NIOBIUM IN DUAL PHASE AND TRIP STEELS

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Abstract

Low alloyed multi phase steels are subject to extensive research efforts especially with regard to automotive applications due to their attractive combinations of mechanical properties. In this paper the microstructural characteristics differentiating multi phase steels from conventional cold formable steels are highlighted. The processing of dual phase steels and TRIP steels is described in detail and various alloying concepts are reviewed. Special emphasis is laid on the decisive role of niobium with respect to transformation behavior and microstructure development. It is concluded that the addition of niobium can provide a further improvement of the mechanical behavior and thus enlarge the application field of multi phase steels.

Introduction

For a great variety of applications the enhancement of strength remains one of the most important targets of material development. Economical and ecological considerations have even reinforced this tendency, as elevated strength levels are a prerequisite for light weight construction in the automotive industry which has gained increasing importance during the last years. To maintain or even strengthen the competitive position of steel for car body applications in spite of the distinctly higher specific weight in comparison with aluminum, magnesium, and plastic new material concepts have been developed and the maximum strength level of cold formable steels has been widely extended.

The conventional mechanisms to increase the strength of steel such as solid solution hardening or precipitation hardening are all based on an increase of the dislocation density or an increase in the different types of interactions with the dislocations. Unfortunately, in these cases the higher strength is accompanied by a noticeably inferior formability (Figure 1).
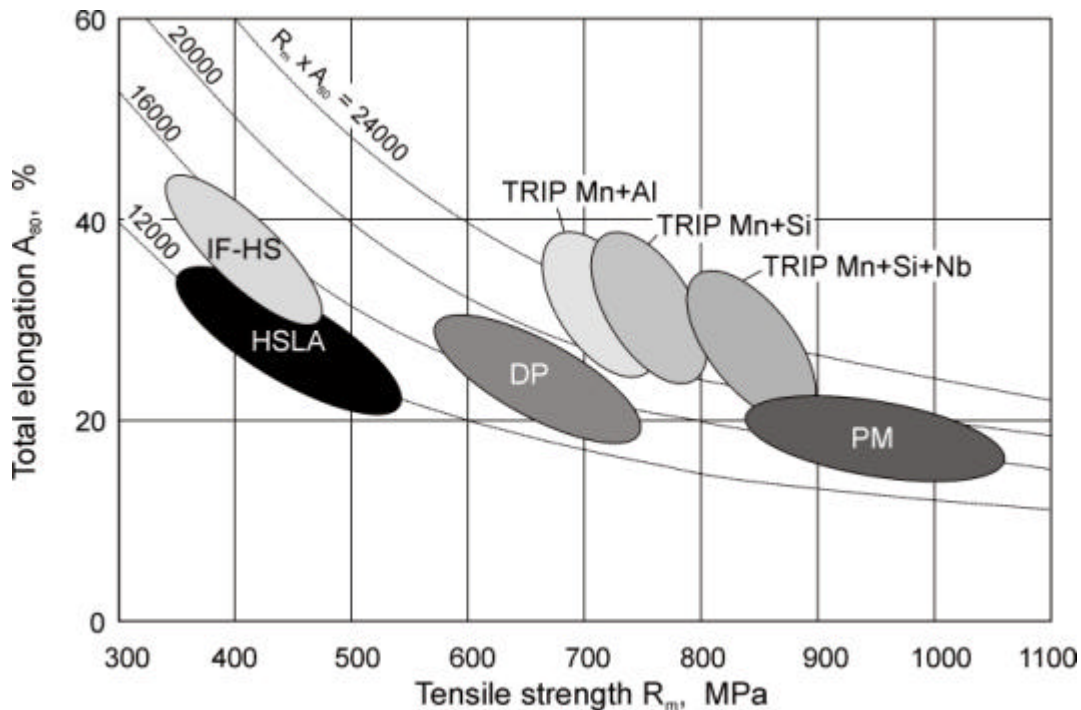

Figure 1: Ductility-strength relationship of several high strength cold rolled steels (IF - interstitial-free; TRIP - transformation-induced plasticity; DP - dual phase; PM - partially martensitic).

The introduction of a new group of steels with a microstructure consisting of at least two different components has led to an increase in the strength level without a deterioration of ductility. The so called multi phase steels therefore offer very attractive combinations of strength and ductility which are due to the coexistence of the different microstructural components, their different mechanical behavior and their mutual interactions. The components of interest are regarded on the scale of light microscopy, i.e. with typical dimensions of a few micrometers. Multi phase steels can for example contain a relatively soft matrix phase being responsible for a low yield strength and a good formability along with a high tensile strength as a result of the presence of a hard second phase as in the case of dual phase steels. It is possible to vary the mechanical properties and to tailor them for the respective application foreseen by adjusting type, morphology and orientation and above all volume fraction, size and distribution of the different phases.

Additional potential for the improvement of mechanical properties can arise from the presence of a phase which does not represent an equilibrium constituent and which transforms during forming when the activation energy necessary is supplied. The transformation induced plasticity of so called TRIP steels associated with the transformation of face centered cubic austenite to body centered cubic martensite has a positive effect on both strength and formability. Figure 1 provides an overview of the combinations of mechanical properties of various cold rolled high strength sheet steels including TRIP steels with different alloying concepts. Apart from the different levels of strength and ductility the flow behavior of dual-phase steels and TRIP steels shows characteristic differences to conventional cold formable steels namely continuous yielding in the first case and high uniform elongation and relatively low necking elongation in the second (Figure 2).
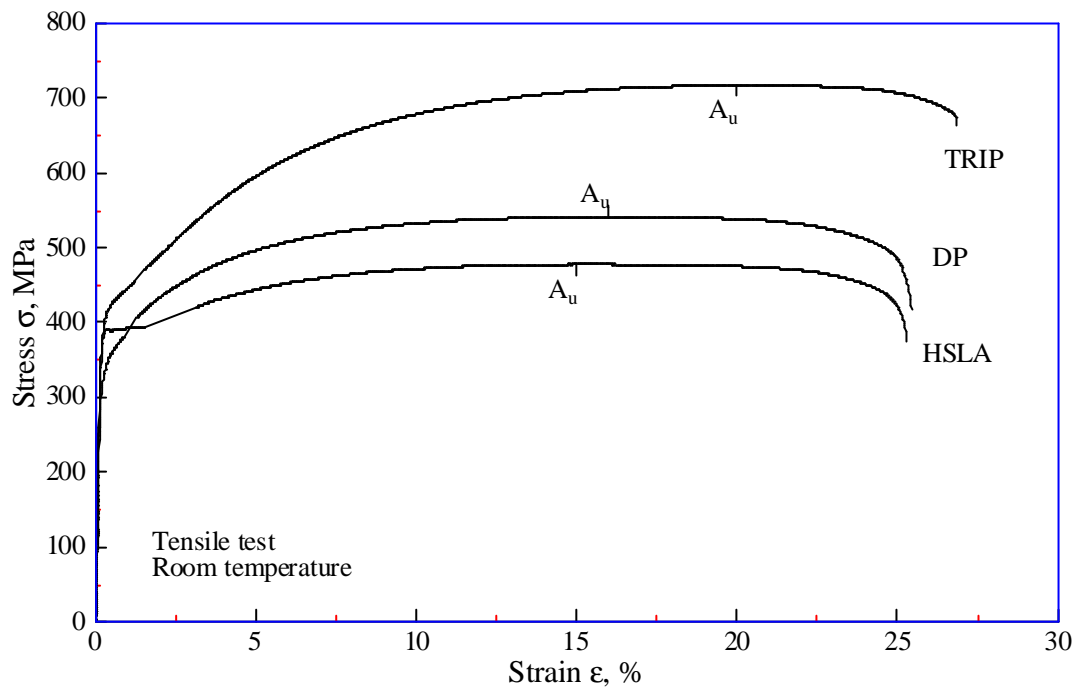

Figure 2: Stress-strain curves of high strength steels with approximately the same yield strength level and the specific features of different steel concepts.

Niobium is known as an alloying element by which austenitization, recrystallization, grain growth, phase transformation, and precipitation behavior can be controlled in a very efficient way and by which the mechanical properties can be varied in a wide range. With regard to multi phase steels niobium affects among other things the transformation of austenite to ferrite and to bainite and thus volume fraction and stability of the retained austenite which is the key to the outstanding mechanical behavior of TRIP steels. Thus niobium can be utilised as a metallurgical tool to adjust microstructure and properties.

In this paper the physical background, important process parameters, and different properties of multi phase steels with low alloying levels for thin flat products are discussed.

Microstructural Features of Multi Phase Steels

The microstructure of multi phase steels is significantly different from single phase microstructures and therefore requires additional information such as volume fraction, size, distribution, and morphology of the different phases to be completed described (Figure 3).

Dual-phase steels are characterized by small martensite islands in a ferritic matrix, while in duplex steels both phases are of the same order of magnitude with regard to amount and size. Duplex microstructures with austenite and ferrite components are of importance for stainless steels. TRIP steels are triple phase steels with two major fractions of ferrite and bainite and a minor fraction of retained austenite. In order to characterize this metastable microstructural component it is even necessary to determine the local chemical composition (1).
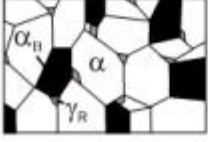

Figure 3: Microstructural characteristics of multi phase steels.

The mechanical properties of different microstructural components of steel are compiled in Figure 4. The data is based on the literature as well as on experiments and can only serve as an initial guide, as the distribution of alloying elements between the components is usually inhomogeneous and as the properties strongly depend on the process parameters, especially the formation temperature of the microstructural components (2).

The formability of multi phase steels is mainly determined by the ratio of yield stress and the volume fraction of hard and soft microstructural components. Figure 5 provides an overview of the combination of these important parameters for several multi phase steels. Additional importance is ascribed to the size and morphology of the constituents. The design of dual-phase steels makes use of a small volume fraction of the second phase and a great difference in yield strength. Due to the generation of martensitic islands associated with volume increase and shape accommodation, localized dislocation accumulations and internal tensile stresses in the ferrite occur, which result in a low yield strength and a high strain hardening rate. For duplex steels, however, a good combination of strength and formability is obtained with similar volume fractions and yield strength levels of the two phases (3, 4).

Phase	R$_e$, MPa	R$_m$, MPa	A, %	Hardness, H$_v$
Interstitial free ferrite	100 - 150	~ 280	~ 50	-
Ferrite (mild steel)	~ 220	~ 300	~ 45	-
Ferrite (0.7% Ni, 0.6% Cr)	~ 330	~ 550	~ 35	~ 180
Ferrite (13% Cr)	~ 300	~ 500	> 18	-
Pearlite	~ 900	~ 1000	~ 10	-
Cementite	~ 3000	-	-	800 - 1150
Nb carbonitrides	-	-	-	2500 - 3000
Bainite (~0.1% C)	~ 400 - 800	500 - 1200	≤ 25	~ 320
Martensite (~0.1% C)	~ 800	~ 1200	≤ 5	~ 380
Martensite (~0.4%)	~ 2400	-	-	~ 700
Austenite (18% Cr, 8% Ni)	~ 300	~ 600	> 40	~ 240

- no data available

Figure 4: Characteristic mechanical properties of different microstructure components in steel; the figures indicate the approximate range of properties.
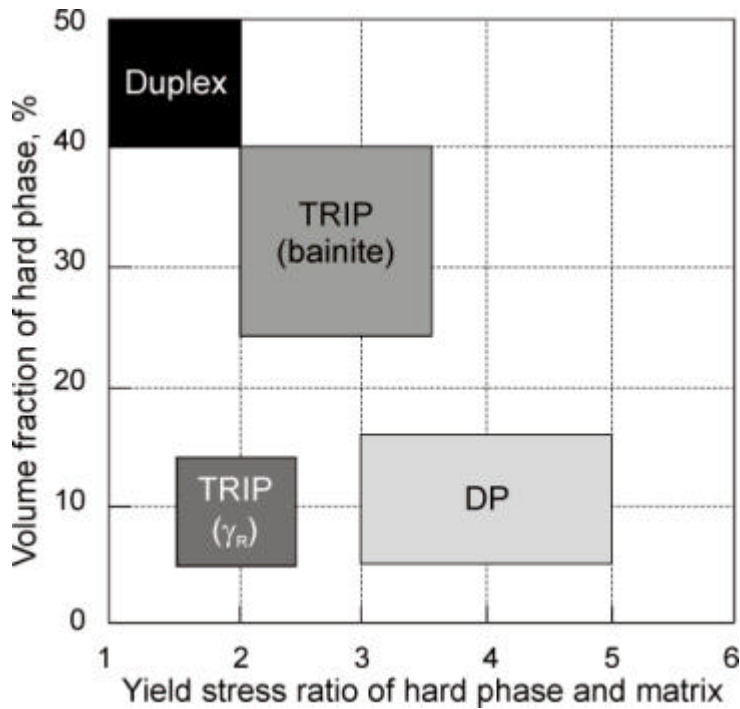

Figure 5: Volume fractions and yield stress ratios in multi phase steels.

Processing of Multi Phase Steels

Various process routes for dual phase steels and TRIP steels are either already in use or are subject to discussion depending on the product. Figure 6 summarizes the different routes and demonstrates that hot rolled, cold rolled and galvanized end products can be produced for all of

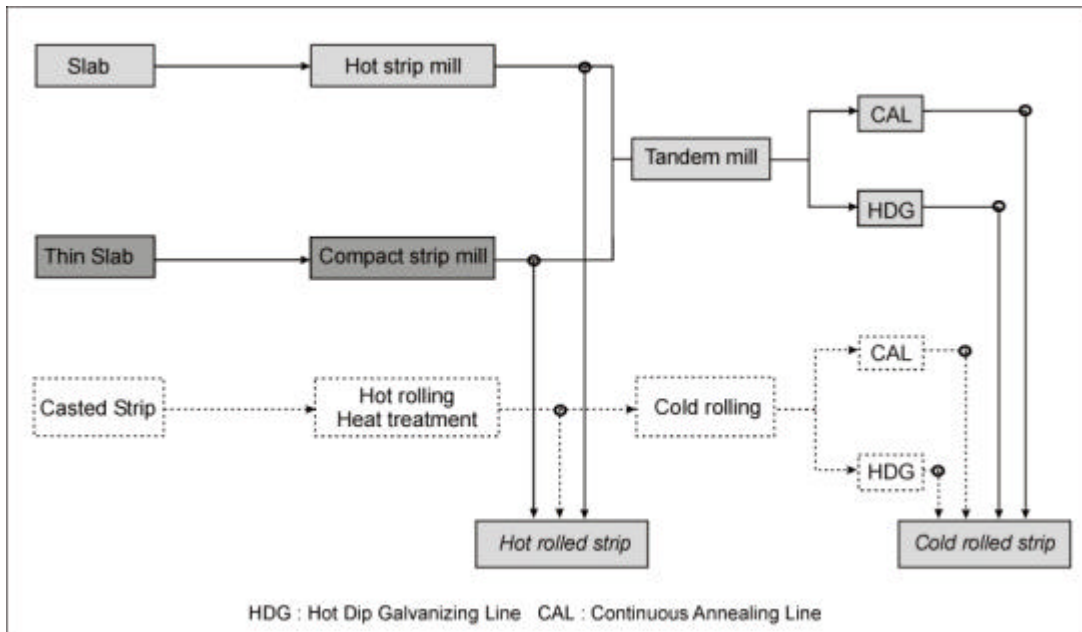them. Other concepts of multi phase steels such as complex phase steels are not discussed in this paper.

Figure 6: Process routes for hot and cold rolled multi phase steels.

Special attention has to be paid to the cooling strategy when producing hot rolled multi phase steels. After austenitization and the different steps of rolling in the roughing and finishing mills the microstructure and the mechanical properties are finally adjusted in the cooling section consisting of the runout table and coiler. A variation of the cooling intensity and the coiling temperature provides a change in the transformation behavior and allows the strength level to vary within in a wide range (5).

The temperature-time-schedule for the production of hot-rolled dual-phase and TRIP steels is presented schematically in Figure 7. The development of the desired microstructure might prove to be difficult as some transformations are necessary whereas others would be detrimental, which restricts the possible cooling path. Furthermore, the limited range of possible cooling rates on the runout table has to be taken into account. For dual phase steels the cooling rate must be low enough to enable the transformation of about 85 % austenite to ferrite to take place, associated with an carbon enrichment of the austenite, and at the same time high enough to avoid the formation of pearlite and bainite and to ensure the formation of martensite at low coiling temperatures of about 200 °C. Hence, a holding step has to be inserted in the temperature range of the maximum ferrite formation kinetics or the alloying concept has to be altered in order to accelerate ferrite formation (6).

For TRIP steels a lower cooling rate is utilised, since ferrite formation is delayed due to the different alloying approach in general and a higher carbon content in particular and since bainite is the desired microstructure. Coiling is therefore carried out in the temperature range of bainite formation at round about 500 °C and the final microstructure comprises 50 to 60 % ferrite, 25 to 40 % bainite, and 5 to 15 % metastable retained austenite that does not transform to martensite, since the carbon enrichment during ferrite and bainite transformation shifts the martensite start temperature below room temperature.

For cold rolled multi phase steels two different hot rolling approaches prior to cold rolling are conceivable the first of them leading to a soft material with a microstructure of ferrite and

pearlite. This microstructure stems from the application of a high coiling temperature of around 700 °C. The material is well suited to subsequent cold rolling. Alternatively, a hot rolling cycle making use of a lower coiling temperature in the range of bainite formation can be applied. As bainite forms at about 500 °C and represents a relatively hard microstructural component, distinctly higher rolling forces must be applied when cold rolling the material. However, it is expected that this process route leads to a more homogeneous and fine grained microstructure and superior properties of the as annealed product.
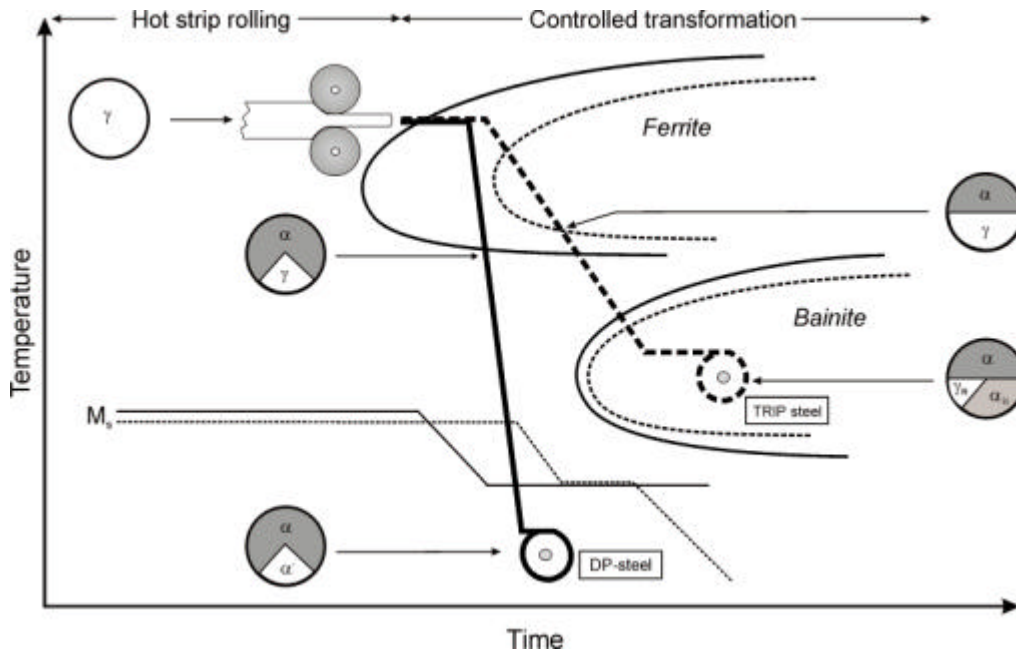

Figure 7: Temperature-time-schedule for hot rolled multi phase steels.

After cold rolling the sheet material has to undergo a heat treatment that can be realized in continuous annealing lines and in hot dip galvanizing lines. The volume fractions of the different phases of the final products are almost identical to those of the hot-rolled multi-phase steels, but the respective microstructures are created in the course of completely different processes. Consequently, details such as the distribution of elements as well as the precipitates in the phases and between the phases are most likely to differ from each other.

No full austenitization is performed during the heat treatment of dual-phase steels (Figure 8a). An annealing temperature slightly above A_{c1} is applied and only a small fraction of 10 to 15 % of the microstructure composed of ferrite and pearlite or ferrite and bainite is transformed to austenite during this intercritical annealing. Afterwards the material is quenched and the austenite transforms to martensite during cooling to room temperature, so that the final microstructure consists of a dispersion of martensitic islands in a ferritic matrix (7).

TRIP steels are subjected to a two step heat treatment with intercritical annealing in the temperature range between 780°C and 880°C, cooling and another isothermal annealing between 350°C and 450 C which is then followed by cooling to room temperature (Figure 8b). The microstructure of TRIP steels after intercritical annealing contains almost identical percentages of ferrite and austenite, but in contrast to the microstructure of hot rolled TRIP steels the ferrite has been part of the microstructure prior to annealing and does not form during cooling from the intercritical temperature. Cooling is interrupted in the temperature range of the bainite formation for several minutes, before finally cooling to room temperature. During the second isothermal holding the austenite is mostly transformed to bainite thus leading to a

final microstructure of approximately 50% to 60% ferrite, 25 to 40% bainite, and 5 to 15% retained austenite (8).

Again, the austenite of dual phase and TRIP steels is enriched in carbon due to the phase transformations and the martensite start temperature is lowered. The high carbon content of the austenite in dual phase steels retards ferrite, pearlite and bainite formation and reduces the critical cooling rate to ensure complete transformation of austenite to martensite, whereas the carbon enrichment in TRIP steels is even sufficient to suppress martensite formation during cooling to room temperature. This can be attributed to the higher carbon content of TRIP steels and the favorable conditions during the second annealing step. Bainite formation enables a further carbon enrichment to occur, whilst carbide formation that withdraws carbon from the austenite can be suppressed, which is indispensable in preserving a considerable amount of metastable retained austenite even at room temperature. If the second annealing is too short and if therefore too little bainite is formed, carbon enrichment of the austenite is not sufficient and the unstable austenite is transformed to martensite during cooling down to room temperature. Nevertheless, efforts are made to minimize or even to omit the isothermal bainite transformation step.
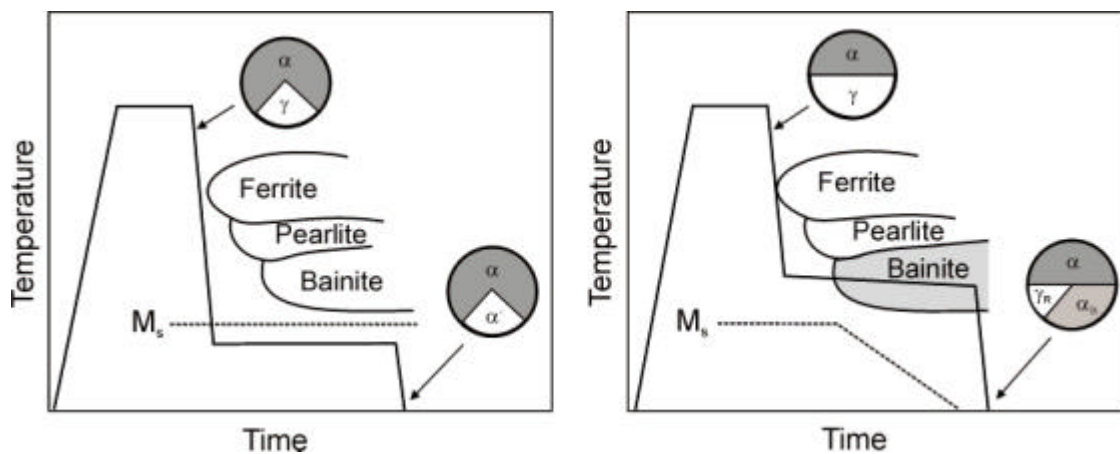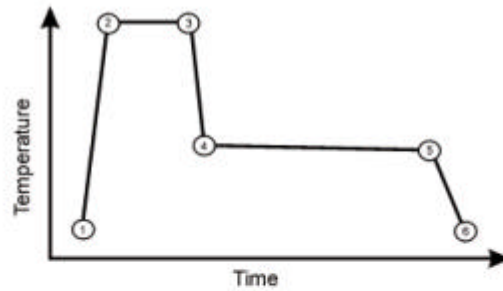

Figure 8: Temperature-time-schedule for cold rolled dual phase (a) and TRIP (b) steels.

The processes taking place during the various steps of the heat treatment of TRIP steels are listed in Figure 9. The essential requirement is to transform part of the initial microstructure to austenite, to bring all carbon atoms in solid solution and to prevent any precipitation in the austenite, so that a sufficient amount of retained austenite with a sufficient carbon content is obtained. The significant role of niobium is hinted at and will be discussed in detail.

The description of the production process already implies strict control of the process parameters which is necessary in order to produce the desired microstructure and mechanical properties. Hence, it is recommended that multi-phase steels are produced on hot strip mills with a coil box, which ensures a defined cooling rate, a constant finishing temperature without use of a speed up, and a homogeneous temperature distribution over the strip width. Figure 10 presents the demands that are characteristic of the production of multi phase steels, some drawbacks and handicaps, but also numerous advantages connected with the processing of dual-phase and TRIP steels. For hot-rolled products the combination of the improvements arising from a thermomechanical treatment and the benefits caused by multi-phase behavior must be highlighted; for cold rolled material the possible production of high strength steels by the application of relatively low rolling forces is highly remarkable.

Steps	Metallurgical features	Niobium effect
1 → 2	Recrystallization	Delay
	Dissolution of cementite	Acceleration
2 → 3	Alloying element and carbon segregation	Acceleration
	Grain growth	Delay, Fine grain
	Precipitation of carbonitrides	Formation of NbC
	Transformation of pearlite & ferrite to austenite	Acceleration / delay
3 → 4	Transformation of austenite to ferrite	Acceleration / delay
	Carbon segregation	Acceleration / delay
4 → 5	Transformation of austenite to bainite	Delay
	Carbon segregation	Acceleration / delay
	Possible precipitation of cementite	Acceleration
5 → 6	Possible transformation of austenite to martensite	Control of M_s-temperature

Figure 9: Metallurgical features during the processing of cold rolled TRIP steels.

Figure 10: Comparison of the process routes of hot and cold rolled multi phase steels.

Alloying Concepts

Numerous alloying concepts have been developed for dual phase steels as well as for TRIP steels in order to adjust the desired microstructure and properties. The alloying elements change the thermodynamic stability of the phases and the kinetics of transformation whereby the transformation temperatures are shifted, the transformations are either promoted or hindered

and the phase distribution is altered. Additionally, the elements might act as solid solution or precipitation hardeners and affect the grain size. Some of the most common alloying concepts for dual phase and TRIP steels are summarized in Figure 11 (9-14).

Steel	C	Mn	Si	Al	Cr	Nb	Other
DP	0.1	1.0	0.5				
"	0.1	1.0	0.1				P: 0.05
"	0.1	1.5	0.1	1.2	0.5		
"	0.12	1.2	0.1		0.8	0.04	
"	0.07	1.4				0.03	P: 0.03
TRIP	0.2	1.5	1.5				
"	0.2	1.5	0.1	1.8			
"	0.3	1.5	0.3	1.2			
"	0.15	1.5	0.6				P: 0.1
"	0.15	1.5	0.1	1.0			Cu, Ni
"	0.2	1.5	1.5			0.04	
"	0.2	1.5	1.1			0.04	Mo: 0.3

Figure 11: Alloying concepts of multi phase steels, contents in mass%.

The main alloying element is carbon by which all transformations are noticeably affected and by which the final microstructure and the mechanical properties are controlled. Carbon stabilizes the austenite which leads to the formation of martensite in the case of dual phase steels and to the retention of austenite in the case of TRIP steels. However, other requirements such as weldability limit the use of carbon to round about 0.2 mass %.

Other important alloying elements for dual phase steels are silicon and manganese. Silicon in the first place promotes ferrite formation, whereas manganese retards pearlite and bainite formation and allows martensite formation to occur during cooling from the intercritical temperature. Phosphorus or aluminium (sometimes in combination with chromium) can be used to reduce or replace silicon, which may cause problems during hot rolling and coating.

Niobium in the state of solid solution retards static and dynamic recrystallization during hot deformation as well as the austenite to ferrite transformation. Small carbonitrides which form when niobium combines with carbon and nitrogen also delay recrystallization, constitute an effective obstacle to grain growth and result in significant strengthening.

The addition of niobium to dual phase steels not only provides noticeable grain refinement and thus improved ductility but also renders an additional holding step in the temperature range of maximum ferrite formation unnecessary and enables continuous cooling to be applied after finish rolling (Figure 12). Ferrite formation involves carbon enrichment of the austenite thus retarding pearlite and bainite formation and facilitating martensite formation.

Any static recrystallization within the finishing mill is suppressed, as the solute drag effect caused by the presence of soluble niobium raises the temperature at which recrystallization can occur above the entry temperature to the finishing mill. The possible temperature range of hot deformation is extended, since the addition of manganese lowers the A_{c1} temperature. Consequently, the accumulation of strain finally either leads to dynamic and metadynamic recrystallization within the last stands of the finishing mill or even to a complete suppression of

recrystallization combined with a full pancaking of the austenitic microstructure, if strain induced precipitation takes place. In both cases the nucleation rate and the growth rate of ferrite is strongly enhanced due to the creation of grain boundaries, twin boundaries, dislocations, and deformation bands on a massive scale, so that ferrite formation is accelerated. The dynamically recrystallized structure is considered to be even more advantageous with respect to grain refinement, because the high fraction of soluble niobium shifts ferrite formation to lower temperatures and acts in the same way as accelerated cooling (15).
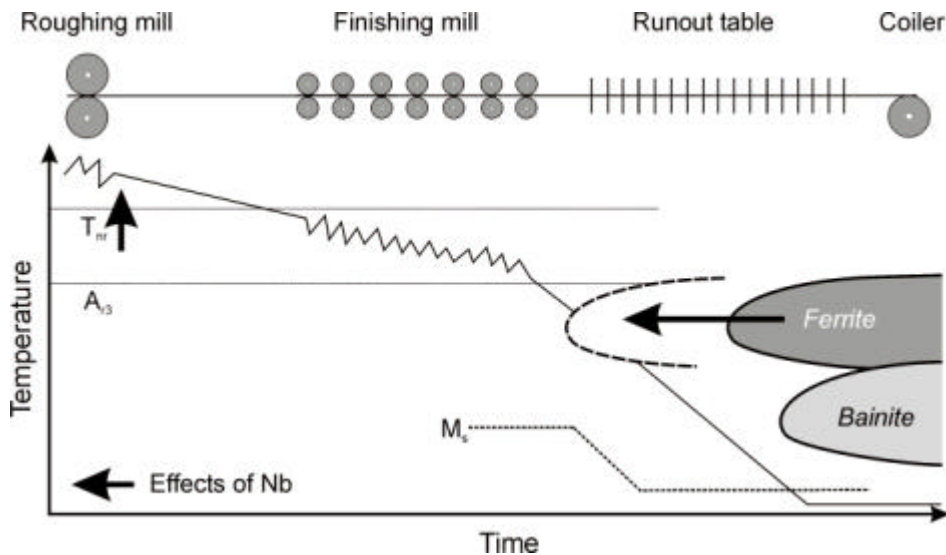

Figure 12: Effect of niobium on hot strip processing of DP steel.

Higher carbon levels are used for TRIP steels which is due to the necessity of producing a highly enriched austenite with a carbon content above 0.8 mass%. Apart from carbon, manganese and silicon play an important role to control the transformation behaviour and to stabilize the retained austenite. Silicon not only promotes a possible formation of some proeutectoid ferrite during cooling after intercritical annealing, but as an element which is not soluble in cementite silicon also prevents or at least retards carbide precipitation during bainite formation and allows the carbon to diffuse into the retained austenite. Manganese as an austenite stabilizing element lowers the transformation temperature of the austenite and in this way helps to avoid martensite formation during cooling to room temperature. Manganese also enhances the possibility of carbon enrichment, by increasing carbon solubility in austenite and extends the applicable range of cooling rates as it delays pearlite formation. Moreover, manganese, as well as silicon, increases the strength of the material by solid solution hardening. Too high manganese levels must be avoided, because manganese lowers the carbon activity in austenite and in this way promotes carbide formation. The inhibiting effect on ferrite formation must also be taken into account, since it minimizes a possible carbon enrichment during cooling. Concerning the addition of silicon and manganese, the relative amounts must be adjusted in order to control the phase distribution and volume fraction as well as the carbon content of the austenite in particular (16).

As a higher silicon content can be responsible for a poor surface quality of hot rolled steel and a poor coatability of cold rolled steel, a partial or complete substitution of silicon by other alloying elements has been the focus of attention of several research activities. Another element, which is not soluble in cementite and therefore has to diffuse from the interface of the carbide particle before the particle can continue to grow, is aluminium, which also promotes the generation of ferrite. But as the effect of aluminium is weaker and as the solid solution hardening potential of aluminium is lower, it is often used in combination with higher carbon

contents or in combination with phosphorus, which reduces the kinetics of cementite precipitation on the one hand and is a very effective solid solution hardening element on the other hand. Further disadvantages connected with the use of aluminum are the rise in the transformation temperature, which extends the intercritical region and makes full austenitization impossible, and the rise of the martensite start temperature (17, 18).

Phosphorus can be used when the addition of silicon and aluminium has to be limited. Low additions of less than 0.1 mass% prove to be sufficient to prevent the precipitation of iron carbides and to provide a clear enhancement of strength. The beneficial effects of phosphorus on the formation and retention of retained austenite can only be accomplished in combination together with silicon or aluminium. If the addition of phosphorus exceeds a certain amount, phosphorus might segregate to the grain boundaries and cause a deterioration in the ductility (19).
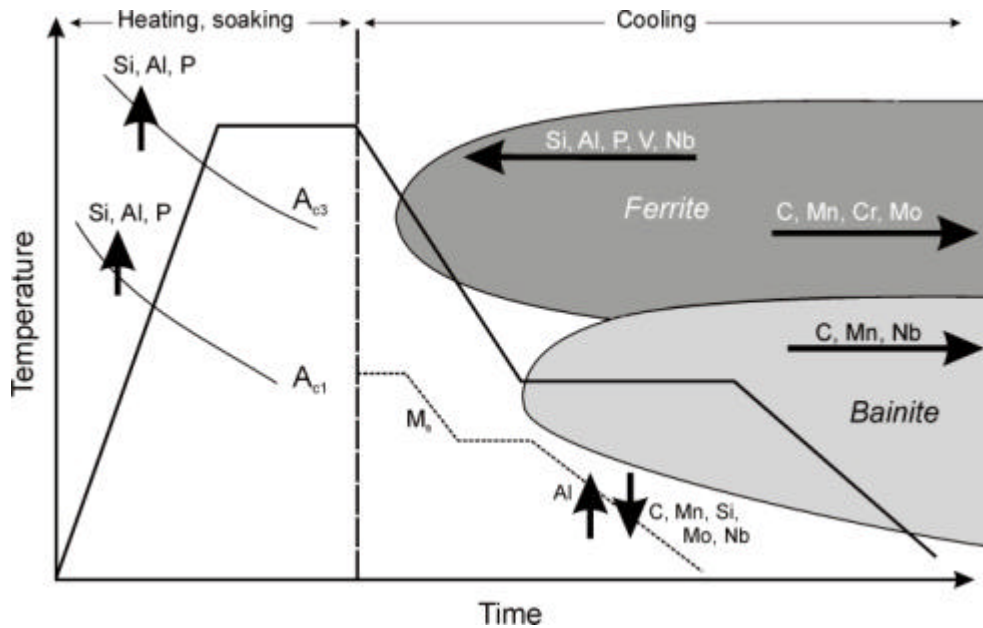

Figure13: Effect of alloying elements on transformation behavior during continuous annealing of TRIP steels.

The well known effects of micro alloying elements such as niobium on grain boundary motion and element migration also influence the TRIP effect. The different phase transformations during hot rolling, intercritical annealing, cooling, isothermal holding in the temperature range of bainite formation, and straining are modified thus resulting in superior mechanical behaviour. Direct effects of niobium on the retention of austenite at room temperature and indirect effects arising from the improved carbon enrichment of the austenite must be highlighted. Vanadium might also be used to control the transformation behaviour of TRIP steels (20, 21).

Additions of molybdenum to niobium-bearing steels are reported to bring about a further improvement in the combination of strength and ductility and to constitute another possibility of lowering the silicon level required. Molybdenum exerts an important solute drag effect and delays the transformation of austenite to ferrite and to pearlite strongly. Although molybdenum lowers the activity of carbon in austenite and promotes carbide formation from the thermodynamic point of view, the contrary effect is observed in practice. Due to the strong solute drag effect the carbide precipitation is actually retarded in the presence of molybdenum. Moreover, molybdenum raises strength due to solid solution hardening. Figure 13 depicts how

the alloying elements affect transformation behaviour during each single step of processing (22).

Transformation Behaviour and Microstructure Development

The presence of some metastable austenite after the final heat treatment is considered the main reason for the superior properties of TRIP steels. The development of the microstructure and retained austenite in particular is therefore illustrated in Figure 14. During intercritical annealing a microstructure of almost identical fractions of ferrite and austenite is planned. The carbon atoms are almost completely incorporated in the austenite. Figure 14-1 shows an austenite grain which is surrounded by several ferrite grains and which originates probably from the pearlitic or the bainitic region of the microstructure prior to heating. During cooling part of the austenite might transform to ferrite, again connected with carbon enrichment of the adjacent austenite. The ferrite is likely to nucleate at the grain boundaries and especially the triple points of proeutectoid ferrite and austenite (Figure 14-2).
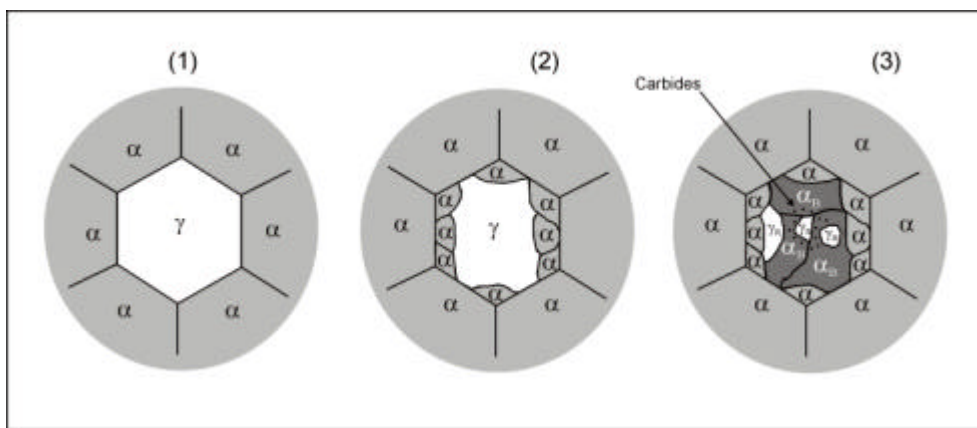

Figure 14: Microstructure development during heat treatment of TRIP steel; (1) two phases during intercritical soaking, (2) ferrite formation in austenite during first cooling, (3) bainite formation and retained austenite enrichment during second soaking.

In the course of the second isothermal holding step, part of the austenite is transformed to bainite and the remaining austenite is further enriched in carbon, so that the martensite start temperature is shifted below room temperature. The retained austenite is predominantly found adjacent to recently formed ferrite and bainite and within the bainite (Figure 14-3). Some very small grains might remain austenitic in spite of a moderate carbon enrichment, which derives from the adverse nucleation situation due to the small size. The final microstructure contains 5 to 15% of retained austenite, which is partly or completely transformed to martensite during forming, as the driving force is increased by a mechanical contribution. The beneficial effects of this transformation depend heavily on the resistance of the austenite to strain induced martensite formation which is determined by the driving force available for the austenite to martensite transformation and the activation energy needed to transform the highly metastable austenite to the more stable martensite. The lack of activation energy, that must be compensated by the mechanically provided driving force in order to induce martensite formation, is often referred to as stability of the austenite, though the austenite does not represent a stable phase in terms of equilibrium. If the so called stability is either too low or too high, the transformation does not progress steadily during forming and its beneficial effect is limited. The austenite stability is mainly determined by the carbon content. Additional

importance is attributed to other alloying elements, which also affect the chemical driving force available and the activation energy necessary for the martensitic transformation, the austenite grain size, which defines the probability of finding a nucleation site in the austenite particle, the strength of the matrix, which determines the mechanical driving force contribution to the total driving force, and the stress state applied, as the volume change due to the transformation is encouraged or impeded by the tri-axial nature of the stress (23, 24).

The phase distribution and especially the distribution of the elements is different from that predicted from equilibrium. Calculations incorporating thermodynamic as well as kinetic data clarify the processes during intercritical annealing. It can be concluded from Figure 15a that the volume fraction of austenite in a typical TRIP steel increases as a function of intercritical annealing temperature, until finally full austenitization is reached. In consequence the carbon content of the austenite declines. After 90 seconds, however, the actual amount of austenite is lower than predicted by thermodynamics. Figure 15b reveals that neither at 780°C nor at 850 °C equilibrium can be attained within 90 seconds. The concentration peak at the interface of austenite and ferrite is due to carbon diffusion from ferrite to austenite and the high carbon content remote from the interface stems from the presence of pearlite, which was chosen as a component part of the initial microstructure for the calculations. Since substitutional elements diffuse much more slowly than interstitial elements, only little alteration of the distribution of these elements between the phases is expected (25).
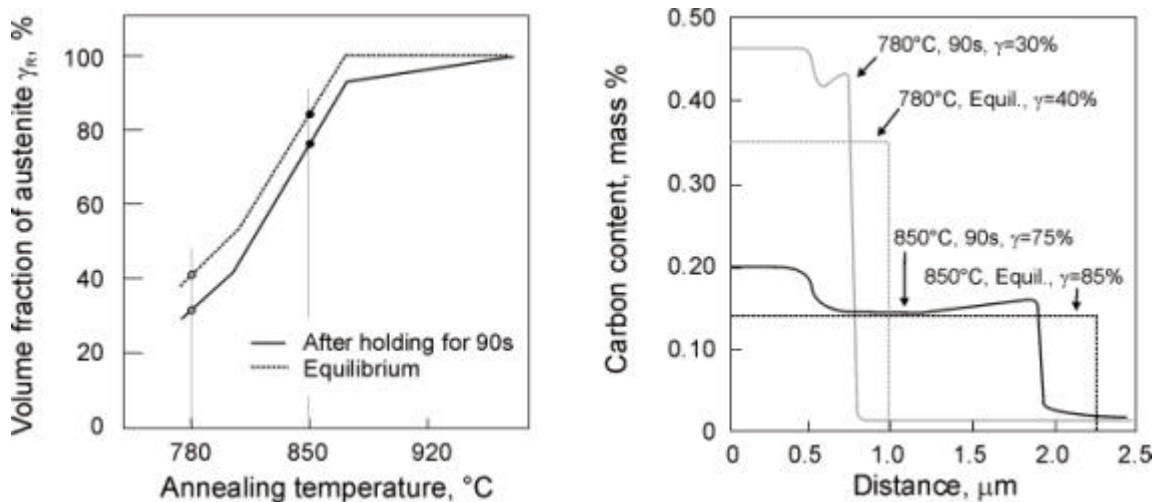

Figure 15: Calculated austenite formation and carbon enrichment during intercritical annealing of a TRIP steel, Composition in mass%: C: 0.14, Mn: 1.35, Si: 1.3; carbon distribution and austenite fraction at the temperatures marked in the left figure is displayed in the right figure.

In the temperature range between 450°C and 350°C, in which the second annealing step is performed, only carbon possesses a mobility worth mentioning. Holding temperature and time applied are vitally important so as to create a sufficient amount of retained austenite with an adequate stability. This combination is the prerequisite for the favorable mechanical properties of TRIP steels. Figure 16 outlines the development of the phase distribution at a constant temperature as a function of holding time. As the ferrite fraction does not change, it is excluded from this schematic graph. Too short holding times lead to the formation of little bainite and prevent the degree of carbon enrichment which is indispensable to avoid martensite formation during cooling to room temperature. As bainite formation progresses, the amount of retained austenite increases steadily, before it reaches a maximum. Even higher amounts of the hard bainitic phase and consequently lower amounts of retained austenite are directly correlated

with higher strength values but noticeably poorer elongation values. Very long holding times must be avoided, as carbide formation might start, which withdraws carbon from the austenite and reduces the volume fraction of retained austenite. If the holding temperature is raised, shorter holding times, which are of special interest to industrial usage, provide the desired microstructure and the favourable austenite characteristics (26).
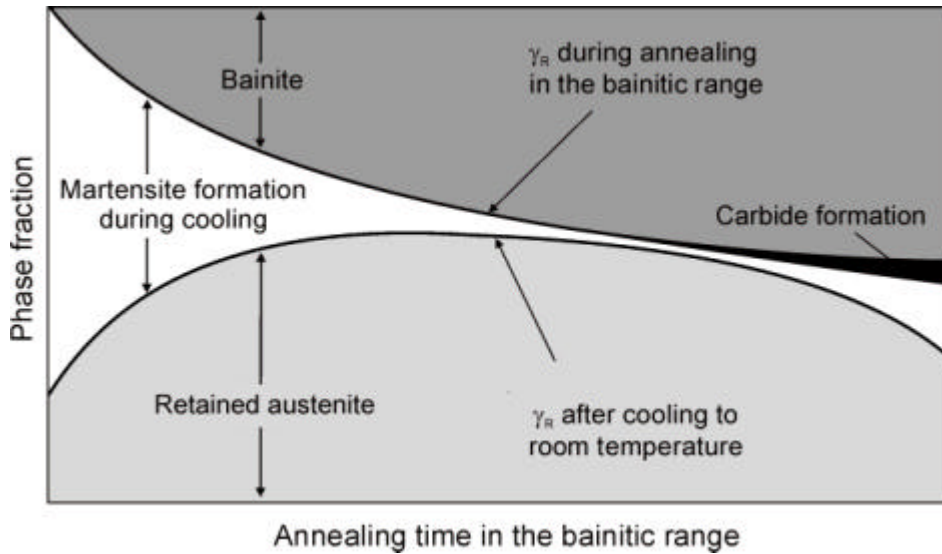

Figure 16: Development of the volume fractions of the microstructural components during second soaking and final cooling of TRIP steel.

Niobium changes the recrystallization and transformation behaviour during hot rolling, intercritical annealing, bainitic transformation and possibly during intermediate cooling as well. The effect of niobium on the martensite start temperature after intercritical annealing for three minutes at 800°C is shown in Figure 17 and reveals the strong inhibition of martensite formation that can be achieved by niobium. This is particularly true, if a low coiling temperature of about 500°C is applied after hot rolling. The low coiling temperature leads to a decrease of the M_s–temperature of the niobium free TRIP steel as well, which has been suggested to be explained by the homogeneous and very fine structure of the bainite and the homogeneous distribution of carbon. The microstructure of ferrite and pearlite resulting from the use of a coiling temperature of approximately 700°C, however, contains coarse cementite that possibly does not dissolve completely or at least entails an inhomogeneous carbon distribution during intercritical annealing.

A similarly positive effect is ascribed to the addition of niobium which accounts for the generation of a fine grained ferrite due to the retardation of recrystallization and grain growth and due to a ferrite formation that takes place at lower temperatures, but possesses a distinctly higher driving force and a higher nucleation rate. However, the effect of niobium is restricted, when coiling is performed at 700°C, as precipitates become coarser and thus less effective concerning the suppression of grain growth during the slow cooling. If coiling is carried out at round about 500°C, only 50% of the niobium is precipitated after hot rolling, whereas 50% is determined to be chemically soluble which is the result of chemical isolation analysis and includes niobium in solid solution as well as very small precipitates with a diameter of less than 20 nm. As the measurements indicate that the precipitated fraction of niobium after intercritical annealing is irrespective of the hot rolling conditions and amounts to more than 90%, extensive precipitation must take place during intercritical annealing in the case of the low coiling temperature. The favourable nucleation conditions surely result in the precipitation of very small particles. These precipitates not only control the grain size, encourage carbon diffusion

and ensure a highly homogenized microstructure, but apparently also impair martensite nucleation considerably. The result is another decrease of the M_s–temperature by almost 100°C.
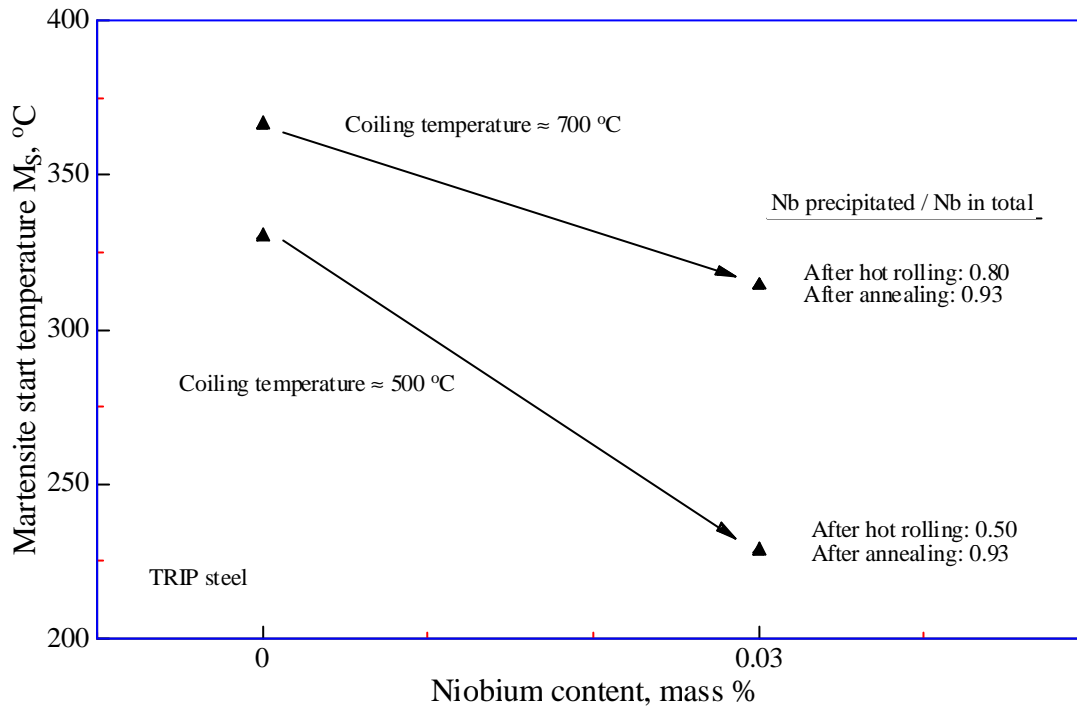

Figure 17: Effect of niobium content and coiling temperature on the M_s-temperature of TRIP steels, Composition in mass %: C: 0.2, Mn: 1.4, Si: 0.5, Al: 0.75, P: 0.04.

Figure 18 demonstrates that bainite formation is also delayed by niobium, which might also be attributed to the blocking of the nucleation sites by the very fine dispersed carbonitrides. This hypothesis is supported by the fact that the retarding effect is much more pronounced at a lower transformation temperature of 350°C, at which the displacive character of the bainite formation should prevail. At a temperature of 450°C the velocity of bainite formation is hardly affected by the niobium content. Another possible mechanism accounting for reduced bainite formation is ferrite formation during cooling due to the fine grained microstructure and the enhanced nucleation and growth conditions. A similar effect might be accomplished by using a lower cooling rate. Dilatometric experiments confirm that a transformation actually takes place during cooling. A ferrite formation would enable further carbon enrichment of the austenite and lower the bainite start temperature. This assumption conforms to the fact that aside from the velocity of bainite formation the total amount is reduced (27).
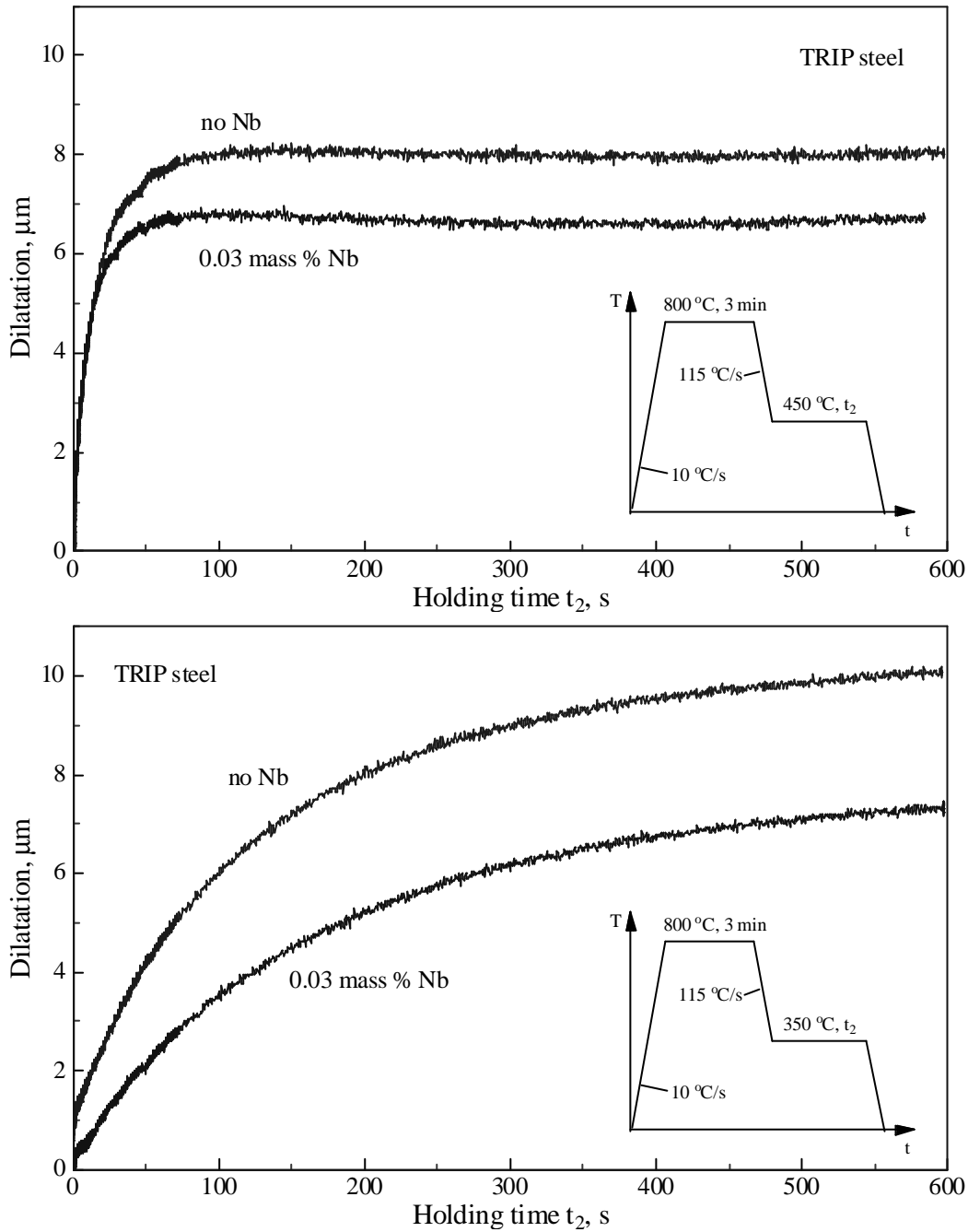
Figure 18: Effect of niobium on isothermal bainite formation of TRIP steels at 450°C and 350°C, Composition in mass%: C: 0.2, Mn: 1.4, Si: 0.5, Al: 0.75, P: 0.04.

Figure 19 illustrates the grain refinement in multi phase steels due to the addition of niobium and reveals an even stronger effect than in normalized steels, which originates from the lower reheating temperature applied to TRIP steels and especially dual phase steels. An additional contribution of the multi phase microstructure to the small ferrite grain size is also conceivable. Finally the positive effect of niobium on the volume fraction of retained austenite is displayed in Figure 20. The high austenite contents are a result of the combination of the different mechanisms described, including grain refinement, carbon enrichment, and martensite nucleation inhibition (28).
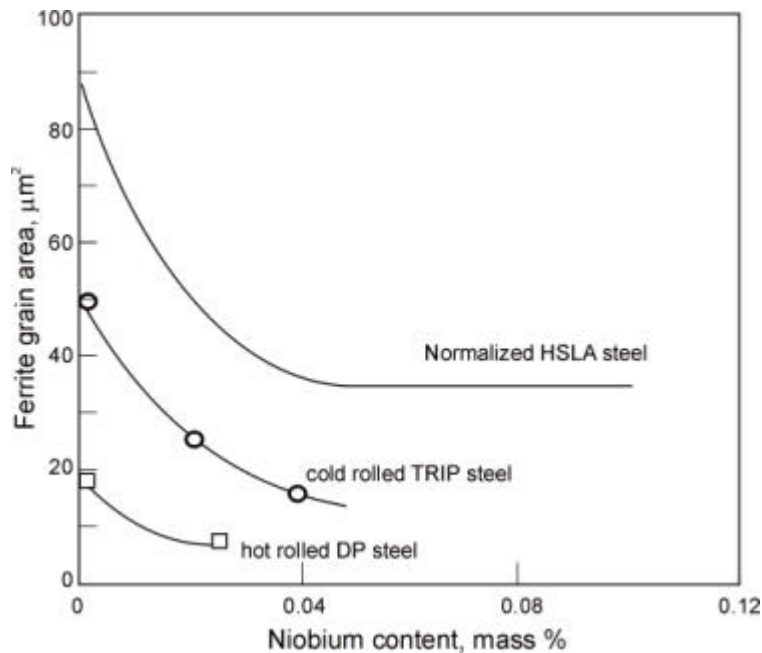
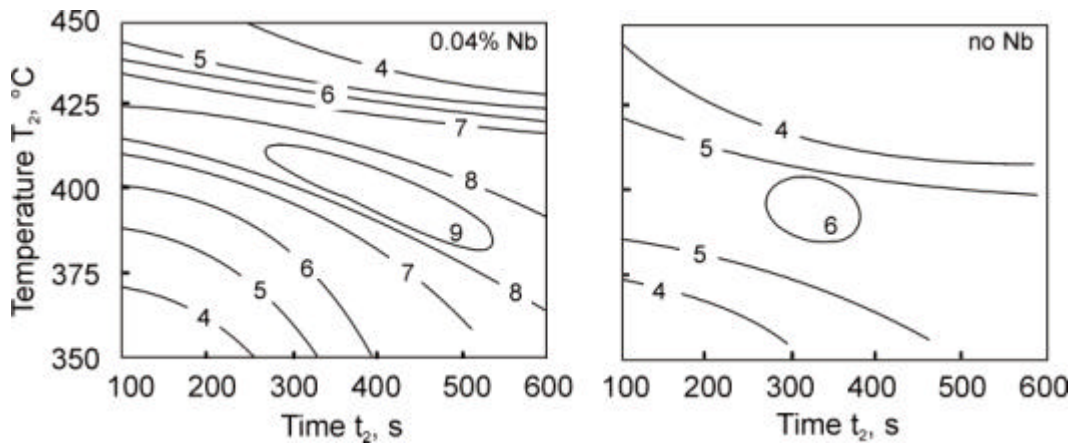Figure 19: Effect of niobium on the ferrite grain size of different steels.

Figure 20: Effect of niobium on the volume fraction of γ_R of TRIP steels for different process parameters, Composition in mass%: C: 0.17, Mn: 1.4, Si: 1.5.

Mechanical Properties

Each elevation of the strength level usually causes a deterioration in formability as can be observed in Figure 21, which summarizes the developments of the past years in the field of cold-rolled sheet steel. Only by introducing TRIP steels was this conflict was at least partly defused and a steel grade became available that offers enhanced strength with tensile strength values up to 1000 MPa in combination with high elongation values surpassing those of micro alloyed steels by far and almost reaching those of mild deep drawing steels. These findings are supported by strain hardening characteristics which allow a closer look at the material behaviour during forming. Regardless of the deformation stage the strain hardening values of TRIP steel are much higher than those of single phase ferritic steel (Figure 22), which is a consequence of the coexistence of several phases with different strength levels and different formability characteristics, the extensive generation of dislocations within the martensite and the adjacent regions of ferrite and austenite, and the strain induced transformation of retained

austenite to martensite, accounting for the suppression of local necking. The strain hardening of TRIP steels is unequalled by other low alloyed steels. For high strains it can be three times as high as that of soft deep drawing steel DC 04 and even distinctly higher than that of dual-phase steel H 300 X. The n-value shown in Figure 23 testifies to the excellent strain hardening behaviour as well.
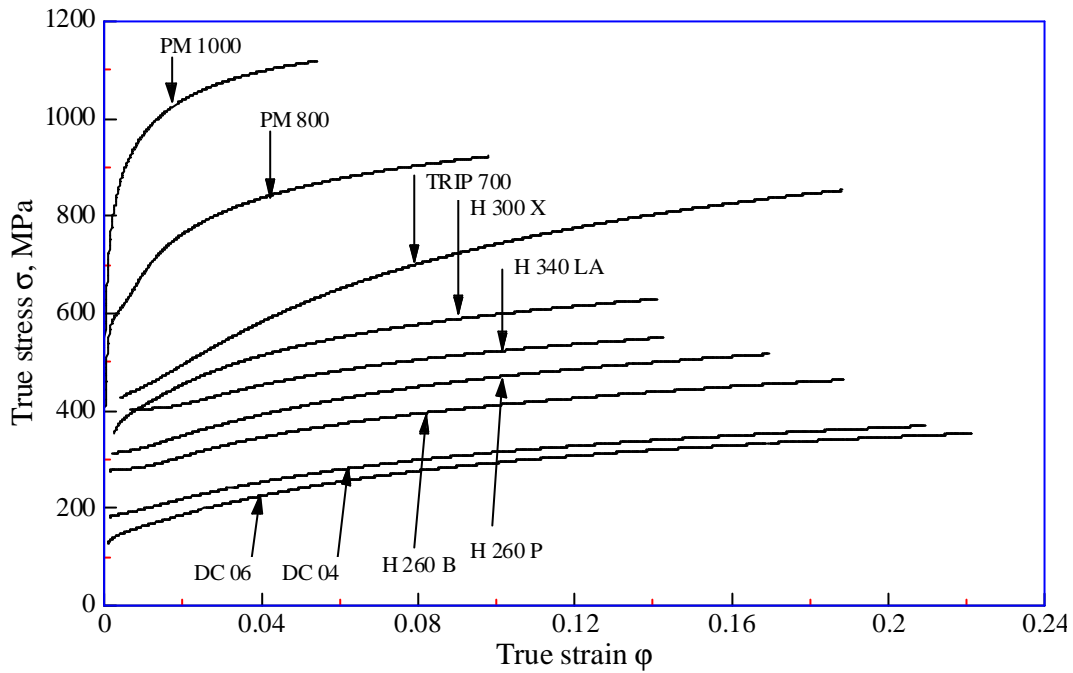

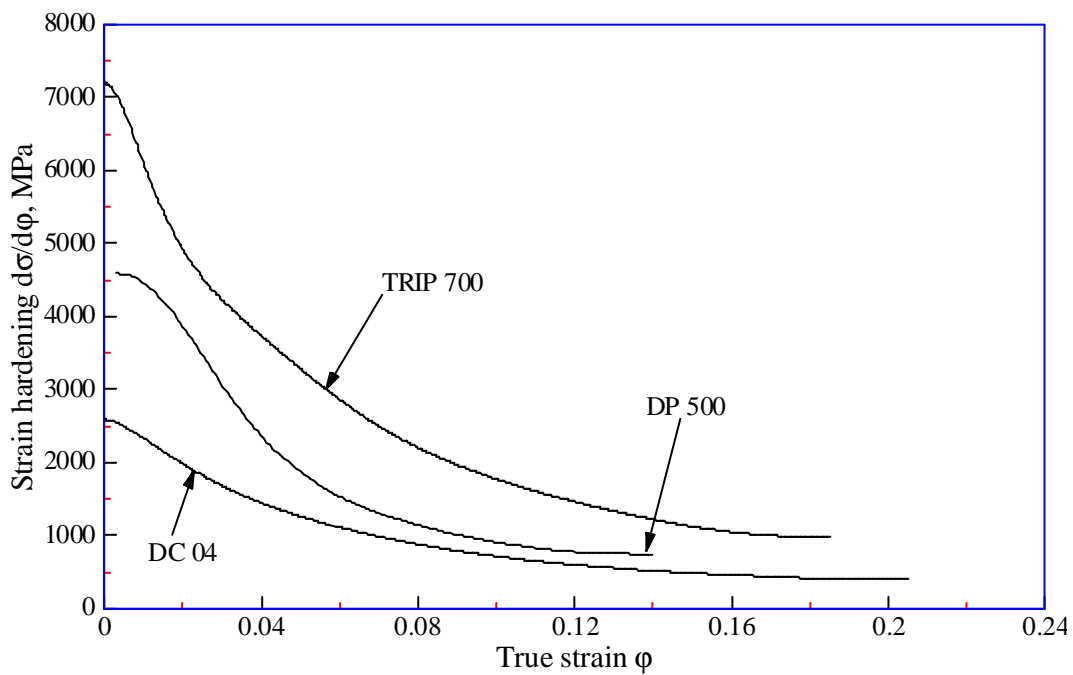Figure 21: Flow curves of different cold rolled steels.

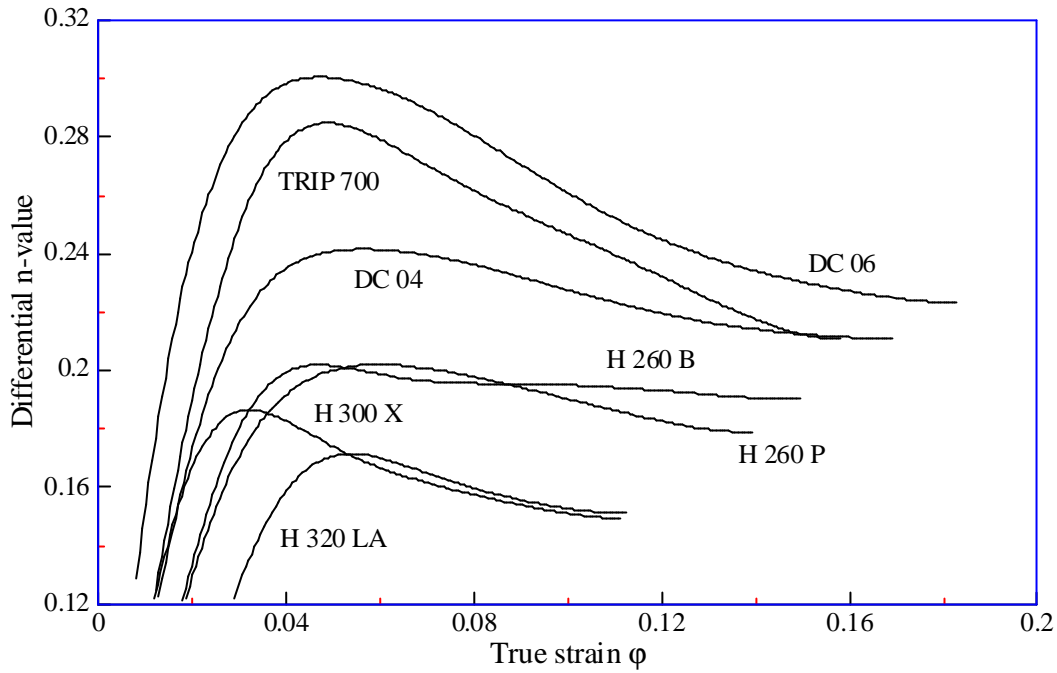Figure 22: Strain hardening of different cold rolled high strength steels.

Figure 23: Differential n-value of different cold-rolled steels.

An addition of niobium to dual phase steels raises yield and tensile strength considerably as illustrated in Figure 24. An increase of up to 200 MPa can be achieved without a great loss in ductility being based on the strong grain refinement. Medium coiling temperatures produce a certain amount of bainite and reduce the stress concentrations existing at the grain boundaries between ferrite and martensite thus improving ductility (29, 30).
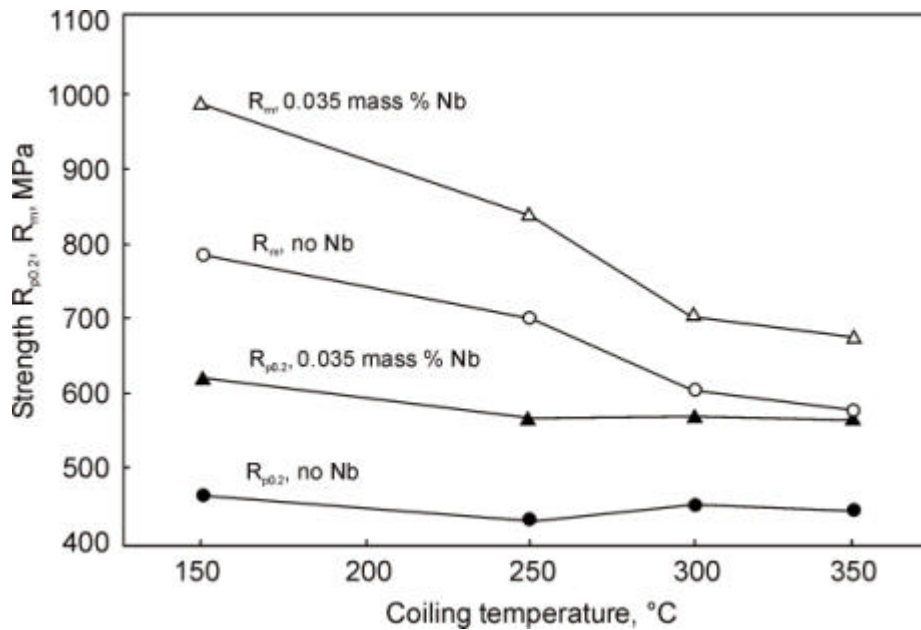

Figure 24: Effect of niobium content and coiling temperature on the mechanical properties of hot rolled DP steel, contents in mass%: C: 0.12, Mn: 1.2, Si: 0.1, Cr: 0.8.

As the strain induced formation of martensite is subject to a distinct temperature dependence, the mechanical properties of TRIP steels respond sensitively to a change in temperature. By

varying the chemical composition and the processing parameters the microstructure can therefore be adapted to the respective forming or usage temperature in order to obtain the required mechanical properties. The mechanical properties of dual phase steels exhibit a similar temperature dependence as those of single-phase ferritic steels (Figure 25).
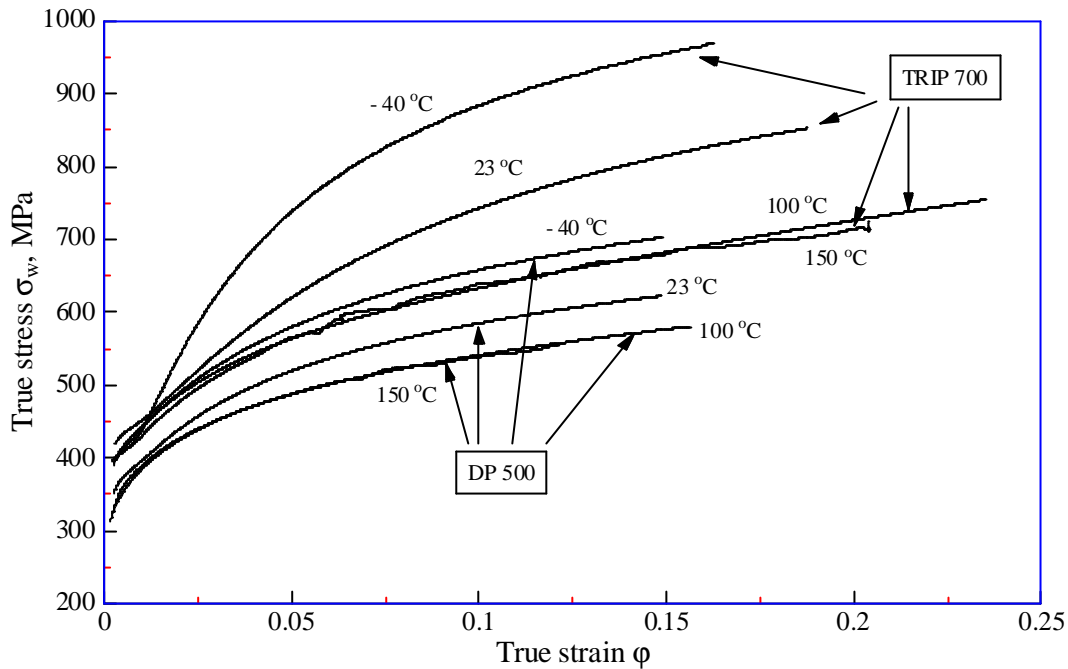

Figure 25: Effect of temperature on the flow curves of DP and TRIP steels.

The strength values of TRIP steels decrease as temperature increases, as shown in Figure 26. Obviously the tensile strength is much more affected by the temperature variation. The behaviour of the yield strength as a function of temperature does not greatly differ from that of conventional steels and can be explained by the effects of thermally activated plastic flow, whilst the transformation induced plasticity has a significant impact on the tensile strength. The increase of the latter at lower temperatures is distinctly more pronounced than the well known temperature dependence of steels not showing any kind of phase transformation during deformation and can therefore not derive from thermally activated processes. Actually the stress- and strain induced formation of martensite is responsible for the rise in tensile strength observed at low temperatures, whereas the yield strength is hardly influenced by the TRIP effect, as the stress is not high enough to cause the formation of a considerable amount of martensite (31).

The elongation values possess a maximum in the temperature range between 50°C and 120°C (Figure 26), that can be interpreted as a consequence of the optimum stability of the retained austenite in relation to the stress state applied. The martensite preferably appears in the regions subjected to the maximum stress, which usually leads to the occurrence of necking in this area. The formation of the hard martensitic phase, however, is an adequate device by which local necking can be suppressed and thus strain hardening can be encouraged. A higher temperature results in a higher stability of the austenite, which impedes or even avoids martensite formation. In consequence the amount of retained austenite transforming into martensite will be reduced due to an increase in temperature. Lower temperatures, however, destabilize the austenite by increasing the chemical driving force, so that the transformation already takes place during the beginning of the deformation, which leads to higher strength values, but does not contribute to an improvement in the formability. Superior formability requires a transformation of austenite,

that progresses steadily, as stress and strain grow. Some very small and highly enriched particles do not transform during straining and participate in the plastic flow of the matrix, which is reported to improve uniform elongation and hole expandability (32).
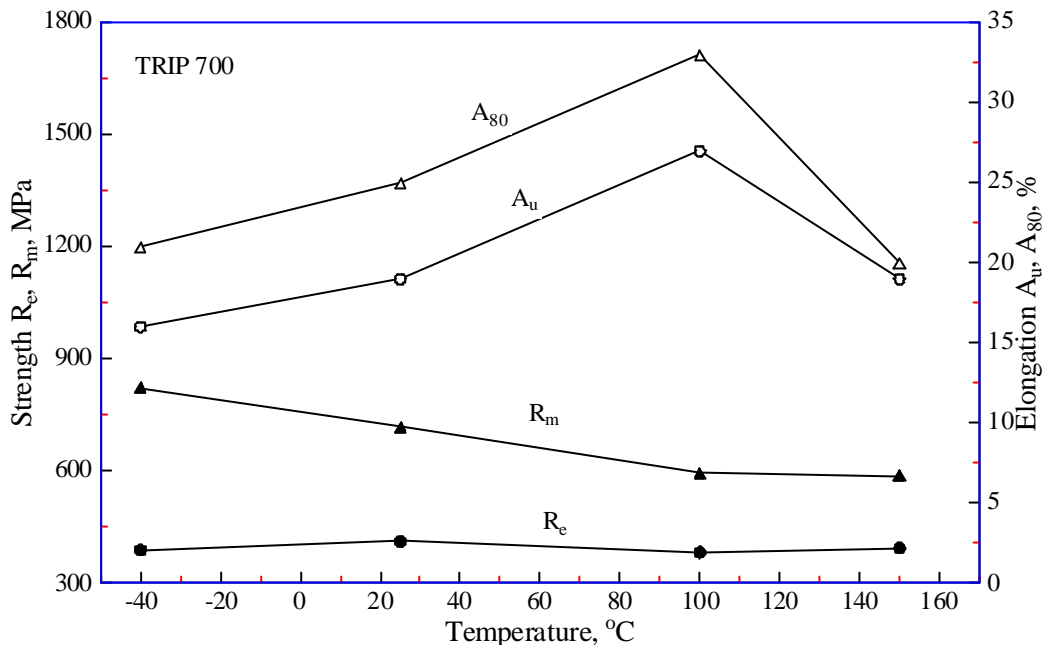

Figure 26: Effect of temperature on the mechanical properties of TRIP steel TRIP 700.

Although the existence of retained austenite in the multi phase microstructure of TRIP steels is accounted the source of the favourable mechanical behavior, no clear relation between the volume fraction of retained austenite and the elongation values for example can be detected. In contrast to the possible assumption that an increase in austenite content is generally followed by higher elongation values, Figure 27 reveals that the maximum elongation values do not coincide with the highest volume fraction of austenite. Nevertheless a minimum amount of retained austenite should be present in order to guarantee TRIP behavior and enhanced strength and formability. Rather poor formability in spite of a high austenite content is related to an insufficient carbon enrichment during annealing and thus a low stability of the austenite, which is then transformed to martensite in the very early stage of plastic deformation or even in the elastic region, so that dual-phase behaviour can be observed. As only a limited amount of carbon is available, a compromise between the austenite content and the carbon content in the austenite must be found. Therefore an increase in austenite content might sometimes even prove detrimental to the formability. The minimum amount of retained austenite necessary to ensure high elongation values rises, if the strength level of the material gets higher, as is true for steels containing niobium. As niobium encourages the formation of high austenite contents, the addition of niobium leads to higher strength and almost equal elongation values. Figure 28 shows the gradual transformation of austenite during straining and the effect of temperature. A shorter annealing time in the bainitic range would result in a higher initial austenite content and a sharper drop as a function of strain (33).
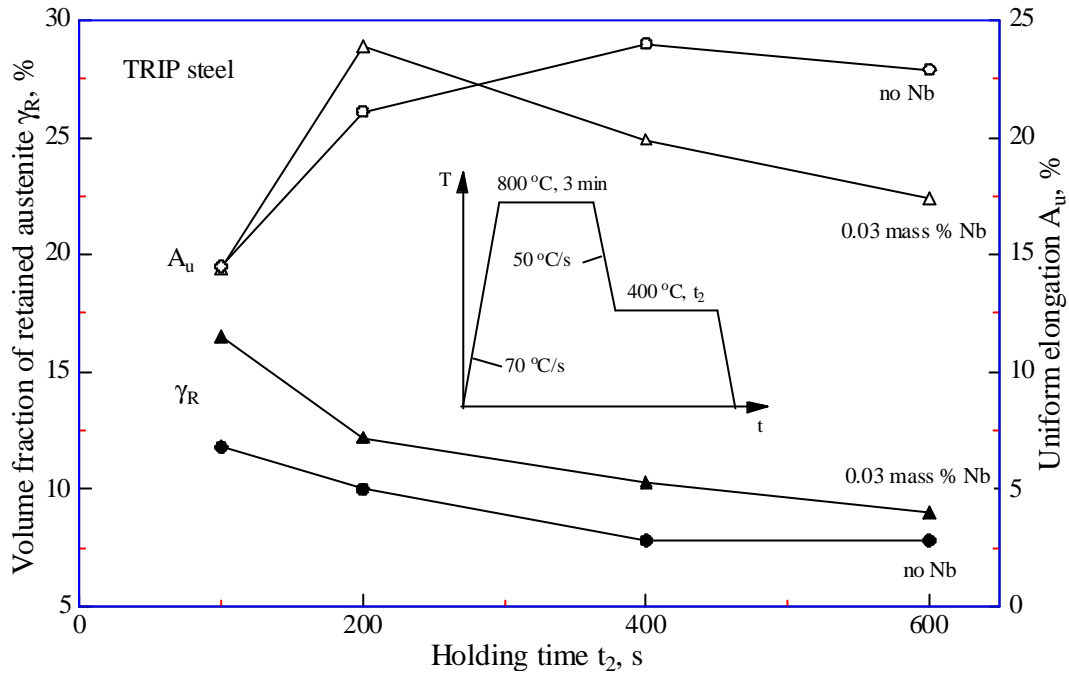
Figure 27: Relation of γ_R-constituent and uniform elongation of TRIP steels for different isothermal bainitic holding times, Composition in mass%: C: 0.2, Mn: 1.4, Si: 0.5, Al: 0.75, P: 0.04.
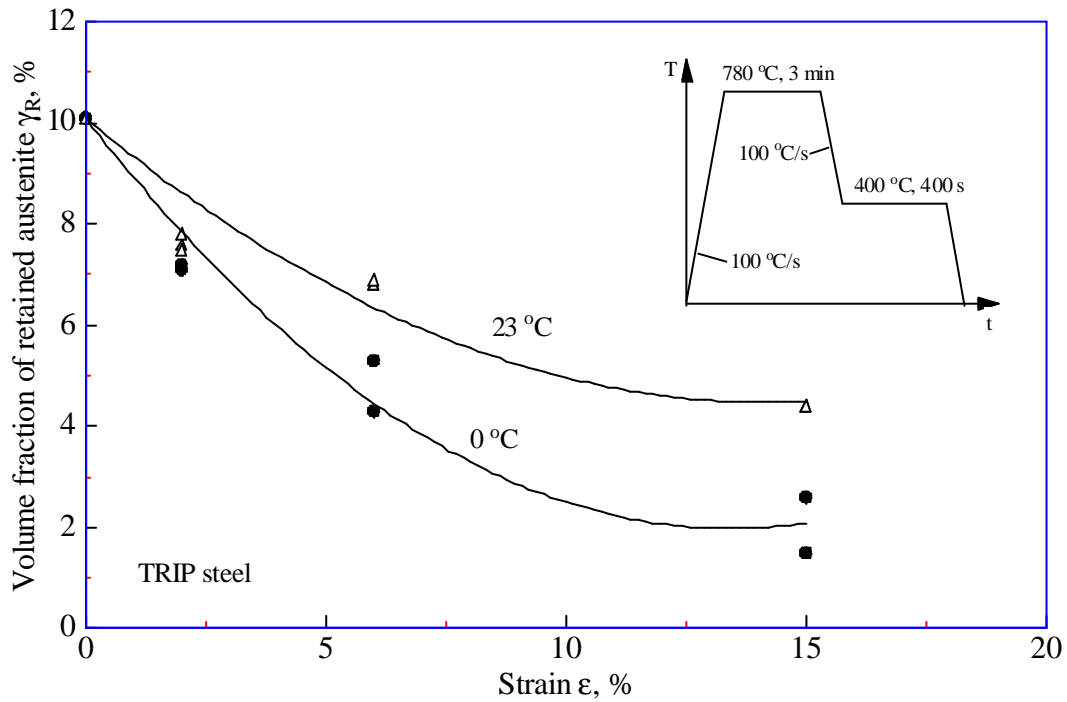

Figure 28: Change in γ_R-constituent of a TRIP steel during straining at different temperatures, Composition in mass %: C: 0.17, Mn: 1.45, Si: 1.1.

The mechanical activation of the martensite transformation is primarily based on the impact of shear stress which is not only the key to plastic deformation by dislocation slipping, but also promotes the nucleation and growth of martensite particles. Moreover, the hydrostatic part of the stress tensor interacts with the volume increase due to the transformation of face centered cubic austenite to body centered cubic martensite, so that hydrostatic tension encourages

martensite formation, whilst hydrostatic compression inhibits martensite formation. Figure 29 reveals how the austenite is stabilized and the martensite start temperature decreases as a consequence of increasing hydrostatic compression. The decline of the martensite start temperature is more pronounced for higher carbon contents (34).
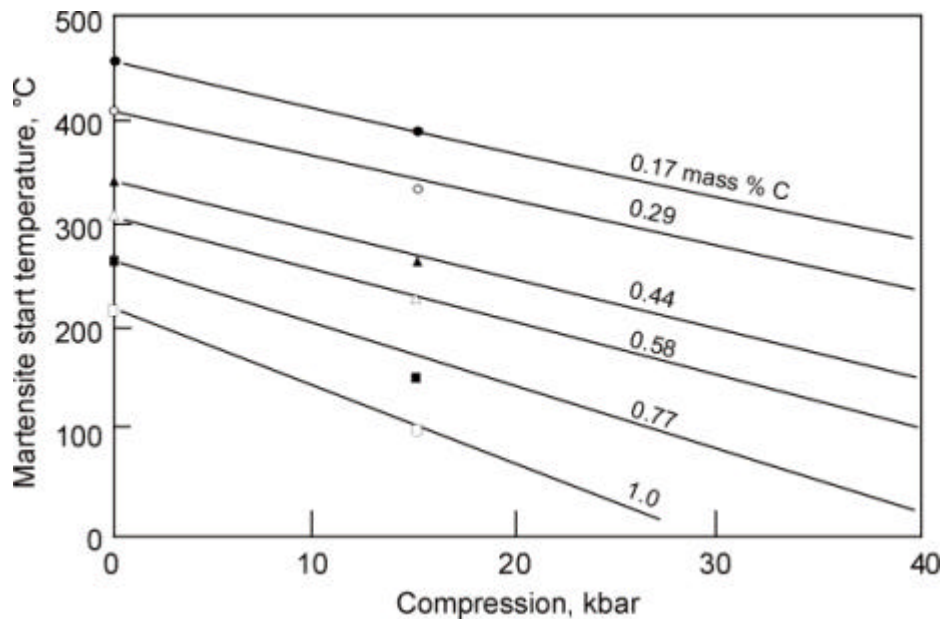

Figure 29: Effect of hydrostatic compression on M_s-temperature for different carbon contents.

Conclusions

The mechanical properties of so called multi phase steels are superior to those of conventional cold formable high strength steels, but the production of these steels requires sophisticated process routes and the observance of narrow parameter windows.

Niobium exerts a remarkable influence on grain size development, transformation behavior, carbon enrichment of the austenite, and martensite nucleation. This renders process control often much easier, allows less demanding process routes and leads to a further improvement of the favourable mechanical properties of multi phase steels.

Material development in future should pay attention to the effect of temperature and stress state on the transformation of metastable austenite in order to adapt the microstructure to the respective forming temperature and stress state thus optimizing strength and formability of TRIP steels.

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