

Advanced metallurgical concepts for DP steels with improved formability and damage resistance

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ABSTRACT

Dual phase steel typically available in the tensile strength range of 450 to 980 MPa is widely used in today's car body manufacturing. Its characteristics of high n-value and good elongation (A80) are connoted with good press formability. However, practical experience has shown unexpected failure in forming operations where tight bending, stretch flanging or hole expansion are predominant. The inhomogeneous microstructure of soft ferrite and hard martensite in combination with local straining is inherent to these problems. However forming behavior can be much improved by microstructural refinement and homogenization of the phases as well as modification of hardness contrasts. Modified alloy concepts in combination with adapted processing have been identified as an appropriate approach to achieve this optimization. Niobium and molybdenum are key alloying elements in this approach. The paper explains the fundamentals, demonstrates respective production concepts and presents examples of industrial implementation of such optimized DP steels.

INTRODUCTION

Dual phase (DP) steels consist of a ferrite-martensite microstructure and have been developed to combine very high strength with increased elongation. Thus DP steel allows forming rather complex part geometries that are not possible to manufacture with conventional HSLA steel of the same strength level. The volume fraction of hard martensite islands determines the strength of DP steel whereas the ductile ferrite matrix provides good formability. The characteristic of as-delivered DP steel is a relatively low yield strength and high initial work hardening resulting in a high n-value. The high n-value provides good protection against local thinning under the forming conditions of drawing and stretching.

Although DP steel allows forming of complex shapes, practical experience repeatedly revealed unexpected failure even when manufacturing quite simple geometries [1, 2]. Shear fracture has regularly been observed in die bending operations (Fig. 1). Another common problem is sheared edge splitting during flanging operations (Fig. 2). Roll profiling leads to splitting in the outer bend when the bend radius is too tight causing severe downtime of production lines. All these situations have in common that strain is highly localized. Thus, severe dislocation pile-up is occurring in individual ferrite grains leading to delamination at the ferrite-martensite phase boundary or to cracking of martensite islands as demonstrated in Figure 3. Such in-situ induced micro damage can grow into a propagating crack. The larger the size of an initial damage site the smaller is the critical stress required for crack propagation. A crack typically propagates along the ferrite-martensite interface.

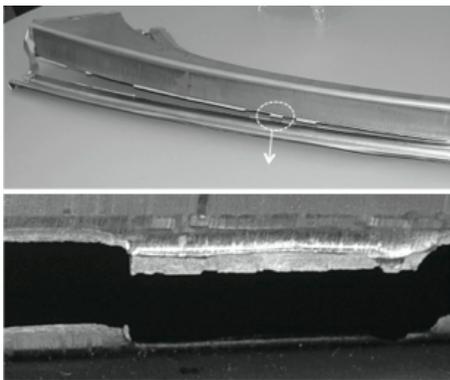


Figure 1. Shear fracture during die bending [1].

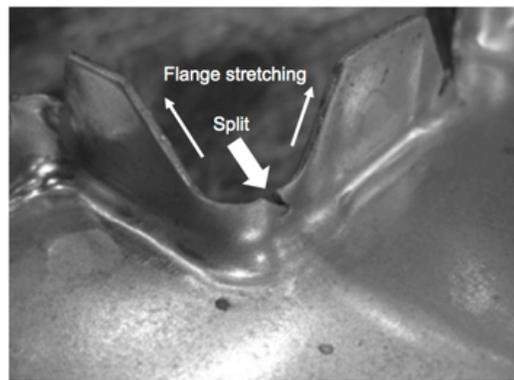


Figure 2. Sheared edge fracture during stretch flanging [2].

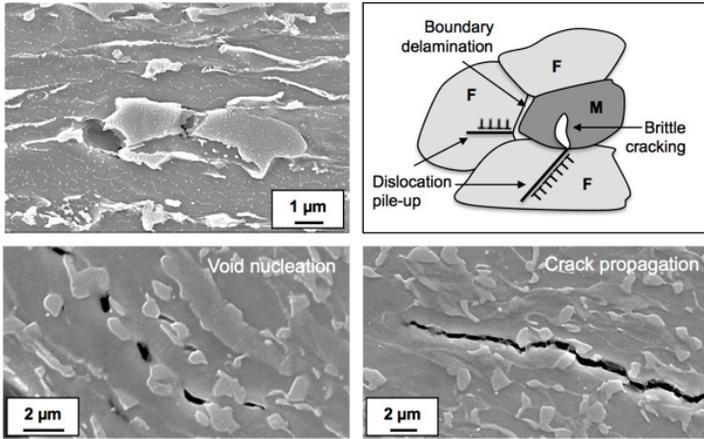


Figure 3. Features and mechanisms of damage initiation and propagation in DP steel.

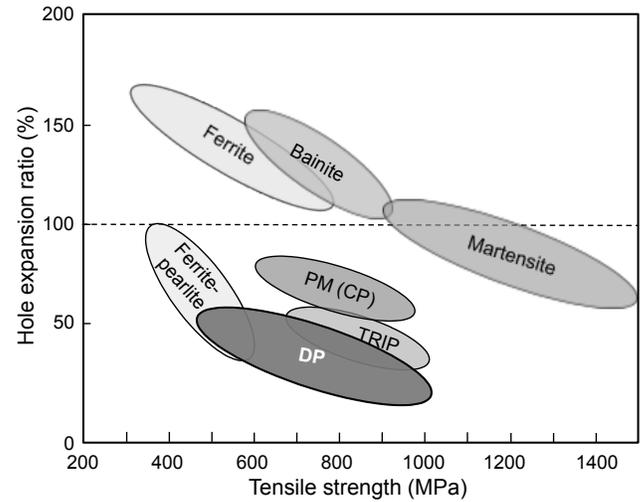


Figure 4. Influence of tensile strength and microstructure on the hole expansion ratio.

It is apparent that the mixture of hard and soft phases in the microstructure of DP steel causes the discrepancy of having good draw and stretch formability versus rather bad bendability and hole expansion ratio. A low hole expansion ratio is a feature of all microstructures containing hard and soft phases (Fig. 4). Although this is an inherent handicap of multiphase steels that cannot be completely removed, there are approaches of improving hole expansion ratio and bendability to an acceptable level. The challenge will be generally not to compromise the drawing and stretching behavior. On the other hand, for particular manufacturing processes where drawing or stretching demands are on a low level (e.g. roll-profiling), compromises could be made to specifically optimize for high bendability or high hole expansion ratio. The present paper details microstructure-property relationships that allow laying the foundation for optimized DP steels. Subsequently alloying and respective processing concepts are being worked out as a guideline for industrial implementation.

OPTIMIZING DUAL PHASE MICROSTRUCTURE

Microstructural refinement

Dislocation pile-up causing micro damage under the condition of localized straining cannot be principally avoided in DP steel. Nevertheless countermeasures exist to reduce the criticality of this phenomenon. Most important in this respect is refining the microstructure, i.e. reducing the size of the ferrite and martensite grains. Consequently the size of initial damage is reduced raising the critical stress for crack propagation to higher level according to the principles of Griffith's theory. Furthermore it is important to homogenize the microstructure, i.e. to avoid the formation of pearlite bands in the prior hot strip microstructure. Both microstructural characteristics, a refined grain size as well as homogeneous phase distribution can be achieved by Nb microalloying in combination with appropriate hot rolling conditions leading to austenite pancaking. Application of this concept leads to a refined ferritic-pearlitic microstructure of the as-hot rolled strip. Fast cooling after the last rolling pass further refines the microstructure by delaying the austenite-to-ferrite transformation. Solute niobium and molybdenum provoke this transformation delay. Bainitic coiling can be an option to obtain a particularly fine hot strip microstructure. Although the enhanced yield strength of bainitic hot strip may form a concern with regard to cold rolling forces it still remains feasible for thicker cold rolled gages.

The final cold rolled annealed DP microstructure is in a significant way determined by the prior hot strip microstructure [4] although some grain coarsening usually occurs during annealing [3]. A secondary effect of the refined hot strip microstructure is an accelerated nucleation rate of newly formed phases during the intercritical annealing (ICA) cycle since the new phases preferably nucleate on the grain boundary. In the refined hot strip microstructure the total grain boundary area is significantly enhanced. Therefore the statistical probability of separating martensite islands by soft ferrite islands capable of arresting a crack is enhanced. Most of the niobium remaining in solid solution after hot rolling is likely to precipitate during the intercritical annealing cycle. Molybdenum on the contrary stays in solid solution after annealing due to its very good solubility at the typical carbon contents utilized in DP steels.

The effectiveness of microstructural refinement by Nb microalloying is exemplary demonstrated for DP780 in Fig. 5 [5]. Conventional and grain refined DP steels are benchmarked in an instrumented 3-point bending test. It is obvious that the grain-refined steel supports a higher bending force at simultaneously increased bending angle. The standard steel fails at a bending angle of around 90 degrees so that the production of typically U-shaped profiles is critical. Small scatter in the materials bendability may lead to significant production losses. The grain-refined steel on the contrary offers sufficient margin for the forming process providing clearly improved process robustness. Similarly positive results were obtained with DP980 material where microstructural refinement allows

tighter bending radii during roll profiling (Fig. 6). In this case a profile with specified inner width w has a more compact outer dimension W consuming less space in the assembly and reducing total weight of the structure.

Besides better bendability and hole expansion behavior, microstructural refinement also results in an increase of yield and tensile strength as shown in Fig. 7 [6]. Thus, the volume fraction of martensite phase can be reduced in grain-refined steel while the minimum specified tensile strength is still reached. The enhanced amount of ferrite results in increased elongation, lower yield strength and thus higher n -value, improving the steel's overall forming characteristics. The generic tensile curve (Fig. 8) highlights all influencing effects on microstructural scale. From this it can be concluded that microstructural refinement is indisputably beneficial for all relevant mechanical properties of DP steel.

Low-carbon alloy concepts

A very significant influence on the hole expansion ratio of DP steel originates from the difference in hardness between the soft ferrite phase and the hard martensite phase. Hosoya et al. have demonstrated this effect by performing different degrees of low temperature tempering treatments to quenched DP steel in the strength range of 800 to 1000 MPa. Their results suggest a linear relationship between the hardness difference and the hole expansion ratio (Fig. 9). Thereby the volume share of martensite in the steel appears to be of less significant impact, provided that its distribution is homogeneous.

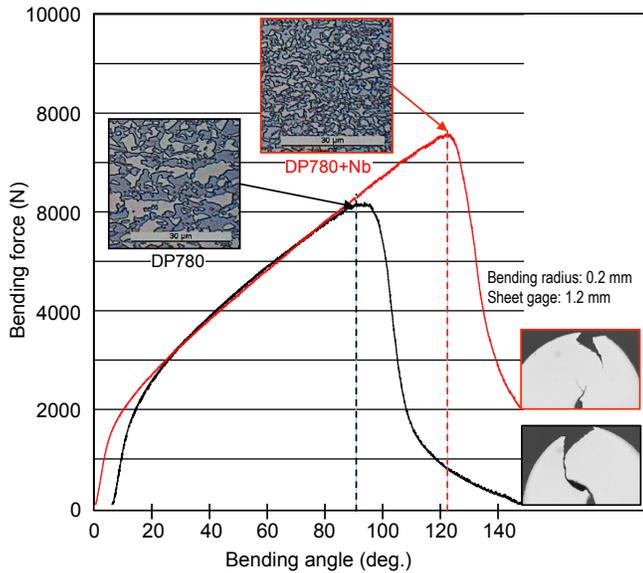


Figure 5. DP780 steel under 3-point bending conditions before and after microstructural refinement.

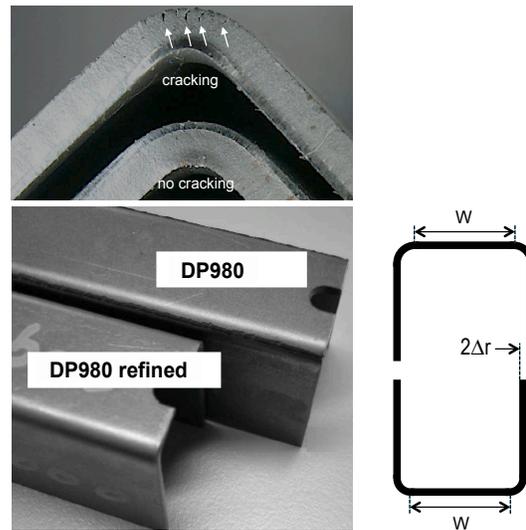


Figure 6. Effect of microstructural refinement with regard to corner radius during roll profiling of DP980 steel.

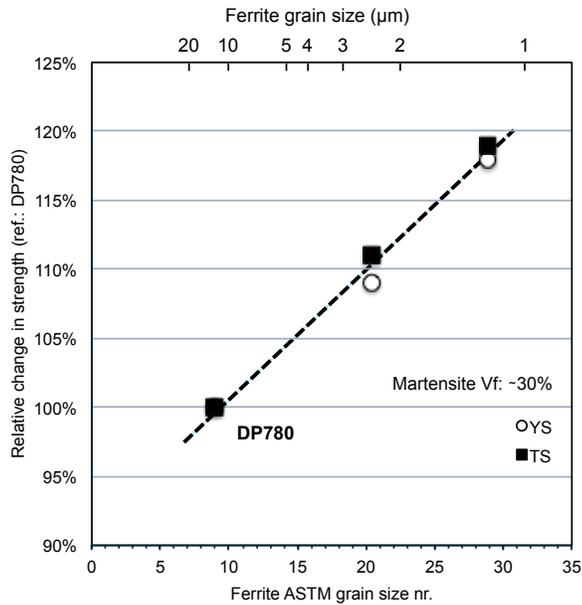
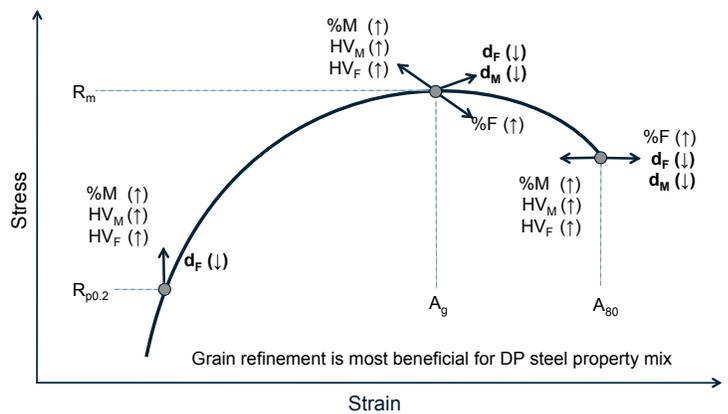


Figure 7. Effect of microstructural refinement on yield and tensile strength of DP steel [6].



%F, %M: volume fraction of Ferrite, Martensite
 HV_F, HV_M: hardness of Ferrite, Martensite phase constituent
 d_F, d_M: grain size of Ferrite, Martensite
 (↑): increasing, (↓): decreasing

Figure 8: Microstructural influences on key characteristics in the tensile curve of DP steel.

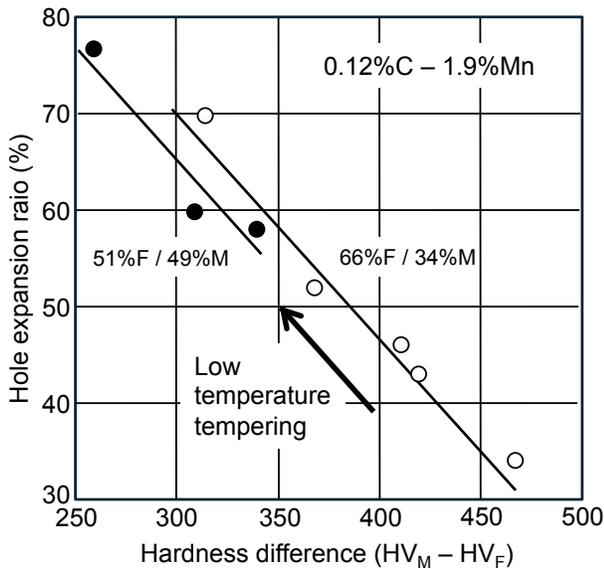


Figure 9. Relationship between M-F hardness difference and hole expansion ratio for DP steels between 800 and 1000 MPa [10].

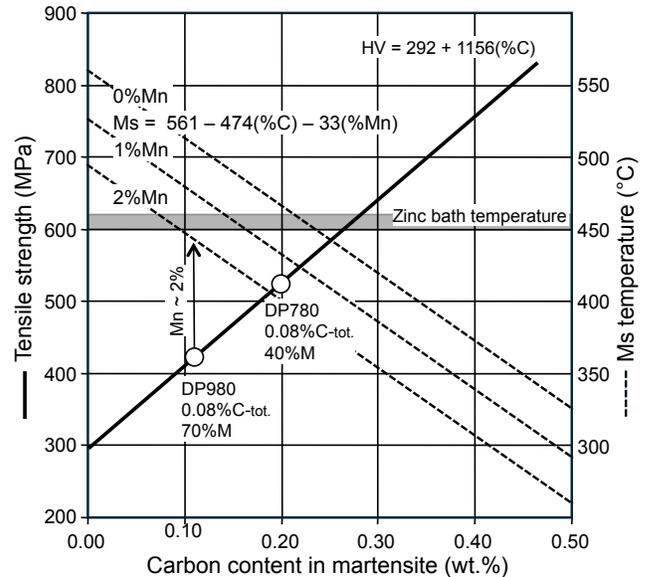


Figure 10. Tensile strength and martensite start temperature as a function of the carbon (manganese) content.

A tempering treatment is not really practical in the production sequence of DP steel. Therefore other ways of reducing the hardness difference should be explored. One can make ferrite harder by either solid solution hardening (increased Si, Mn, P), precipitation strengthening (Nb microalloying) or by incomplete recrystallization. Grain refinement will also contribute to increase the ferrite hardness. On the other hand martensite can be softened by reducing its carbon content instead of performing a tempering treatment. The hardness (strength) of martensite is known to depend directly on its carbon content as shown in Figure 10. Traditional DP steel typically has an over-peritectic carbon content (i.e. 0.15-0.16%). For a martensite share of either 30% (DP780) or 50% (DP980) the carbon content after repartitioning would be around 0.5% or 0.3%, respectively. Such martensite has a hardness of between 600 and 800 HV. Taking a ferrite hardness of 130 to 150 HV, the hardness difference will be over 450 HV resulting in a very low hole expansion ratio according to Figure 9.

In a sub-peritectic alloy concept, the carbon content is below 0.1%. With a total carbon content of 0.08% a larger share of consequently softer martensite is necessary to achieve the specified strength as indicated by Figure 10. With lower carbon content also the martensite start temperature increases. This can become a critical issue when producing low-carbon DP steel via a hot dip galvanizing line. If martensite is formed at temperatures above 460°C, it will be in situ tempered when passing the strip through the zinc bath resulting in a severe loss of strength. Increasing the Mn level to 2% or slightly above can be a remedy to prevent premature martensite formation. In low-carbon DP steel hardness differences between ferrite and martensite can be expected to be below 350 HV providing the conditions for a much improved hole expansion behavior. It was recently demonstrated that DP980 based on a low-carbon chemistry shows a reduced sensitivity to notch damage [11]. The low-carbon DP980 could tolerate a higher local fracture strain despite its lower elongation as compared to the higher carbon versions. An additional benefit of the low-carbon chemistry is related to weldability since the risk of cold cracking in the heat-affected zone is being reduced [7].

PROCESSING ASPECTS OF OPTIMIZED DP STEELS

Effects caused by niobium microalloying

Niobium microalloying is a key aspect in producing microstructure-optimized DP steel as outlined above. At this point some important process implications will be discussed. The grain refinement intended to improve the mechanical properties of cold rolled annealed DP steel is inherited from the prior hot strip [4]. Therefore the initial microstructure before cold rolling has to be grain refined. Grain refinement increases the flow stress of the as hot-rolled steel resulting in higher rolling forces during cold rolling. Monitored production trials with DP780 have indicated that the flow stress of Nb-microalloyed grain-refined strip is consistently 100 to 150 MPa higher for the same degree of cold reduction.

Depending on the amount of Nb added as well as the rolling and coiling conditions in the hot strip mill, not all Nb will be precipitated in the finished hot strip. Niobium remaining in solid solution after hot rolling has the potential of precipitating during the intercritical annealing cycle. Specific experiments in an annealing simulator have indicated that the precipitation of solute Nb is already occurring during the heating up phase. Niobium precipitates exert a recrystallization delay of the cold deformed ferrite. Experience indicates that the top annealing temperature should be raised by about 20°C above that of Nb-free alloy to reach thorough recrystallization (Fig. 11). Solubility products [12] suggest that typical low-carbon DP steel chemistries and appropriate annealing conditions could allow a

maximum solute Nb content of 120 ppm at intercritical annealing temperature. Niobium in solid solution has a retarding effect on bainite formation as it lowers the activity of carbon [13]. Therefore metastable austenite of reduced carbon content is stabilized by solute Nb [14], which is particularly important when producing low-carbon DP steel with an increased amount of martensite. The retarded recrystallization also preserves dislocation networks that act as nucleation sites for austenite. Hence austenite formation should be accelerated in Nb alloyed DP steel. On the other hand, the grain-refined microstructure of Nb microalloyed strip additionally provides an increased grain boundary area as nucleation site for austenite when annealing in the intercritical temperature range. Measurements have indeed confirmed that at a given intercritical annealing temperature the amount of austenite in the Nb added alloy is higher compared to the Nb-free base alloy [14] (Fig. 12). During the soaking phase, carbon partitioning is accelerated and more homogeneous in the finer grained microstructure of the Nb alloyed strip due to the shorter diffusion distances in the smaller grains. By slow cooling to quenching temperature new ferrite is being nucleated reducing the amount of existing austenite. The refined microstructure of Nb microalloyed exhibits a quicker kinetics of new ferrite formation (Fig. 13) [15]. The accelerated transformation kinetics makes the grain-refined alloy also more robust against process deviations. For instance a variation in line speed and hence the actual quenching temperature is affecting the yield strength of the Nb microalloyed DP steel less severely (Fig. 14).

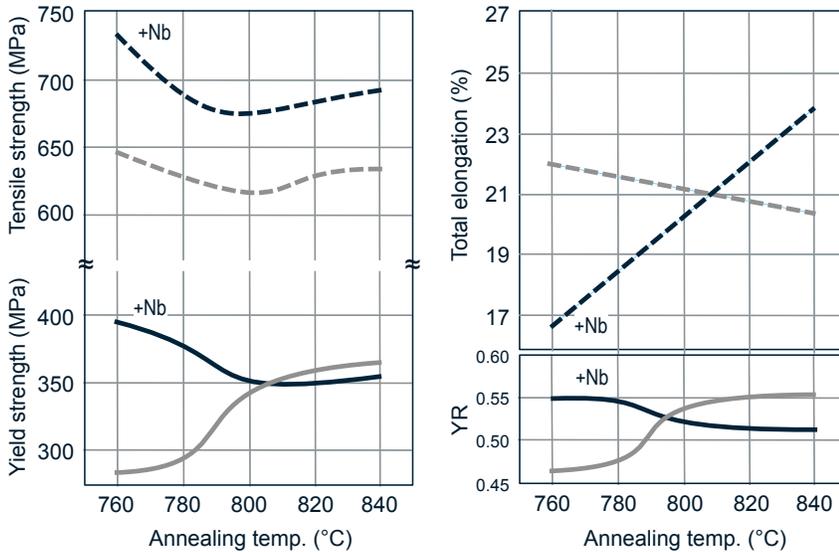


Figure 11. Influence of Nb microalloying on the intercritical annealing behavior (QT = 750°C) of low-carbon DP steel.

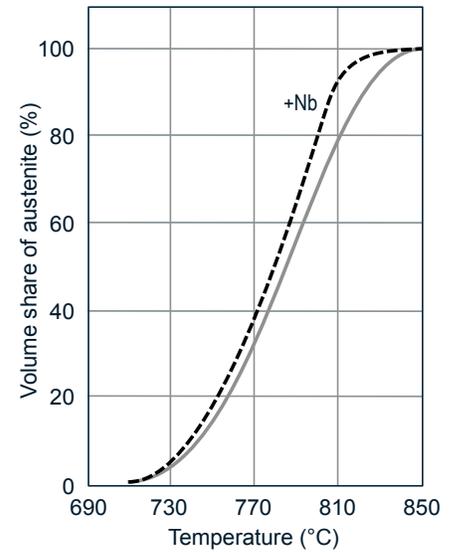


Figure 12. Effect of Nb on austenite formation during reheating [14].

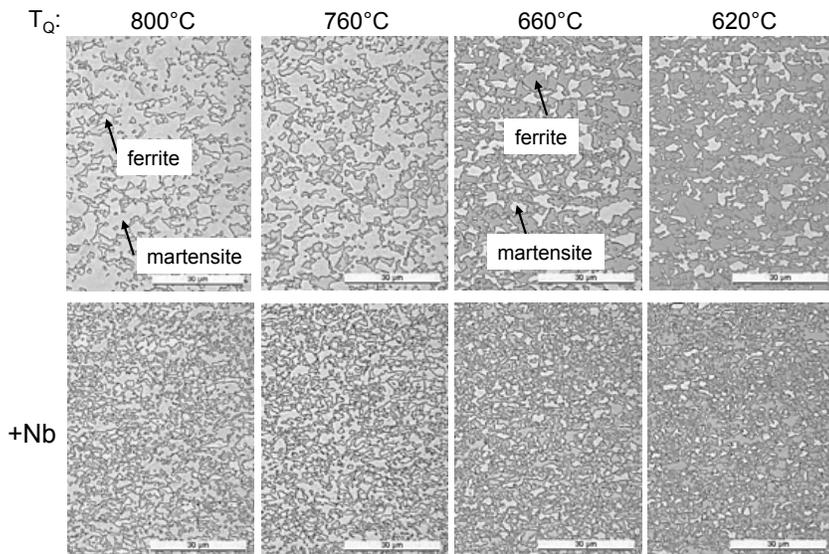


Figure 13. Ferrite and martensite phase distribution after intercritical annealing and slow cooling to quenching temperature T_Q [15].

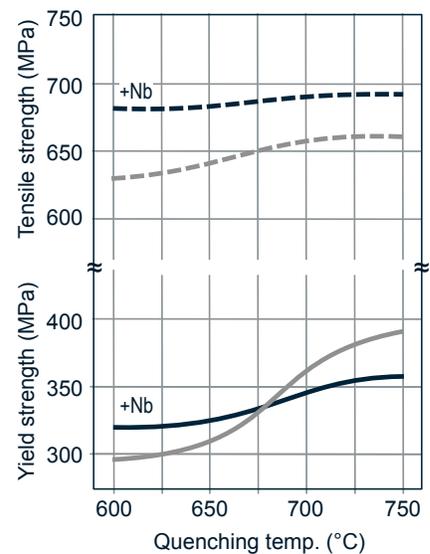


Figure 14. Effect of quenching temperature on strength after intercritical annealing at 840°C.

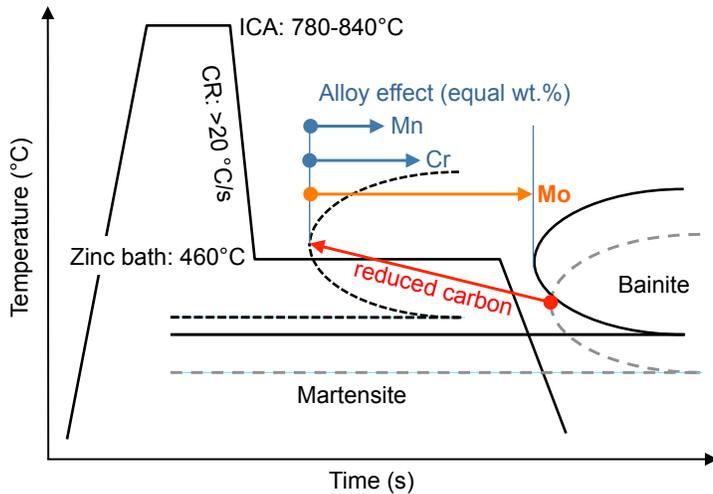


Figure 15. Alloying effects on bainite formation in low-carbon DP steel during a hot dip galvanizing cycle.

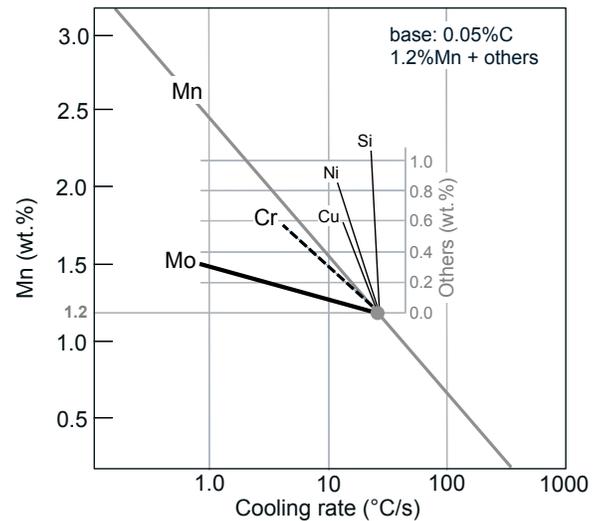


Figure 16. Influence of alloying elements on critical cooling rate for martensite formation in low-C steel.

A consequence of the enhanced ferrite amount is the further enrichment of carbon in the remaining austenite share. Thus the hardenability of the carbon-enriched and smaller austenite grain is improved. With regard to mechanical properties Nb microalloyed DP steel has less but stronger martensite as a second phase when subjected to a given annealing cycle compared to the Nb-free alloy. Systematic variation of annealing and quenching temperature revealed that Nb microalloyed DP steel always has the higher tensile strength (Figs. 11 and 14). Its yield strength decreases at increased annealing temperature after overcoming the recrystallization delay. In combination this leads to better formability being characterized by a lower yield-to-tensile ratio and a higher total elongation. The variation of the quenching temperature shows a significantly smaller impact in the Nb microalloyed variant since the transformation kinetics in the finer grained material is quicker. The work hardening rate is directly proportional to the square root of martensite share and inversely proportional to the square root of martensite island size [16, 17]. Hence, it is particularly interesting reducing the martensite island size with regard to improving the stretch forming behavior as was demonstrated based on experimentally obtained forming limit diagrams (FLD) [3].

Effects caused by molybdenum alloying

Producing low-carbon DP steel of increased strength level such as DP780 or DP980 is challenging, particularly in the hot-dip galvanizing line. The level of carbon enrichment in austenite is relatively low in such an alloy concept (Fig. 10) and hence the risk of partial bainite formation during the holding period at around zinc bath temperature (Fig. 15) increases. The presence of bainite in the microstructure reduces the tensile strength and raises the yield strength, thus deteriorating the key characteristics of low yield-to-tensile ratio and high strain hardening coefficient (n -value) of standard DP steel. Avoiding bainite formation during the quasi-isothermal holding phase requires sufficient shifting of the bainite phase field towards longer time. This can be practically achieved by increasing the amount of alloying elements such as Mn, Cr, and Mo as shown in Figure 15. Generally, Mo has the largest effect on the critical cooling rate followed by those of Cr, Mn, Cu, Ni and Si (Fig. 16) [8]. Molybdenum's effect is approximately 2.6 and 1.3 times larger than that of Mn and Cr, respectively [9]. The solid solution hardening potential of Mo is between that of Mn (higher) and Cr (lower), respectively. Accordingly, partial replacement of Mn by Cr is applicable particularly when low yield strength (and low yield-to-tensile ratio) is targeted. These characteristics are mostly desirable for DP grades in the lower strength range (DP450 – DP590), which are regularly subjected to drawing operations in press forming. On the other hand Mo-based alloy concepts, often in combination with Cr-alloying, are most appropriate for producing low-carbon DP grades of the upper strength range (DP780 – DP980).

One fundamental metallurgical effect of Mo is to remarkably decrease the mobility of the ferrite/austenite interface, which is based on solute drag [18]. Comparing on an equal atomic concentration, the magnitude of this effect caused by Mo is similar to the one of Nb and much stronger than that of Mn and Cr (Fig. 17). Accordingly Mo can retard the growth of austenite during the intercritical annealing phase. As mentioned before, the solubility of Nb is very limited so that in practical applications its retarding effect must be rather small. On the other hand, during slow cooling after annealing Mo similarly impedes the growth of ferrite [18]. Both contributions promote formation of a finer grained microstructure.

Another fundamental effect refers to the diffusivity of carbon [13], [19]. Addition of carbide-forming elements such as Mo (Fig. 18) and Cr decreases the carbon diffusivity by attracting C atoms. Niobium also causes a strong reduction of carbon diffusivity (Fig. 18). However, the amount of Nb available for this effect is limited due to its low solubility. Silicon on the contrary being a ferrite stabilizer increases the carbon diffusivity in austenite.

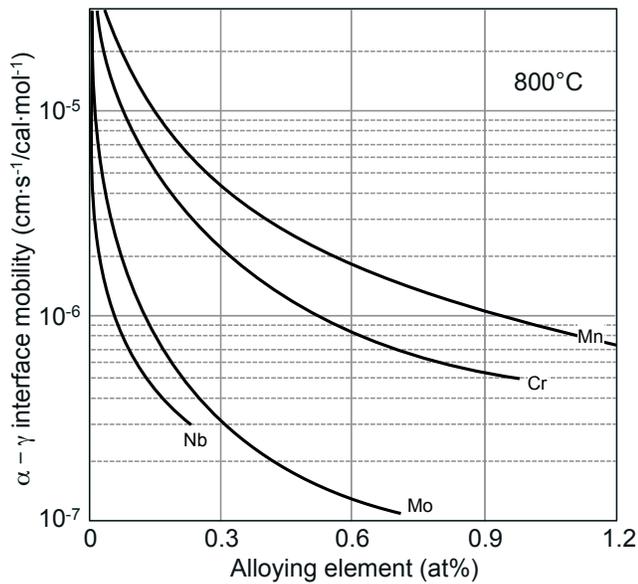


Figure 17. The effect of alloying elements on the mobility of the α/γ interface [18].

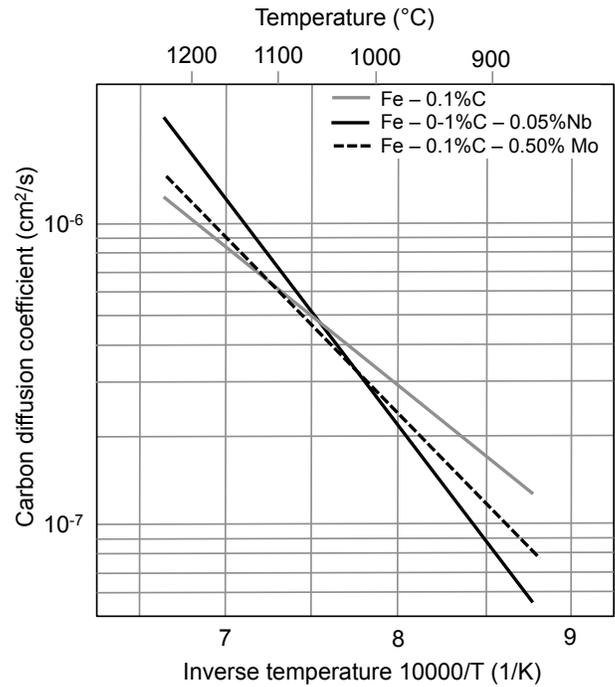


Figure 18. Influence of alloying elements on critical cooling rate for martensite formation in low-C steel [13].

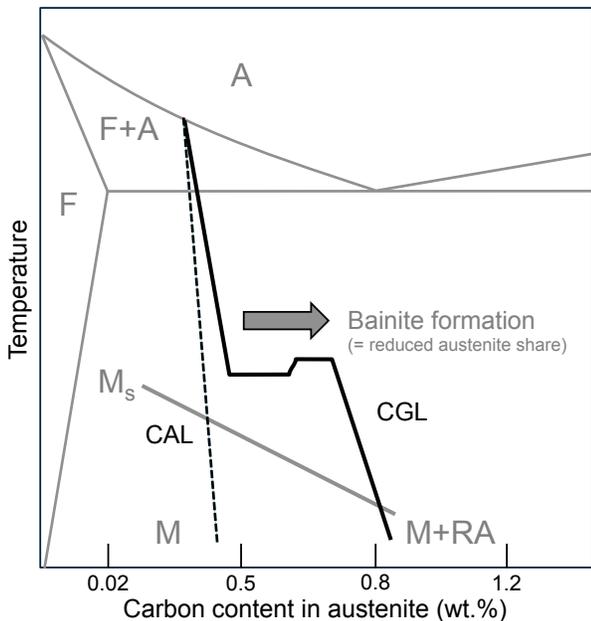


Figure 19. Principal of carbon enrichment and phase formation after intercritical annealing depending on cooling strategy.

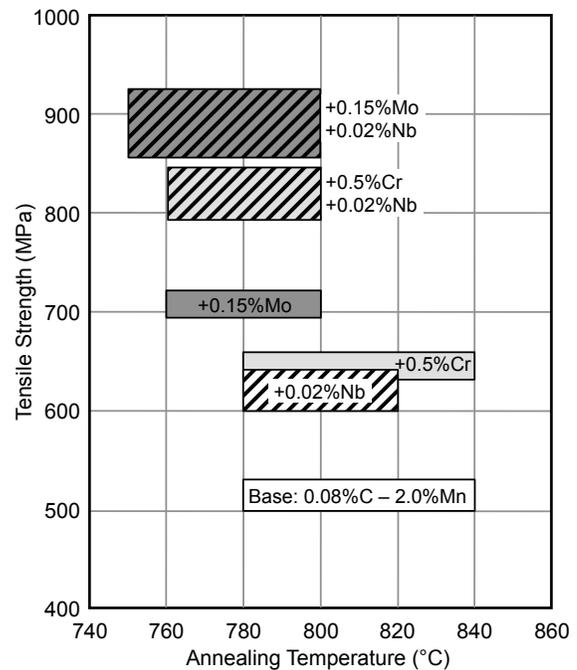


Figure 20. Achievable strength range of low-C DP steels for various alloying concepts.

These two metallurgical effects of Mo have a profound effect on the formation of bainite during the holding period around the galvanizing operation. Elaborate research of the influence of alloying elements in this stage has revealed that Mo makes the bainite transformation extremely sluggish. The growth of (carbide-free) bainite would otherwise result in a decrease of the austenite volume (and simultaneously in carbon enrichment) as schematically shown by Figure 19. It was found that addition of Mo to a multiphase steel alloy maintains the initial martensite start temperature even after extended holding in the bainitic temperature range [21]. Accordingly negligible reduction of austenite volume takes place during holding. The same alloy without Mo addition shows a drop in martensite start temperature already after a few seconds of bainitic holding indicating further carbon enrichment of austenite coming

along with reduced austenite volume. Thus, Mo additions to multiphase steel effectively increase the strength at a given carbon level by providing more (but less carbon enriched) martensite. In this respect Mo-alloying has a clearly stronger effect than Cr.

Besides its impact on phase formation, Mo is the most uncritical hardenability alloying element with respect to surface oxidation in a hot-dip galvanizing process. Oxidation of alloying elements is unavoidable during the continuous annealing process. Oxides remaining on the steel surface prior to entry into the zinc bath cause bad wetting of zinc on the steel surface [20]. Aluminum, silicon and chromium tend to enrich at the surface during ICA and all have a high affinity to oxygen. On the contrary the oxygen affinity of Mo is similar to that of iron. Even if Mo would oxidize it could be reduced in a suitable furnace atmosphere. Consequently Mo is the preferred alloying element in dual phase steels especially when an increased level of hardenability is required such as in low-carbon DP grades with higher strength level.

Combined alloying effects

The effect of various alloying elements added to a base alloy of 0.08%C and 2.0%Mn was evaluated with respect to the strength potential after intercritical annealing treatment (Fig. 20). The base alloy by itself can only achieve a strength level of DP490. Microalloying of 0.02%Nb pushes the strength level into the range of DP590. This strength increase is due to the significant grain refinement as is evident from Figure 21. Addition of 0.5%Cr to the base alloy also achieves DP590 strength level. However the microstructure remains coarse-grained providing a low yield strength. Addition of 0.15%Mo to the base alloy provides a higher strength increase than Cr, which can be in part due to a finer microstructure and in part to more efficient martensite formation. The combined addition of Cr and Nb as well as that of Mo and Nb lifts the strength level to DP780. It can be concluded that a combination of Mo and Nb seems to provide the best platform for reaching the level of DP980.

It is also obvious from Figure 20 that the higher strength levels are achieved towards lower intercritical annealing temperature. This observation was similarly confirmed by Girina et al. [14]. Higher ICA generates more but less carbon-enriched austenite that is less stable and hence tends forming bainite. Detailed investigations of the impact of carbon-enrichment on the required cooling rate can be used to indicate the annealing strategy [22]. At higher carbon enrichment in austenite (>0.2%) the cooling rate depends not very strongly on the actual carbon content. It is in this range easier to adjust a larger amount of austenite during ICA and convert it only partly into martensite. This strategy requires the lower cooling rate. When a larger fraction of martensite is required to reach DP780 or even DP980 strength level, carbon enrichment can drop to values below 0.2% (cf. Fig. 10). In terms of cooling rate, it then becomes more efficient to convert a high share of intercritical austenite into martensite, i.e., using a lower ICA temperature. Chromium and particularly molybdenum alloying to a C-Mn base steel help bringing the cooling rate into a range that can be achieved by modern galvanizing lines.

It is important to note that at lower ICA temperatures the existing cold deformed ferrite will not fully recrystallize. This reflects in relatively high yield strength as well as reduced elongation particularly in Nb microalloyed variants [14]. Therefore it is necessary to anneal at higher ICA temperature and to reduce the austenite volume during a slow cooling phase by forming new ferrite.

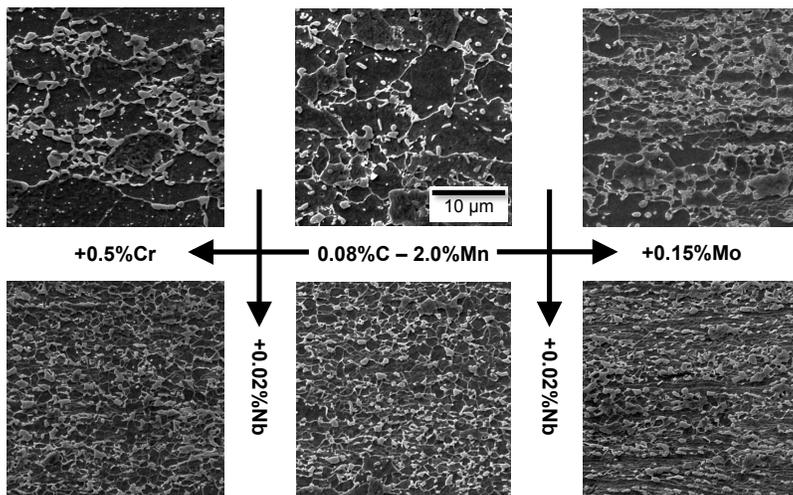


Figure 21. Microstructural effect Nb, Mo and Cr addition to a low-C high-Mn base alloy.

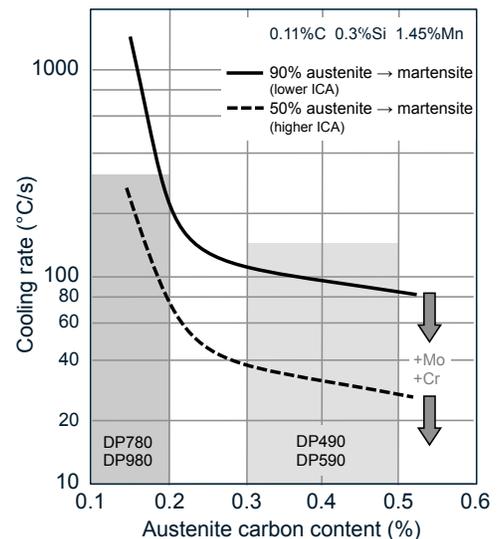


Figure 22. Effect of carbon enrichment on critical cooling rate for austenite-to-martensite conversion [22].

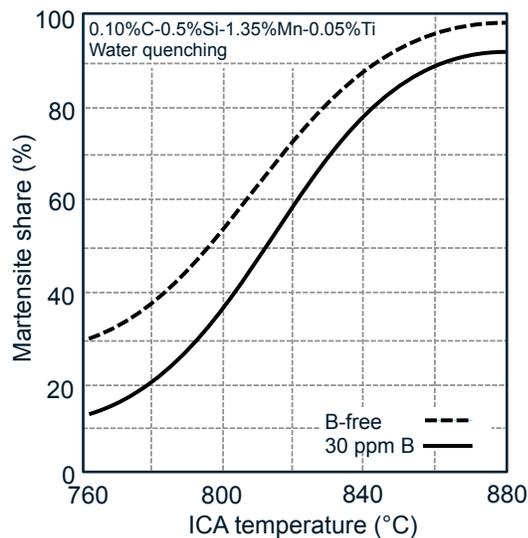


Figure 23. Effect of intercritical annealing temperature on martensite content after water quenching [23].

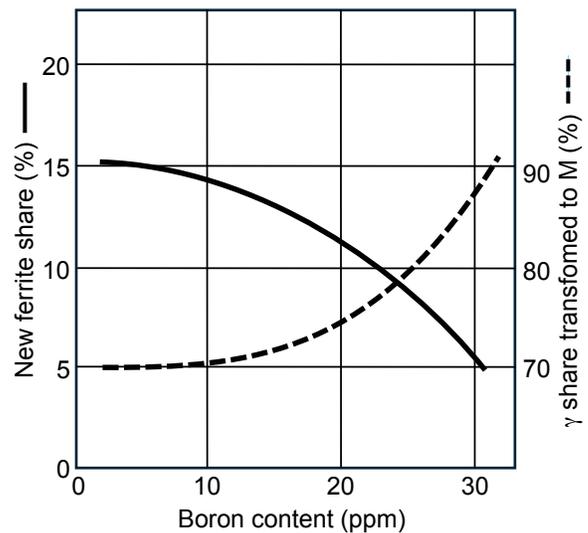


Figure 24. Effect of boron on the volume fraction of new phases formed after ICA at 800°C [23].

Effects caused by boron microalloying

The well-known metallurgical effect of boron is related to its ability of segregating to austenite grain boundaries and inhibiting the grain boundary nucleation of ferrite. Its action, therefore, is to delay the formation of ferrite relative to the formation of lower temperature transformation products. In the production of dual-phase steels, inhibition of ferrite formation would be expected to increase the share of martensite. However, in contrast to the cooling of fully austenitized steels, where boron delays ferrite nucleation, ferrite forms during cooling from the ($\alpha + \gamma$) phase field by the growth of old ferrite that does not require nucleation.

Adding boron to low-carbon manganese DP steel was found to reduce the amount of austenite formed at a constant intercritical annealing temperature (Fig. 23) [23]. The lower share of austenite naturally has a larger enrichment of carbon and is thus more hardenable. Although boron cannot prevent growth of existing “old” ferrite as mentioned before it can however impede the formation of “new” ferrite after ICA. This means that a higher share of the existing intercritical austenite will be converted into martensite upon sufficiently strong quenching as is indicated by Figure 24. Thus, one can expect that the final amount of martensite and thus the strength properties become quite insensitive to the cooling rate by adding a sufficient amount of boron that is usually in the order of 20 to 30 ppm. Since these effects are only effective when boron is in solid solution a Ti addition to fix free nitrogen is also required.

It can be concluded from the observed effects that boron does not directly contribute to austenite hardenability after ICA in contrast to Mo or Cr. Accordingly the addition of boron is practically only of interest for low-carbon DP grades of the upper strength level where it prevents the decay of diluted austenite into new ferrite after ICA. The delay of austenite formation during ICA can also help to better match the requirement of an increased annealing temperature for thorough recrystallization with the actual austenite share needed for the final martensite volume.

CONCLUSIONS

Dual phase steels (and other multiphase steels) are sensitive to local damage upon forming due to the hardness difference between phases. Although this effect is principally unavoidable, countermeasures can be taken to reduce the criticality of this phenomenon. It was demonstrated that the combination of microstructural refinement, homogenization of the individual phases and reduced carbon content in dual phase steel provides the platform for an optimal spectrum of application properties.

A small addition of Nb in combination with appropriate hot-rolling conditions is most effective in achieving such an optimized microstructure. This is reflected in a clearly improved hole expansion ratio and bending angle. Grain refinement also leads to increased strength allowing to reduce the martensite content. Accordingly the increased amount of ferrite results in improved elongation and n-value. Niobium microalloyed DP steel requires a slightly increased intercritical annealing temperature to achieve thorough recrystallization of cold deformed ferrite.

Low-carbon DP steel not only offers improved weldability but also better performance under the conditions of high local straining. Processing of low-carbon DP grades is challenging especially during hot-dip galvanizing treatment. Molybdenum alloying is the most efficient means of making a low-carbon DP concept robust for processing under such conditions. Thereby Mo particularly enhances the stability of weakly carbon-enriched austenite during holding in the temperature range between 400 and 500°C. In that respect Mo

is significantly more powerful than Cr. There is obviously also a synergy when Mo and Nb are alloyed in combination. Molybdenum due to its strong effect of reducing grain boundary mobility seems to further to emphasize the initial microstructural refinement caused by Nb. Furthermore, Mo is uncritical with regard to formation of surface oxides during intercritical annealing and thus avoids problems of surface wetting by liquid zinc, which occur when higher amounts of alternative hardenability alloys are being used.

Boron has no direct impact on the hardenability during processing of intercritically annealed DP steel. However, B can be useful to impede the formation of new ferrite after intercritical annealing. In practice this might be applicable when a larger volume of weakly carbon-enriched austenite is being adjusted during annealing, as is the case for producing low-carbon DP780 or DP980.

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