium (Nb) and Phosphorus (P)

Niobium is a ferrite stabilizer, and in solid solution it increases the amount of retained austenite. The reason is not clear, since fine Niobium carbide precipitates and complex NbMoC precipitates can also strengthen ferrite. Furthermore, niobium makes TRIP steel less sensitive to changes in processing parameters and refines the microstructure [44].

Molybdenum can be considered as an addition to conventional TRIP steel. Molybdenum in solid solution is a ferrite hardener and retards pearlite formation. In low silicon steels, its addition can lead to mechanical properties comparable to those of high silicon steels with ultimate tensile strength (UTS) greater than 1000 MPa with a total elongation of about 36%. On the other hand, it has been shown that Mo has a strong solute dragging effect on steel, which causes a delay in recrystallization and precipitation [104].

Furthermore, Mo has been reported to increase the solubility of C in austenite, which decreases the driving force of precipitation, as well as the diffusivity coefficient of precipitating carbide-forming species (i.e., NbC) in TRIP steel [127]. Molybdenum can also be used to suppress polygonal ferrite formation and promote the development of acicular ferrite structure or a mixed acicular/polygonal ferrite [128]. Moreover, Mo and Cr elements can increase the hardenability of the Steel, as well as the acceleration of the bainite transformation, which could affect the stabilization of the retained austenite during the isothermal bainite formation (IBT).

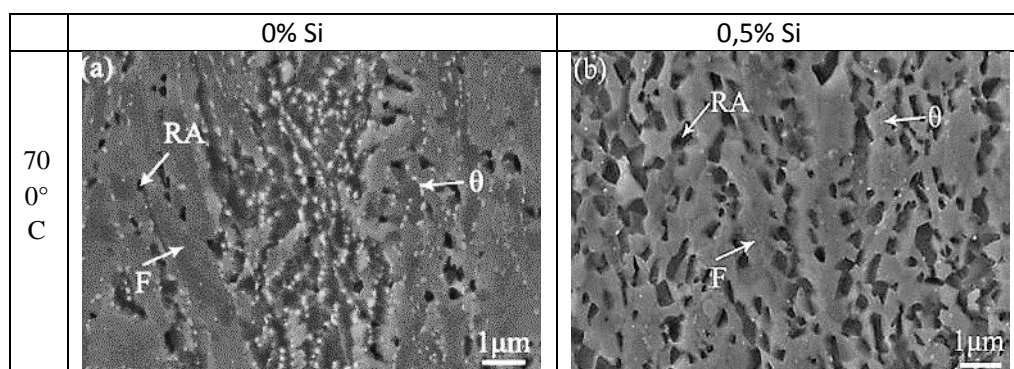
Molybdenum helps balance properties between strength and elongation in TRIP steels. Moreover, it has been observed in many studies on the TRIP effect that the highest tensile strength was obtained in the steel with the highest C and Mo content.

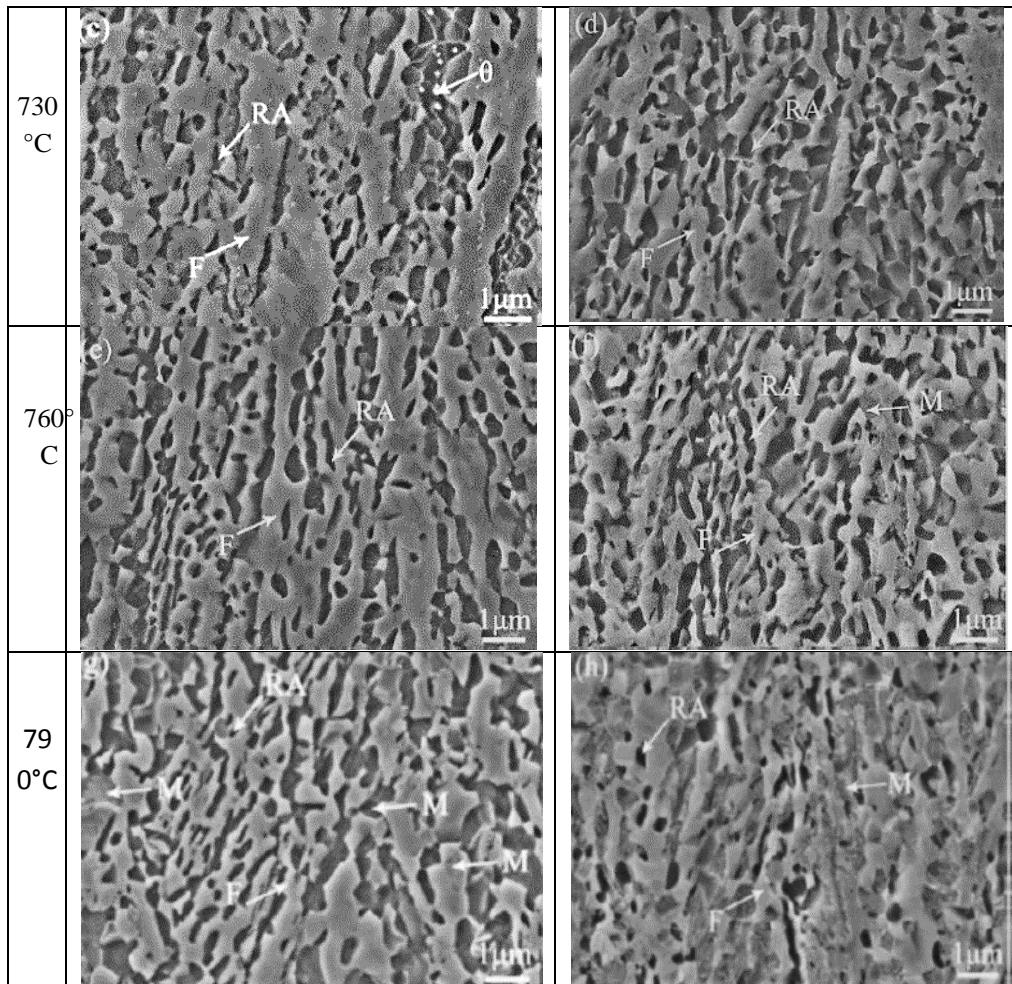
In the production of TRIP steels by hot working, increasing C and Mo contents have little effect on any change in the area fraction of recrystallized ferrite grains at low winding temperature, but decrease their fraction at high winding temperature. In short, higher winding temperature, indicating that ferrite recrystallization is delayed when TRIP steels are produced by this method.

Multiple addition of Cr and Mo in TRIP steels provides excellent impact toughness, which is mainly caused by uniform fine acicular bainitic ferrite matrix structure and a large amount of metastable retained austenite.

Multiple addition of Cr and Mo in TRIP steels provides excellent impact toughness, which is mainly caused by a uniform fine acicular bainitic ferrite matrix structure and a large amount of metastable retained austenite.

Chen et al. [129] showed that both phosphorous and silicon inhibit carbide formation, leaving more carbon in solution for segregation to austenite. In this study, fewer carbides were observed in phosphorus steel than in non-phosphorus steel. They also carried out in their study how to determine the effects of phosphorus in any TRIP steel. They found that increases in phosphorus increased  $f_y$ , especially in the presence of silicon. They attribute it to several different and simultaneous processes.

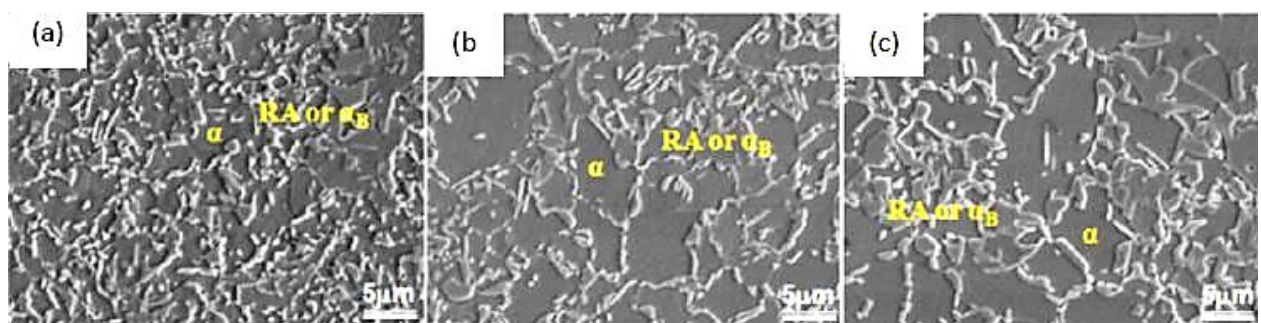




**Figure 10.** SEM micrographs. Effect of intercritical annealing temperature IA and two grades of %Si content in two TRIP steels. One of them does not contain Si (0%) and the other has 0.5% Si. It is observed that the addition of Si has little effect on the phase variation in the first stage of intercritical annealing. [103]

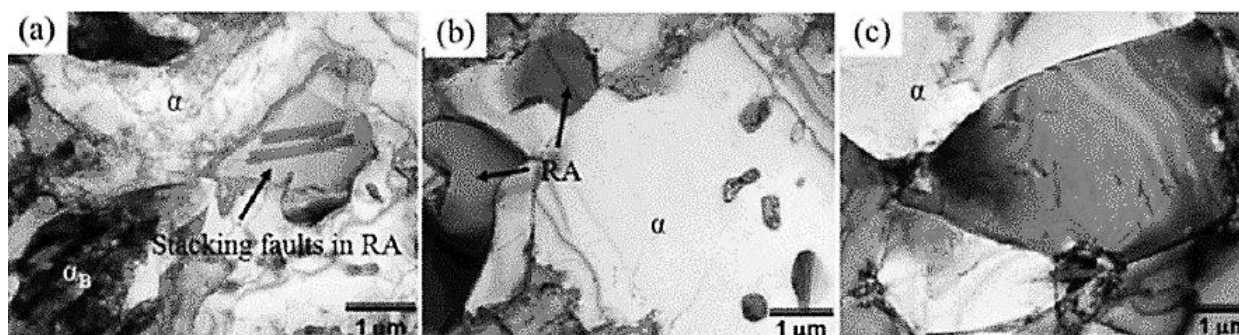
#### 4.5. Effect of Aluminum

Aluminum (Al) generally prevents the formation of carbides, leading to further C enrichment in the retained austenite RA [104, 105]. Like Si, Al also decreases the activity of C in the ferrite and thus increases the solubility of C in the ferrite. Furthermore, Al can accelerate the formation of bainite [106]. However, there are some disadvantages to adding Al: it can reduce solid solution hardening, greatly increase  $M_s$  transformation temperature even at room temperature, and reduce RA stability [107]. NamSuk Lim et al [108]., investigated the effect of aluminum in CMnSiAl (TRIP) steels with different aluminum contents (0.04, 1.00 and 2.00% by weight) but with the same percentages of C (or .20%) Mn(1.99%) and Si (1.87%) manufactured by thermomechanical processes. The microstructure shown in Figure 11 was obtained where it can be seen that the microstructures consisted of polygonal ferrite ( $\alpha$ ) and second phases such as bainite ( $\alpha_B$ ), martensite and RA.



**Figure 11.** Electron micrographs of CMnSiAl (TRIP) (a) 0.04-Al, (b) 1.00-Al, and (c) 2.00-Al with three different percentages of aluminum.

The grain size of the ferrite matrix increased with increasing Al content. In the case of 2.00-Al steels, a non-uniform microstructure containing coarse ferrite bands was observed. Figure 12 corresponds to a TEM microscopy where many thick stacking faults are observed in the RA as shown in Figure 12(a). This could be attributed to the increase in internal stress in the areas close to RA during the IBT bainitic transformation process. Stacking faults formed to relieve these internal stresses of the austenite. Figure 12(b) shows that most of the RA (is granular in form) and a small amount of RA is also observed in the form of a film or isolated inside the grains. Figure 12 (c), reveals that the number of second phases decreases with increasing Al content [108].



**Figure 12.** . TEM micrographs showing Al content-dependent microstructural changes of (a) 0.04-Al, (b) 1.00-Al and (c) 2.00-Al steels. [108]

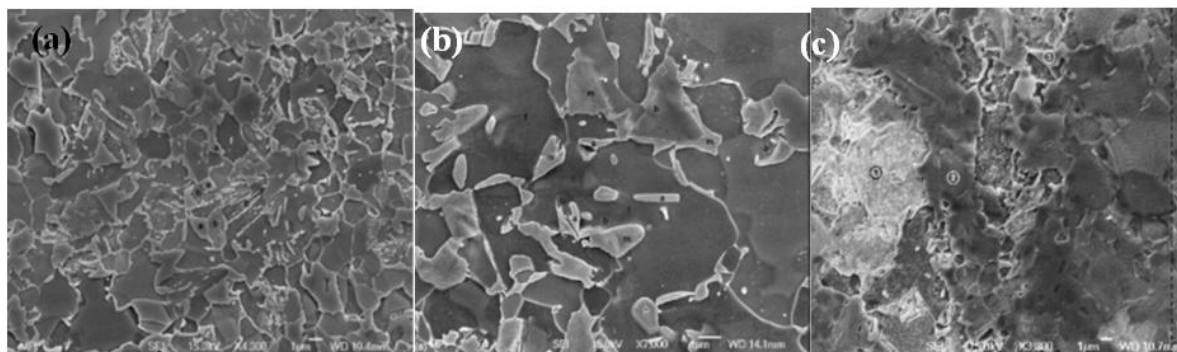
From the above analysis it appears that low levels of Al cannot completely suppress the formation of carbides. On the other hand, for higher levels, Al atoms diffuse into the ferrite from austenite only during the IA intercritical annealing process.

**Table 1. Summary the influence of alloying elements in the formation of TRIP steels - Most outstanding aspects**

Element	influence
<b>C</b>	Determines phase distribution. Determines stability of the retained austenite. It is the main hardener of martensite. Reduces toughness of martensite in plates. Determines fraction of austenite retained in the microstructure.
<b>Mn</b>	Stabilizes austenite. Reduces activity of carbon in the ferrite. Suppresses pearlite formation. Increases ferrite resistance.
<b>Si, Al, P</b>	reject carbon during the isothermal transformation to bainite. suppress formation of iron carbide Fe <sub>3</sub> C. stabilize the ferrite. accelerate formation of equiaxed ferrite at high temperature. harden ferrite (Si and P)
<b>Cr, Mo</b>	Decrease the activity of the carbon in the ferrite and stop the formation of pearlite.

### V. Effect of Intercritical Annealing (IA) Time and Temperature on Microstructure

As previously stated, the microstructure obtained in these steels depends on the various formation parameters, such as the chemical composition, time and temperature of IA, time and temperature of the isothermal process of bainite formation IBT, degree of deformation, etc., but all These factors must have the mission of having a material with a combination of the highest possible Strength-ductility ratio. S. Zaeferrer et al., [109] shows us the results of a TRIP steel: (0.2% C, 1.4Mn, 0.5Si, 0.7Al, 0.04 P), cold rolled up to approximately 65% reduction and controlled cooling with 20 C/s. Three samples exposed to intercritical annealing with the following parameters were used. At (400°C, 200s); B(350C, 200s); C (400°C, 400s). After the microscopy tests, the structures of ferrite, bainite, and martensite are exposed in figure 13.

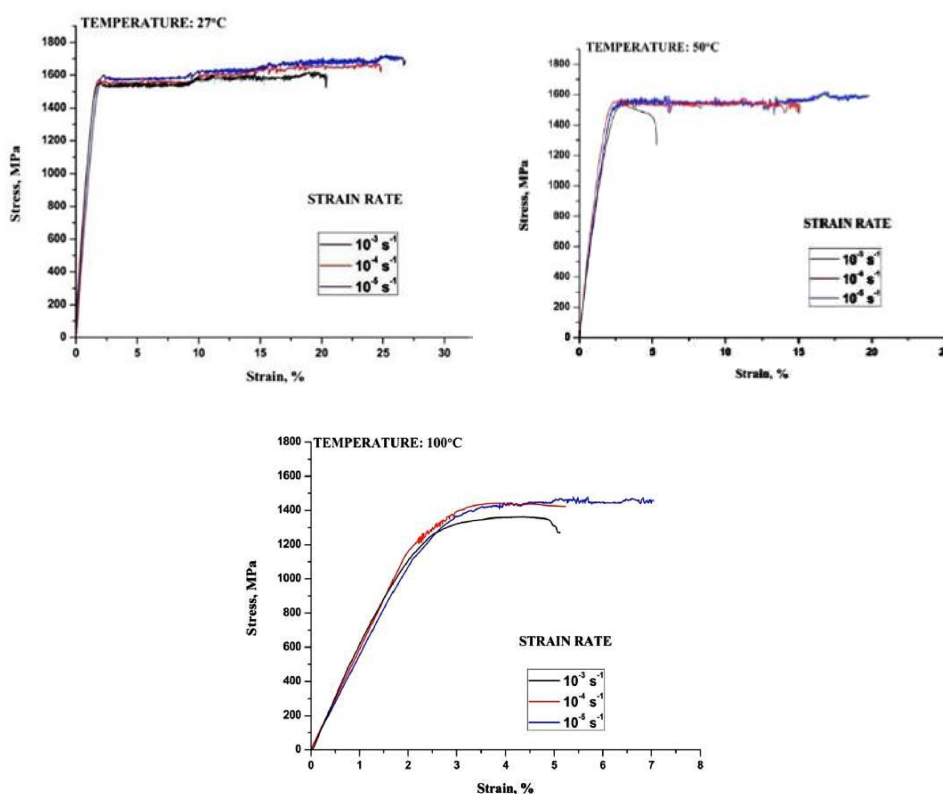


**Figure 13.** SEM micrograph showing typical structure of TRIP steels for three different IA treatment conditions: A) 400°C, 200s; b) 250°C, 200s; c) 400°C, 400s. In all three cases, the structures of Ferrite, bainite and martensite with islands of residual austenite are shown. [109]

### VI. Effect of Temperature and Strain Rate

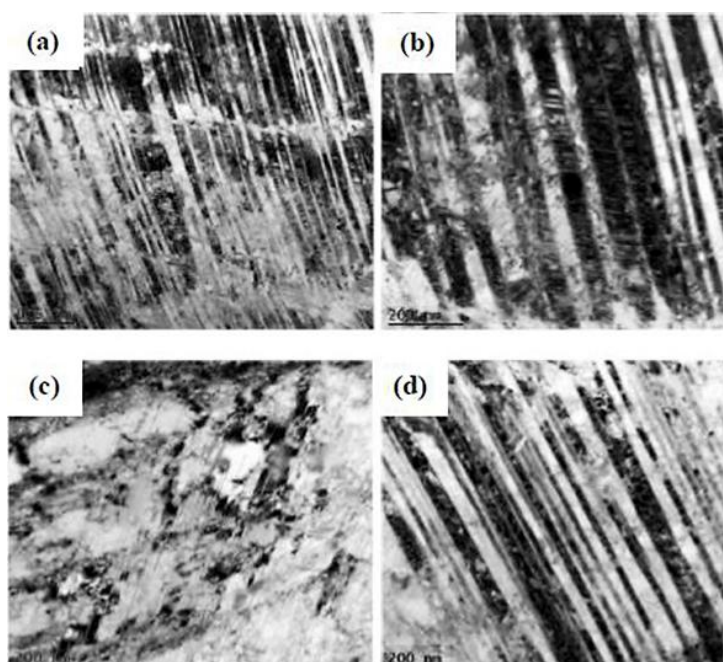
Apart from the above factors, the transformation of martensite in TRIP steel is also dependent on strain rate and strain temperature. It is important to predict this TRIP behavior in steels to be able to design according to the application. Several theoretical models have been proposed to quantitatively estimate the amount of martensite formed as a result of the phase transformation, taking into account temperature, strain, strain rate, and the amount of austenite present in the initial microstructure [110–112].

In high-alloy TRIP steels with medium manganese content, Lüders bands frequently appear [113]. The mechanism underlying this unstable plastic deformation resulting in the formation of Lüders bands is still unclear. However, there are many theories. One of which claims that reactions involving dislocations and C-Mn aggregates are the reason for these instabilities during plastic deformation [113, 114]. The formation of these bands is undesirable as it results in a poor surface finish on the final product, making it aesthetically unsuitable for most applications. It is important to note that these instabilities are highly dependent on strain temperature and strain rate and often disappear at sufficiently high strain rates [114]. It is also important to note that these instabilities are highly dependent on strain temperature and strain rate and often disappear at sufficiently high strain rates [114].

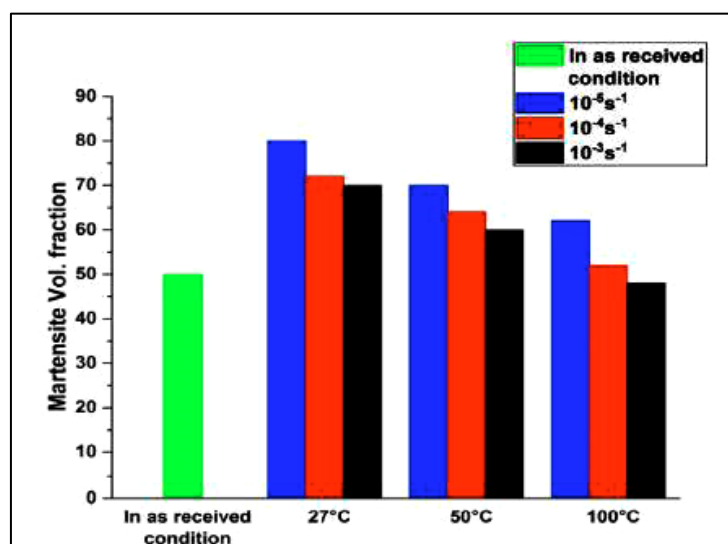


**Figure 14.** Engineering stress – strain curves for TRIP steel tested at (a) 27 °C, (b) 50 °C, and (c) 100 °C. [115]

J.V. Tilak Kumar et al. report a study on the influence of strain rate and temperature on an austenitic steel with TRIP effect [115]. Tensile strength, yield stress and elongation at fracture were observed to decrease with an increase in strain rate and strain at temperatures in the range of ( $10^{-5} \text{ s}^{-1}$  a  $10^{-3} \text{ s}^{-1}$ ) and 27 °C to 100 °C. The TRIP effect is suppressed when the strain rate and strain at temperature are increased, which allows attributing strength and ductility. Figure 14 shows some of the effects in relation to the microstructure and mechanical properties.



**Figure 15.** TEM bright field images after tensile deformation at a strain rate  $10^{-5} \text{ s}^{-1}$  at 27 °C (a) showing bundles of long twinned laths, (b) short twins within the long twinned laths, (c) twin planes containing islands of retained austenite, and (d) dislocation pattern within the laths. [115]



**Figure 16.** Martensite volume fractions before and after the tensile tests determined at different temperatures and strain rates.[115]

### VI. Correlation: Retained Austenite – Mechanical Properties

The excellent combination of strength and formability of TRIP steel depends mainly on the amount of retained austenite and its stability against transformation during deformation. The transformation of metastable retained austenite to martensite under strain, accompanied by volume expansion, results in a localized increase in the strain hardening coefficient.



Table 2 Retained austenite attributes of the studied TRIP steels alloys [116]

Designations	Property					
	Volume fraction of retained austenite ( $f_{\gamma}$ ), %	Lattice parameter of retained austenite ( $a_{\gamma}$ ), Å	Carbon content of retained austenite ( $C_{\gamma}$ ), wt.%	Total carbon content ( $C_{\gamma} + f_{\gamma}$ )	Transformable retained austenite index ( $C_{\gamma} / f_{\gamma}$ )	Stability coefficient of retained austenite ( $k$ value)
Si-alloy	23	3.6262	1.46	0.336	0.06	3.05
Si/Al-alloy	24	3.6194	1.25	0.30	0.05	2.52
Si/Al/P-alloy	20	3.6275	1.50	0.30	0.08	3.09

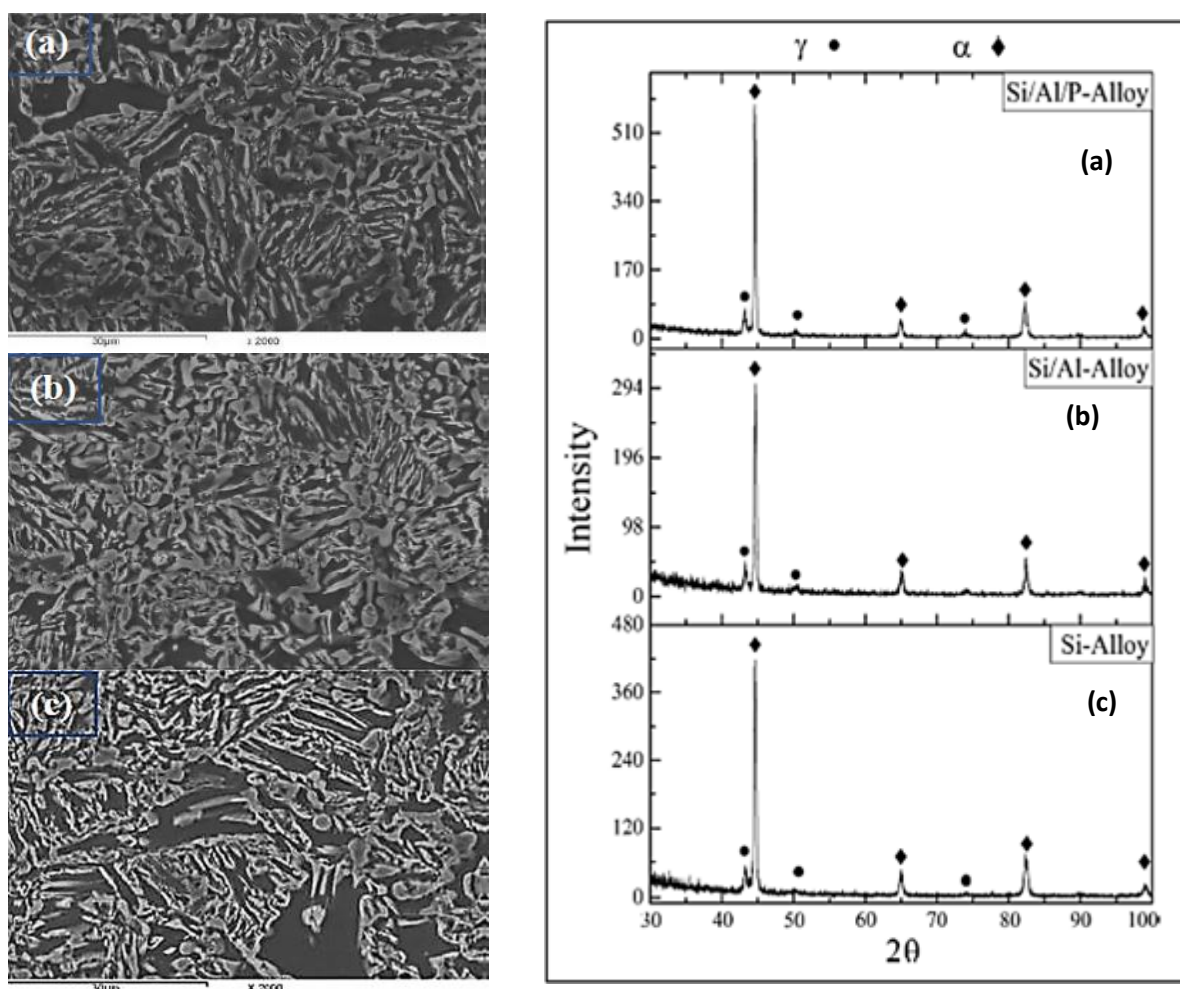


Figure 17. SEM micrographs of hot-forged TRIP alloys: (a) Si-alloy, (b) Si/Al-alloy and (c) Si/Al/P-alloy, with their respective XRD [116]

Hoda Nasr El-Din and Reham Reda [116] made a study on the Attributes of retained austenite and mechanical properties in TRIP steel alloys with different compositions, keeping constant the contents of C (0.32%), Mn (1.24%), Mo (1.22%) and varying the contents of Si, Al and P. The results on the austenite attributes found for these three samples of TRIP steels are shown in table 2. The most outstanding aspects are that the retained austenite content ranges from 20% to 23%. The Si/Al alloy contains the most retained austenite with the lowest carbon content, while the Si/Al/P alloy contains the least retained austenite with the highest carbon content. The Si alloy has intermediate values. It is a fact that is concluded, that as the volume fraction of retained austenite increases, its carbon content decreases.

Figures 17 shows the microstructures of the same alloys with their XRD where the presence of the same phases in the three samples is confirmed: Austenite saturated with carbon formed in intercritical annealing, which partially transforms into bainitic ferrite during isothermal holding in the bainitic range. This was accompanied by the redistribution and diffusion of carbon from the bainitic ferrite towards the islands of residual austenite. This carbon enrichment in the austenite increases its stability; therefore, austenite can resist transformation to martensite and be retained on cooling to room temperature. Table 3 presents the tensile

properties of the TRIP steel alloys studied and the stress-strain curves are shown in Figure 18. It is observed that the Si/Al alloy shows the highest values of UTS and total elongation of 788, 53 MPa and 27.89%, respectively. While the Si alloy demonstrates comparable tensile property values with the Si/Al alloy, the Si/Al/P alloy shows the lowest UTS values and a total elongation of 747.12 Mpa and 22.59% , respectively. The three TRIP steel alloys investigated exhibit comparable yield strength to UTS ratio values ranging from 0.716 for the Si/Al alloy to 0.733 for the Si/Al/P alloy. Table 5 reveals that the Si/Al alloy shows the maximum performance index level of 28,081.151 Mpa.%, and the Si alloy comes in the middle level of 28,970.29 Mpa.%, while the Si/Al alloy Al/P exhibits the weakest combination of 25,282.54 Mpa.% The balance between strength and ductility can be estimated using the performance index that describes the ability of the steel to absorb energy until fracture. This index can be obtained by multiplying the UTS and the total elongation [117-119,]

Table 3. Tensile Mechanical Properties of TRIP steel alloys [116]

Designations	Property					
	Yield strength ( $\sigma_y$ ), MPa	Ultimate tensile strength (UTS), MPa	Uniform elongation, %	Total elongation, %	$\sigma_y$ /UTS ratio	Performance Index, MPa.%
Si-alloy	570.96	787.69	26.49	35.65	0.725	28,081.151
Si/Al-alloy	564.59	788.53	27.89	36.74	0.716	28,970.29
Si/Al/P-alloy	547.76	747.12	22.59	33.84	0.733	25,282.54

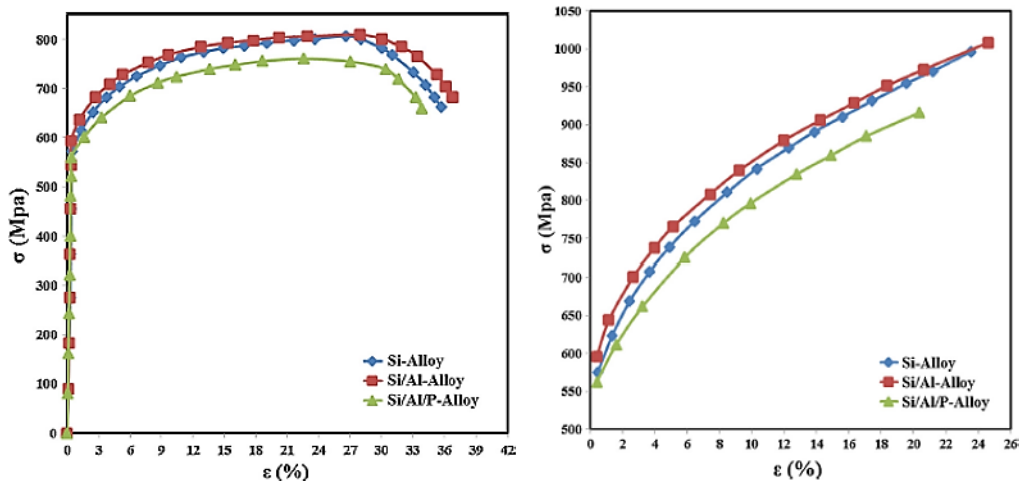


Figure 18. a) Engineering stress–strain curves of the studied TRIP alloys b) Engineering stress–true strain curves of the studied TRIP . These graphs correspond to the data in Table 3 and the microstructure obtained in Fig. 17. [116]

The main contribution of this section is reflected in the ability of retained austenite to improve mechanical properties. The key is not the amount of retained austenite that transforms to martensite during deformation, but the mode of transformation, that is, the regularity and gradation of transformation with deformation. It has also been observed that the replacement of Si by Al in the Si/Al alloy exhibits the best combination of retained austenite attributes and outstanding mechanical properties. On the other hand, the transformable retained austenite index  $C_{v_0}/\sqrt{V_0}$  and the retained austenite stability coefficient (k-value) are the most effective retained austenite attributes that can be used to reflect the efficiency of retained austenite in promoting mechanical properties. Lower values of these attributes express greater contribution of retained austenite in the improvement of mechanical properties.

### VI. Effect of Isothermal Bainite Transformation (IBT)

The critical step in the heat treatment of TRIP steels is to maintain the samples at temperature within the bainitic reaction range; that is, within the Austempered; in particular the holding time and the final cooling rate.

The final volume fraction of bainite increases as austempering temperature increase, which is typically between 325 and 475 C. Soaking at temperatures close to “bainite nose” depends on the composition of the steel. The vast majority of articles on TRIP steels describe simulations of isothermal maintenance at 400 °C.

Likewise, various authors have studied the effects of immersion temperature in the range: (465–475 °C) [120; 121; 122; 123].

The effects of temperature holding on bainite transformation kinetics depend largely on the chemical composition of the steel. IBT temperature investigations generally show increased elongation near the tip of the nose of the TTT diagrams. This phenomenon can be interpreted as the increase in the volume of bainite in a short retention time; which results in the enrichment of the  $\gamma$  phase, in addition to a significant amount of retained austenite. For example, experimental data for TRIP steels with 1.3–1.6% Al showed a relatively slow bainite reaction at 400 C, but at 490 C, a small driving force for the bainite reaction was determined, and consequently almost no carbon enrichment in austenite was observed. The temperature of 430 °C was proposed as optimal for hot galvanizing lines since the experimental data showed a significant formation of bainite and a notable amount of retained austenite [124].

If the immersion is carried out at high temperatures within the bainite region, it would negatively affect the TRIP steels by facilitating the formation of cementite.

At the beginning of retention in the bainite region, a large amount of austenite remains. However, the carbon level in austenite is still too low and therefore the MS temperature of austenite is relatively high. Consequently, most of the austenite should transform to martensite during the final cooling of the steel to room temperature.

The longer the immersion time, the higher the degree of completion of the bainitic transformation and the smaller the amount of austenite remaining. If the steel contains alloying elements that prevent the precipitation of carbides in the bainite, the stability of the remaining austenite is increased by carbon partitioning. As the bainitic transformation progresses, the amount of retained high-carbon austenite increases until it reaches a maximum.

The transformation of the bainitic is accompanied by a decrease in the MS temperature of the retained austenite. The “fresh” martensite progressively disappears from the final microstructure. With sufficient enrichment of austenite with carbon, the MS temperature drops below room temperature. This facilitates the formation of a microstructure containing ferrite, bainite and retained austenite stable at room temperature.

As a result of the opposing effects of the potentially available amount of remaining austenite and its stability, some investigators observed the extreme variation of the fraction of retained austenite with holding time in the bainite region. This is shown in Figure 19 which illustrates the general evolution of the microstructure during the isothermal process in the IBT range [125].

As shown in figure 19, ferrite fraction remains unchanged; increase in the bainite fraction until nearing completion initially leads to decrease in the martensite content at expense of increasing amount of retained austenite and its stability. However subsequently, reduces the retained austenite due to the formation of carbides in too long retention. The schematic presented illustrates the potential impact at the end of the process.

The effect of chemical composition and isothermal holding time at 400 °C on the amount of retained austenite was studied, in particular, by Zhao et al., [126] comparing two steels: (0.2C–1.52Mn + 1.8 %Al) with steels containing (1.4%Al + 0.25%Si or 0.96%Al + 0.25%Si). As shown in Figure 20, after ~3 min of retention, almost all available carbon was partitioned into retained austenite in high-Al and non-Si steels. At that time, the bainitic transformation became slowed down and the carbon-enriched austenite began to decompose. Higher Al and Si additions accelerated carbon enrichment of austenite [126]. Figure 20 shows the graphs obtained showing the volume fraction of retained austenite and the carbon content as a function of austempering time.

It can be stated: The isothermal holding time of austempering affected the stability of retained austenite. Prolonging holding lead to the cementite precipitation, which destabilized the austenite.

When the route of producing TRIP steels by hot rolling is followed in austempering step, is generally observed that a multiphase microstructure containing polygonal ferrite, granular bainite and retained austenite. So we have that LI Zhuang et al., studied the mechanical properties of steel Hot Rolled with Si Mn, observing these microstructures and also obtained the maximum values (776 MPa, 33% elongation using 25 min de austempering) [130].

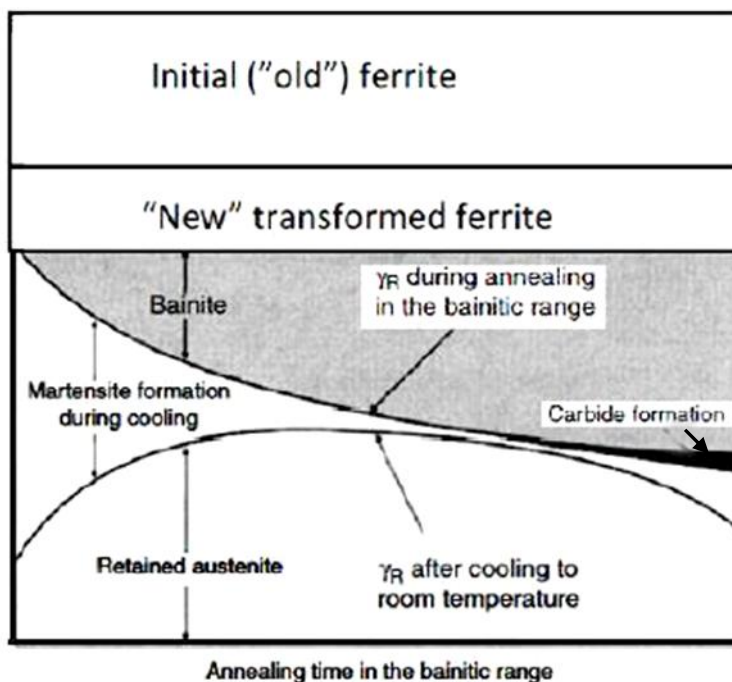


Figure 19. Schematic diagram of microstructural evolution in TRIP steels during austempering [125]

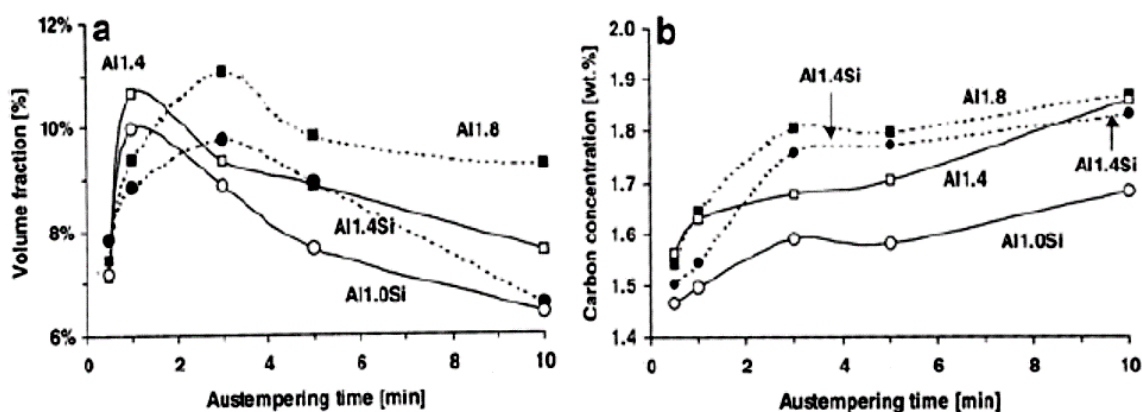


Figure 20. Volume fraction of retained austenite (a) and carbon content in retained austenite (b) as a function of austempering time [126]

#### IV. Conclusion

An extensive and detailed review of the factors that most influence the formation, Mechanical Properties and Microstructure of TRIP steels has been carried out. It has been based on the specialized and updated literature that exists on the subject; After the review, the following conclusions can be drawn:

1. All TRIP steels exhibit excellent combinations of high mechanical strength and high ductility. These steels have a wide range of mechanical resistance and elongation, which puts them at the forefront of the latest generation of advanced AHSS steels. But although these properties are high, they are related. Therefore, an adequate balance of properties must be taken into account in its manufacture. This balance must be in accordance with the conditions of use of this material, especially in applications in the automotive industry, especially in bodywork manufacturing.

This proper balance between strength and ductility comes from the strain-induced transformation of metastable retained austenite into martensite during forming. This quality is due to the fact that these steels have a multiphase microstructure that contains metastable ferrite, bainite and retained austenite, which stabilizes as deformation occurs.

2. There are many factors that influence the TRIP effect and therefore affect the mechanical properties of the steel, where the amount and stability of retained austenite plays the most important role. The stability of the austenite is what guarantees that the TRIP effect occurs.

3. The fundamental alloying elements are C, Si and Mn. The other alloying agents play a secondary role both to prevent harmful effects such as the formation of Fe<sub>3</sub>C carbides and to improve and control the efficient performance during the formation of these steels. The carbon content is the dominant element of the process; due to the formation of iron carbides, which is the most harmful effect in all the treatments that are followed to form TRIP steels. The content and percentage of carbon is another quite relevant factor, since it is closely related to the stability of the retained austenite.
4. The rate of deformation, the treatment temperature and the cooling rate are other factors that alter the microstructure, increasing percentage of retained austenite and its stability. Adequate monitoring of these variables should prevent formation of harmful phases. Rather, the microstructure corresponding to the TRIP steel with the desired properties must be optimized.
5. There are many theoretical models to predict the TRIP effect based on thermodynamic and kinetic aspects, but the hypotheses on which these models are based are restricted to special cases. Work is still under way to make a general behavioral model of the TRIP effect. But some particular models are validated by experiments and others are not. A general model has not yet been proposed or validated by experience; with which it can be affirmed that the TRIP effect is not yet fully understood.

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