

# Developments in Advanced High Strength Steels

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**Abstract:** Two major drivers for the use of newer steels in the automotive industry are fuel efficiency and increased safety performance. Fuel efficiency is mainly a function of weight of steel parts, which in turn, is controlled by gauge and design. Safety is determined by the energy absorbing capacity of the steel used to make the part. All of these factors are incentives for the U.S. automakers to use Advanced High Strength Steels (AHSS) to replace the conventional steels used to manufacture structural parts in the past. AHSS is a general term used to describe various families of steels. The most common AHSS is the dual phase steel that consists of a ferrite-martensite microstructure. These steels are characterized by high strength, good ductility, low yield to tensile strength ratio and high bake-hardenability. Another class of AHSS is the multi-phase steel which has a complex microstructure consisting of various phase constituents and a high yield to tensile strength ratio. Transformation Induced Plasticity (TRIP) steels is the latest class of AHSS steels finding interest among the U.S. automakers. These steels consist of a ferrite-bainite microstructure with significant amount of retained austenite phase and show the highest combination of strength and elongation, so far, among the AHSS in use. High level of energy absorbing capacity combined with a sustained level of high n value up to the limit of uniform elongation as well as high bake hardenability make these steels particularly attractive for safety critical parts and parts needing complex forming. In this paper, recent developments in the U.S. in all these various classes of AHSS will be discussed. Finally, martensitic steels with very high strengths are also in use for certain parts.

## 1 Introduction

Two major factors have been driving the development of new steel products for the automotive industry. The first is the desire for increased fuel economy, which in addition to design, is largely controlled by steel thickness. In order to address this aspect of reduced steel gauge, products with higher strengths and at least equivalent ductility/formability are needed. Initially, the focus was on body panels where development of high strength IF or bake hardenable steels allowed reduction of steel thickness. However, in addition to dent resistance, stiffness is another requirement of outer body panels and even with higher steel strength, thickness can not be reduced below a certain minimum from the stiffness point of view. So, while some efforts are still ongoing in this area, the majority of activity in the gauge reduction arena has been the development of advanced high strength steels (AHSS) for structural applications. The second driver is increased focus on safety performance and the major factor in this case is the ability of the steel for increased energy absorption. For this aspect, AHSS with increased total elongation and, in particular uniform elongation, are the steels of choice for development activities. In this paper, current

developments, both in the plant and in the laboratory, of AHSS are briefly discussed. It should be made clear, at the very outset, that this paper will not cover classical high strength steels such as rephosphorized LCAK or classic HSLA steels, but will include the following AHSS; dual-phase, multiphase (or complex phase), TRIP, and martensitic steels. Both cold rolled uncoated as well as coated (galvanized and galvanized) steels will be covered.

## 2 Dual Phase Steels

Among AHSS, dual phase steels are gaining the widest usage among automakers. This is because they provide an excellent combination of strength and ductility while at the same time are widely available due to the relative ease of manufacture. As a result, a large number of dual phase products have been developed and used in the U.S.

The following Table 1 is a summary of the dual phase product property requirements. Requirements for the same product sometimes vary widely; hence only representative property targets are listed. Also, requirements of total elongation depend on steel gauge; values shown here are for 1.6-2.0 mm range. Finally,

the test results are for JIS-T samples except for GI 600 DP where it is ASTM-L.

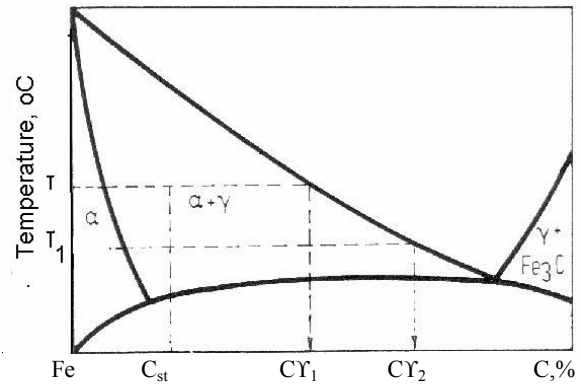
**Table 1 Dual Phase steels and their mechanical property requirements**

Product	TS (MPa) min.	YS (MPa)	TE (%) min.
Cold Rolled 590 MPa Dual Phase (CR 590 DP)	590	305-450	24
Cold Rolled 780 MPa Dual Phase (CR 780 DP)	780	420-550	14
Cold Rolled 980 MPa Dual Phase (CR 980 DP)	980	600-720	10
Galvanized 600 MPa Dual Phase (GI 600 DP)	600	340-410	23
Galvanized 780 MPa Dual Phase (GI 780 DP)	780	420-550	14
Galvannealed 590 MPa Dual Phase (GA 590 DP)	590	300-410	23
Galvannealed 780 MPa Dual Phase (GA 780 DP)	780	440-560	12
Galvannealed 980 MPa Dual Phase (GA 980 DP)	980	600-720	10

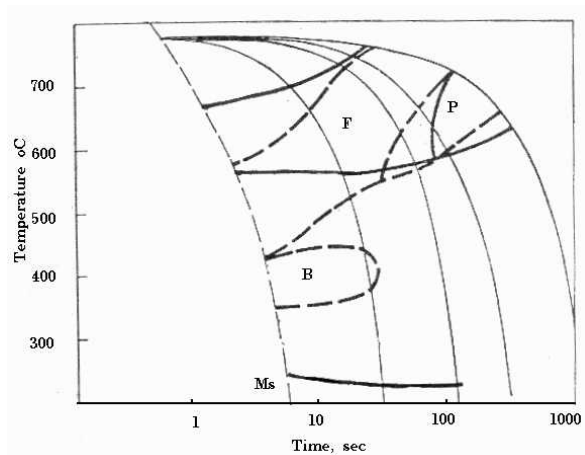
All the steels developed are based on annealing in the two-phase (inter-critical) temperature region and the consequent increase in carbon in austenite in comparison with the average carbon content in the steel. Thus, as shown in Fig. 1, carbon in austenite at a lower inter-critical temperature  $C_{\gamma_2}$ , is higher than carbon at a higher temperature,  $C_{\gamma_1}$ , at the same total steel carbon content.

The comparison of CCT diagrams after intercritical annealing with CCT of the same steel after annealing in  $\gamma$  region, i.e. after complete austenitization, displays some critical features of their difference shown in Fig. 2<sup>[1]</sup>. Higher carbon content in austenite after intercritical annealing results in a significant shift of pearlite transformation towards lower temperature and slower cooling rates. It is clear that the relative fraction of the formed ferrite always increases, and significantly at certain cooling rates.

The effect of intercritical annealing is not confined only to higher  $C_{\gamma}$  in comparison with the fully austenitized condition. The acceleration of “new ferrite” formation due to the presence of pre-existing phase boundaries and corresponding repartitioning of



**Figure 1 Pseudo binary Fe (Me)-C diagram, illustrating concentration of carbon in austenite as a function of heating in two-phase region**



**Figure 2 CCT diagrams for a steel of 0.14C-0.3Cr-1.2Mn-0.5Si-0.002B cooled from inter-critical temperature range (solid lines) and from  $\gamma$  region (dashed lines)**

carbon has very important consequences for production of dual-phase and TRIP steels. In particular, for the steel presented in Fig. 2, the resulting enrichment of remaining austenite by carbon completely suppresses bainite reaction and evidently decreases  $M_s$  temperature at lower cooling rates. In other steels, as shown by the authors<sup>[2]</sup>, it shifts the bainite transformation toward lower cooling rates.

## 2.1 Cold rolled dual phase steels

The cold rolled dual phase steels described here have been developed using the advantages of the water quench continuous annealing line at Mittal Steel.

As is clear from Fig. 1 the closer to  $A_{c1}$  the annealing temperatures are, the higher the  $C_{\gamma}$  (carbon content in austenite) and higher its hardenability. Thus, effects of annealing temperature ( $T_{an}$ ) and cooling rate are interrelated. The lower the  $T_{an}$  in the  $\alpha+\gamma$  region and

therefore the higher  $C_{\gamma}$ , the lower the permissible cooling rate that allows martensite transformation while avoiding pearlite and/or bainite transformation.

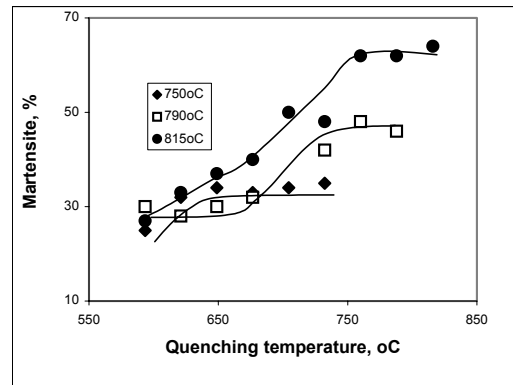
Direct quenching from inter-critical temperature range allows achieving very high strength of steels without expensive alloying. By water quenching directly from the inter-critical region but without initial slow cooling, any desired volume fraction of martensite, which will be equal to the amount of formed austenite, can be obtained<sup>[3]</sup>. However, the final properties of directly water-quenched steels are sensitive to fluctuations of annealing parameters, which affect the amount of formed austenite.

Interrupted cooling cycle is a combination of beneficial features of direct quenching and relatively slow initial cooling. The lowest temperature of water quenching is important for the shape of steel sheet, its YS/TS ratio and partly its elongation. Fig. 3 presents effects of quenching temperature,  $T_q$ , (beginning of water quenching) at various annealing temperatures on properties of 0.1C-1.5Mn- 0.3Si steel. Similar results were presented for various amounts of C and Mn in an earlier work<sup>[4]</sup>. As shown, the lower the annealing temperature (higher stability of austenite), the larger the temperature plateau of quenching temperature where no changes in volume fraction of martensite and therefore TS occur.

As was shown in the earlier work<sup>[4]</sup>, after overaging at 260 C it is possible to meet the necessary requirements of CR 590DP grade (see Table 1) at a rather wide range of C and Mn content,  $T_{an}$ , and  $T_q$ . The optimal chemistry was determined based on the best combination of flexibility of annealing (hardenability) and weldability. A 0.10C-1.0Mn-0.3Si steel is produced commercially using annealing at 770 C, quenching from 730 and overaging at 220 C. Typical properties of this CR 590DP grade are presented in Table 2.

The same principles can also be used to produce CR 780 DP and CR 980 DP. Higher volume fractions of martensite are obviously needed for the higher strength steels and the processing parameters have to be controlled accordingly in order to achieve the required properties. Typical properties for these products are given in Table 2.

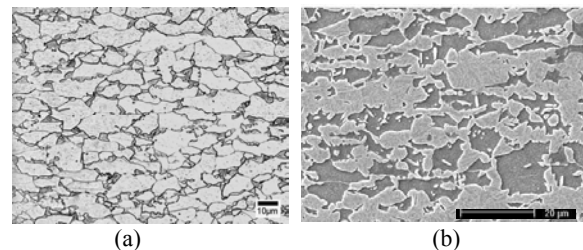
Representative microstructures for two of the products CR 590 DP and CR 980 DP are shown in Fig. 4.



**Figure 3** Volume fraction of martensite as a function of Quenching Temperature after various Annealing Temperatures; for a 0.1C-1.42Mn steel

**Table 2** Mechanical properties of cold rolled dual phase steels

	TS (MPa)	YS (MPa)	TE (%)
CR 590 DP	625	370	26
CR 780 DP	820	470	18
CR 980 DP	1030	675	13



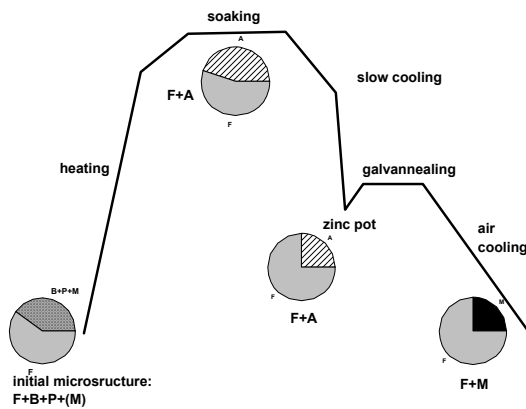
**Fig.4** Typical Microstructure of (a) CR 590 DP and (b) CR 980 DP

## 2.2 Galvannealed dual phase steels

Property requirements of Galvannealed Dual Phase Steels have been presented in Table 1. Low YS and therefore low YS/TS ratio can be achieved only by obtaining ferrite-martensite dual-phase structure.

A schematic of the metallurgical concept to obtain dual phase structure after galvannealing is presented in Fig. 5. Inter-critical annealing, as was described for above, can be used to obtain austenite enriched by carbon. The basic idea is to have such a combination of carbon and manganese content that it ensures a very high stability of gamma-phase, sufficient to prevent, as much as possible, any decomposition of austenite during galvanizing / galvannealing. The final austenite

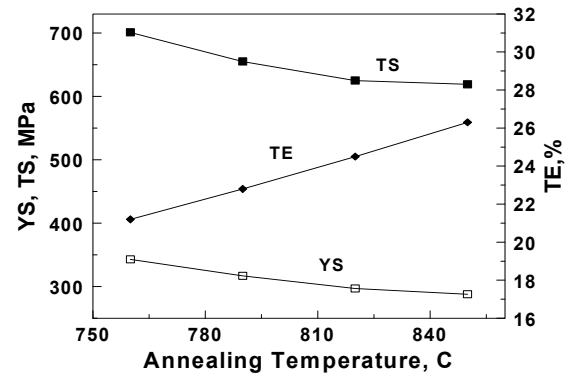
to martensite transformation should take place during final air-cooling.



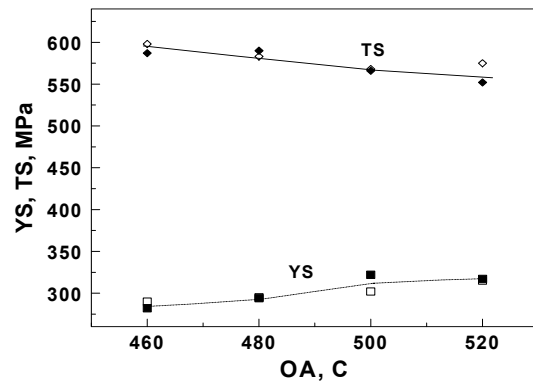
**Figure 5 Metallurgical concept of obtaining dual-phase during the galvannealing process**

Additional contribution to enrichment of austenite by carbon takes place during the initial, relatively slow, cooling typical for all galvannealing lines. As a result of low cooling rate there is enough time for “new ferrite” formation from austenite at sufficiently high temperatures when a near-equilibrium carbon redistribution from ferrite to remaining austenite can be achieved. This phenomenon has some important practical consequences such as significantly decreasing the sensitivity of the final structure and properties to annealing temperature. This feature has been observed by several workers and is sometimes called the effect of “self-stabilization” of dual-phase structure<sup>[5,6]</sup>. In fact, the higher the annealing temperature and higher the amount of initial austenite, the lower its stability due to its lower carbon content and the greater the portion of “new ferrite” formation. As a result of variable amount of “new” ferrite, a roughly constant amount of martensite (and tensile strength) can be produced in dual-phase steels over a reasonably wide “window” of annealing temperatures, sometimes up to 40-80 C.

The GA 590DP steel has been developed using the cycle presented above. The chemical composition is a steel of ~0.06-0.10% carbon alloyed only by manganese, Its high Mn content combined with a high carbon content in the final portion of austenite, which can reach 0.4-0.6%, ensures both high hardenability of austenite and low martensite start temperature,  $M_s$ .



(a)



(b)

**Figure 6 Effect of (a) annealing and (b) overaging (galvannealing) temperature on tensile properties of GA 590DP**

Zinc pot temperature is practically constant (~460 C), and cooling rate is determined by line speed, which cannot be changed without sacrificing productivity. Thus critical variable parameters of processing GA 590DP steel include annealing and galvannealing temperatures,  $T_{an}$  and  $T_{ga}$ , respectively.

Fig. 6a, showing mechanical properties as a function of annealing temperature of this steel, confirms the low sensitivity of tensile strength (volume fraction of martensite) to annealing temperature, as discussed above, inherent to slow cooling from two-phase temperature region. On the other hand, total elongation increases with higher annealing temperatures due to a larger volume of ductile “new ferrite” formed by decomposition of a larger volume of lower carbon austenite. Yield strength decreases in parallel to TS decrease due to a lower martensite volume fraction. So the highest  $T_{an}$  should be limited only by the need to achieve the necessary minimum tensile strength.

Fig.6b presents effects of  $T_{ga}$ . The higher the galvannealing temperature, the more chances that the remaining austenite would decompose partly by bainite reaction before martensite transformation during final air-cooling. This results in lower final strength of steel. Thus,  $T_{ga}$  should be kept as low as possible while still being sufficient to ensure quality of coating.

While the basic metallurgical principles remain the same for higher strength steels such as 780 DP and 980 DP, higher alloying is of course, necessary.

### 2.3 Galvanized dual phase steels

Requirements of Galvanized dual phase product have been presented in Table 1. Obtaining dual-phase structure using common HDG lines can be achieved using the same concept shown in Fig. 5. The key factor of this approach is sufficient, rather high alloying (>2%Mn and/or additions of Cr, Mo, V and so on). However, this approach can result in welding problems on the part of both the manufacturer and the customer.

Mittal Steel has a HDG facility that is equipped with the Zinquent (ZQ) technology<sup>[7]</sup> whereby a strip can enter the zinc pot at temperatures as high as 600-630 C. Very high heat conductivity and a very large volume of zinc in the pot equipped with mixing devices result in very high cooling rate (close to 100 C/sec) in the range of snout temperature ( $T_{sn}$ ) to zinc pot temperature. This technology provides a unique opportunity to keep a higher amount of remaining austenite at leaner alloying. This is further enhanced by the other critical advantage of ZQ, which is to improve the wettability of the steel strip so that it is possible to have good quality coating at Si content as high as 0.65%.

At the same time, the contribution to strengthening by silicon provides the same product strength at less martensite volume fraction and therefore gives an additional option to decrease the content of carbon or alloying elements that could negatively affect carbon equivalent and, therefore, weldability of steels. Typical microstructure of GI contains the dominant martensite type with a very small portion of bainite as strengthening phases in ferrite matrix. Regardless of the combination of annealing and snout temperatures the final strength displays practically linear

dependence on volume fraction of obtained martensite that is typical for dual-phase steels.<sup>[8]</sup>

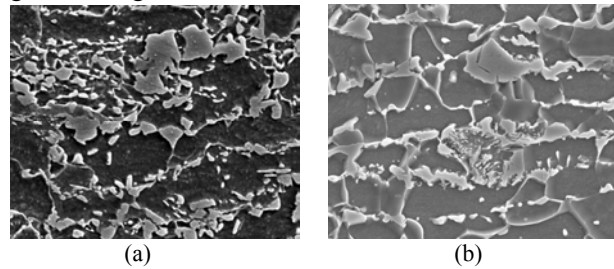
Again, higher strengths such as 780 or 980 MPa need higher amounts of alloying to increase hardenability of martensite under the relatively slow cooling rate of standard hot dip galvanizing lines. Typical properties of these steels are given in Table 3.

**Table 3 Mechanical Properties of Galvannealed & Galvanized Dual Phase Steels**

Product	TS (MPa)	YS (MPa)	TE (%)
GA 590 DP	620	355	26
GI 600 DP	650	360	26
GA 780 DP	795	490	15
GA 980 DP*			

\* under development

Typical microstructure of coated dual phase steels is given in Fig. 7.



**Fig.7 Microstructure of (a) GA 590 DP and (b) GI 600 DP steels, SEM, 3000X**

### 3 Multiphase Steels

Multiphase steels, also referred to as complex phase steels in Europe, are steels with a higher level of yield strength at the same comparable tensile strength levels of dual phase steels. Relatively few such steels are in use in the U.S.

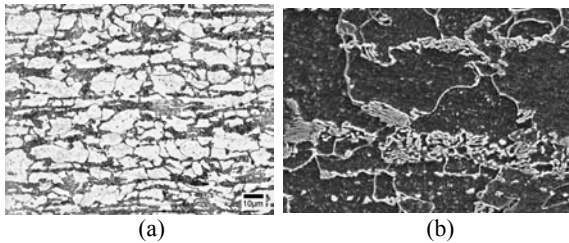
To obtain the high YS/TS ratio, different metallurgical principles need to be used for cold rolled and for galvannealed products. For cold rolled steels, achievement of higher YS/TS ratio of >0.7 is possible with a higher overage temperature on an appropriate dual phase structure. For galvannealed steels, however, higher YS cannot be obtained from an initial dual phase structure since in the galvannealing process, martensite is formed only in the final step of air-cooling and no further overaging is possible. The only way to gain yield strength in galvannealed multiphase structure is to obtain appropriate mixture of pearlite,

bainite as well as ferrite straightened by grain refinement and precipitation strengthening by Nb. Typical mechanical properties for these steels are given in Table 4.

**Table 4 Mechanical Properties of Multiphase Steels**

Product	TS (MPa)	YS (MPa)	TE (%)
CR 590 HY	690	515	23
CR 980 HY	1005	795	15
GA 590 HY	620	505	26

Typical microstructures for a cold rolled and galvanized multiphase steels are shown in Fig. 8.



**Fig.8 Microstructure of (a) CR 590 HY and (b) GA 590 HY, 3000X**

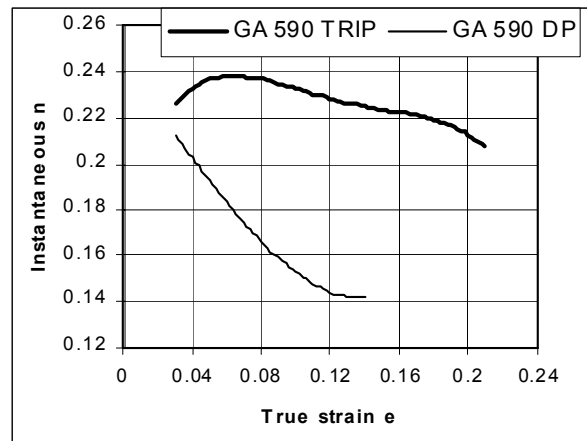
#### 4 TRIP Steels

TRIP steels, based on TRansformation Induced Plasticity effect offer the highest combination of strength and elongation [9], which is a measure of high level of energy absorption. Simultaneously, TRIP steels display high n-value up to the limit of uniform elongation as shown in Fig. 9.<sup>[10]</sup>

In addition, they also show high bake hardening compared to dual phase steels.<sup>[11]</sup>

Initial enrichment of austenite by carbon takes place during inter-critical annealing as shown in Fig. 1. The relatively slow initial cooling and the rather rapid cooling down to the temperature of isothermal holding results in further enrichment of the remaining austenite by carbon, enhancing its stability. Further growth in its stability occurs during the austenite to bainite transformation in the presence of strong ferrite forming elements. This significantly retards the carbide formation part of the bainite reaction and helps to keep all carbon in the remaining austenite. These elements are confined mostly by Si, Al, P<sup>[12]</sup>. As a result of such a high carbon content of more than 1-1.3% (authors data) or even 1.6%<sup>[13]</sup> in the final portion of austenite, the martensite start temperature

becomes lower than room temperature. This stable retained austenite (RA) transforms to martensite under subsequent mechanical stress/strain resulting in Transformation Induced Plasticity as discovered by Zackay<sup>[9]</sup>.



**Fig.9 Comparison of n vs. e curves for TRIP and dual phase steels of the same strength**

Implementation of this phenomenon to commercial production requires the development of a chemistry/cycle combination that ensures the necessary rate of bainite reaction to match with available holding time in the bainite region, inherent to a given facility. In other words, one should ensure overlapping the bainite temperature-time region with the real Bainite Isothermal Temperature Time (BITT) restrictions. For relatively long holding times, this bainite reaction should not occur at a high rate so as to prevent carbide formation and to retain a significant amount of high carbon austenite. In contrast, in HDG lines a composition, which is characterized by a fast bainite reaction at temperatures of zinc pot and galvannealing, should be selected so that even these short times would be sufficient for significant transformation to bainite.

Based on these principles, the following TRIP steels are being developed in the U.S.

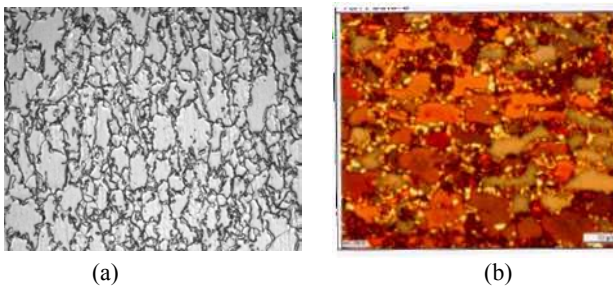
Typical microstructure for TRIP steels using LaPera tint etching<sup>[14]</sup> are shown in Fig. 10.

#### 5 Martensitic Steels

Using water quenching in a continuous annealing line, steels with 100% martensite are produced. These steels offer very high strength although ductility is

**Table 5 Target Mechanical Properties of TRIP Steels**

Product	TS (MPa) min.	YS (MPa)	TE (%) min.
Cold Rolled 590 TRIP	590	350-495	31
Cold Rolled 780 TRIP	780	410-500	21
Galvannealed 590 TRIP	590	360-510	26
Galvanized 590 TRIP	590	380-480	27
Galvannealed 780 TRIP	780	410-560	19
Galvanized 780 TRIP	780	440-500	21

**Fig.10 Microstructures of a typical TRIP steel using (a) Nital (1000X) and (b) LePera (2000X)**

lower than other AHS steels. The strength of the steel is controlled by the carbon content and a complete austenitizing temperature is used to obtain a fully martensitic structure. The martensitic steels in production in the U.S. is given below in Table 6.

**Table 6 Typical mechanical properties of martensitic steels**

Product	TS (MPa)	YS (MPa)	TE (%)
M 130	1054	923	5.4
M 160	1178	1020	5.1
M 190	1420	1213	5.1
M 220	1585	1350	4.7

## 6 Future

While dual phase and multiphase steels are already used in production vehicles in the U.S., TRIP will be in production vehicles starting as early as 2006. Martensitic steels have also been in use for bumpers and door beams for some time now. The next generation of AHSS are likely to be a new class of steels based on TWin Induced Plasticity, called TWIP steels. These offer very high elongations of 60-80% at comparable strength levels. However, since these

steels are heavily alloyed, their manufacturability and hence their eventual use remains to be seen.

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