Niobium Based Metallurgical Concepts and Strategies for the Production of IF-HS and IF-BH Steel Grades

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Abstract

Interstitial Free (IF) steels are being used for- nearly three decades in the construction of car bodies. Originally these were super-soft steel grades offering the highest formability of all carbon steels. Over the last years, however, the share of mild steel is decreasing in car bodies as weight reduction and crash requirements promoted the introduction of high strength steel grades. Also in this area IF based steel concepts have an important role. The yield strength is increased at still excellent formability. In these grades, Nb as a stabilizer of interstitials offers several advantages versus other microalloying elements. This is reasoned by the unique effects of Nb providing efficient grain refinement, transformation control and precipitation hardening. But also as a solute element, Nb can be used to improve relevant properties. The paper discusses the underlying mechansims and indicates how the production process is to be adjusted to obtain the desired property portfolio with respect to formability and weldability.

1. Basic Metallurgical Concepts and Processing of ULC Steels

The production of Ultra Low Carbon (ULC) and Interstitial Free (IF) steels is well known and established in modern steel-making and subsequent processing routes[1-3]. By far the largest consumer of such steel grades is the automotive industry, which continues to place stricter demands upon the steel supplier for a high quality, high performance, cost effective material.

ULC steels are generally classed as steels with carbon content between 20 and 80ppm. In essence IF steels are ULC steels where the interstitial elements of carbon (C) and nitrogen (N) have been completely stabilised. Furthermore, through the use of vacuum degassing units most modern IF steels will have an interstitial level of <30ppm, thus permitting a reduction in the alloy content required for stabilisation. Tab. 1 compares typical ULC, IF and BH products that are available in today's market place. In addition conventional Al-killed drawing quality steel has been included for comparison, the primary difference being the carbon content.

Tab. 1. Properties and chemical composition of typical IF-type products.

Product Type	Spec.	Test Dir.	PS $_{max}$ (MPa)	TS $_{max}$ (MPa)	El. (%)	Ave. r.value	n.value	BH2 (MPa)	C (wt.%)	Ti (wt.%)	Nb (wt.%)	B (wt.%)
Al-killed	---	T	180	310	46	1.7	---	40	0.0250	---	---	---
Formable	FeP06	L	160	300	42	1.7	0.21	-	0.0030	0.070	-	-
	FeP06G	L	160	300	44	1.8	0.21	-	0.0030	0.030	0.020	-
HS-IF	GA180	T	190	350	38	1.5	0.20	-	0.0030	-	0.020	0.0012
	GA220	T	230	370	34	1.4	0.19	-	0.0030	0.030	0.020	0.0020
	GA260	T	270	400	30	1.3	0.18	-	0.0030	0.030	-	0.0015
ULC-BH	GA160BH	L	180	310	42	1.6	0.21	>30	0.0020	0.005	0.005	0.0003
	GA180BH	T	200	350	36	1.4	0.21	>30	0.0025	0.005	0.010	0.0010
	GA220BH	T	240	360	36	1.4	0.21	>30	0.0025	0.005	0.010	0.0010

All these steel types are used extensively in the modern automobile, typical applications being; rear floor pan, hood panels, spare wheel well, front and rear door inners. As will be discussed later, most of these steels are also coated with a protective zinc or zinc-alloy layer.

1.1. Stabilisation of Carbon and Nitrogen

For IF steels, the stabilisation of C and N is typically achieved via small additions of titanium (Ti) and/or niobium (Nb). Although sulphur (S) is strictly not classed as an interstitial, it too must be ‘stabilised’. It is well established that the lower the level of interstitials, the better are the deep drawing properties. This is further illustrated in Fig. 1, which shows that the r-value (measure of deep-drawability) increases greatly with lower carbon levels, a tendency which is enhanced by the stabilisation of the interstitials. Furthermore, Fig. 2 describes the development that occurs in respect to the stabilisation of C by Nb.

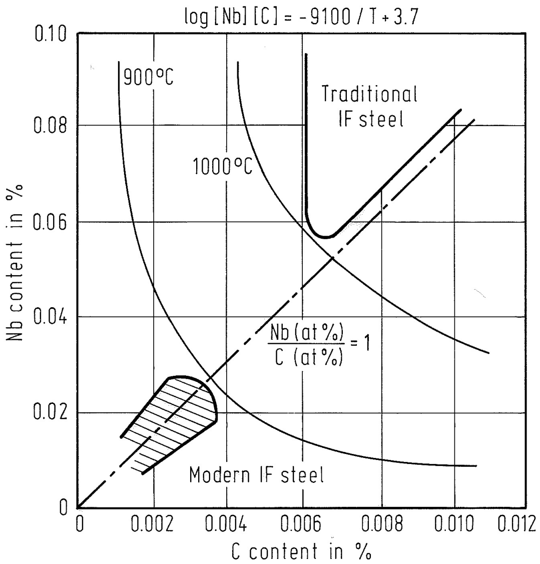

Fig. 1: Effect of carbon content and stabilising on deep-drawability of sheet steel.

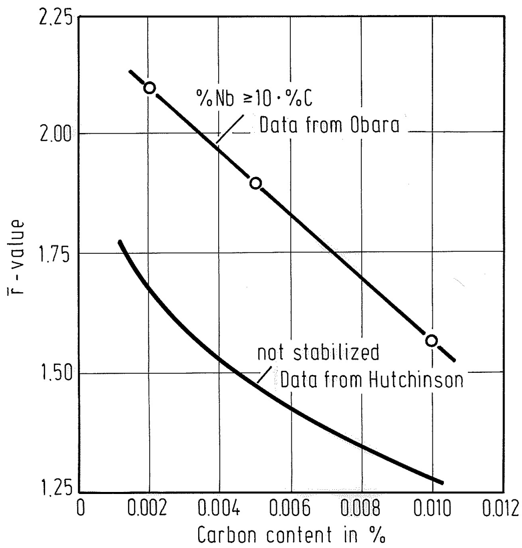

Fig. 2: Carbon stabilisation by Nb in IF steels.

Through stabilisation, the crystallographic texture is improved whilst simultaneously eliminating the yield point (avoiding unsightly Lüders bands) and strain aging characteristics. However, the physical metallurgy surrounding the small additions of Ti and/or Nb is dependent upon which stabiliser is used. Ti shows a high reactivity with all metalloids in the sequence oxygen (O), N, S, C and phosphorus (P). Furthermore, the manganese (Mn) content is also important due to the inherent relationship with S.

For a vacuum treated aluminium-killed Ti-IF steel the first precipitate species that will form is TiN. This will form during solidification, or in the delta ferrite region, then depending on the total S, Ti and Mn content, the formation of TiS is sometimes observed. This sulphide will then be transformed into a titaniumcarbosulphide by picking up carbon in the austenite region. This reaction has been studied in detail and seems to be the preferred one compared to MnS and TiC formation in Ti stabilised IF steels. The minimum amount of titanium required for full stabilisation, based on a stoichiometric approach is:

$$Ti_{\text{stabilised}} = 4C + 3.42N + 1.5S$$

Although it has been proposed that an excess Ti may give rise to high r-bar values, additional free Ti may encourage the formation of an iron-titanium-phosphide in the ferrite region. Fig.

3 demonstrates the precipitation start of the complexes that are generally observed in a modern IF steel.

For Nb stabilised IF steels, the Nb will react with only C and N to form carbonitride precipitates. The manganese available in these types of steels is enough to naturally fix the sulphur in addition the preferential precipitation of AlN will also occur thereby reducing the amount of N available for Nb(CN) precipitation. As Al is always present in such steels, the required Nb level has to be just stoichiometric to carbon:

$$Nb = 7.74C$$

As the bulk N should be scavenged by the Al, only NbC will be formed during or after the austenite to ferrite phase transformation according to the solubility product. In addition, an excess of solute Nb has been reported to have several benefits, and these studies are still continuing. It is likely that the solute Nb will compete for grain boundary sites with phosphorus, which is detrimental to cold work embrittlement resistance[4)] (as will be discussed later).

As a comparison, for a typical IF steel grade with the following chemical composition; 30ppm C, 35ppm N and 0.008%S, at least 0.036% Ti will be required to fully stabilise the steel, whereas only 0.023% Nb is required to achieve an IF status. In either case, with the low amount of stabilising elements actually required, the overall alloying costs for both elements are reasonably low. Nevertheless, by taking into consideration that a higher quantity of Ti is required for complete stabilisation, coupled with its lower level of recovery in alloying, a viable case exists for the use of Nb for full stabilisation, in addition to better coating performance.

For dual stabilisation, the usual practise is to add sufficient Ti to react with N (to form TiN), leaving Nb to scavenge C as NbC:

$$Nb = 7.74C \text{ and } Ti = 3.42N$$

Again, the sulphur in such steels will combine with Mn to form MnS. However, if there is an excess of Ti to account for the N, then TiS will also form in addition to MnS. Furthermore, an excess of Ti may lead to undesirable surface streaking which can be observed in the final coated product (Fig. 4). This type of defect has been more frequently observed in a fully Ti stabilised IF steel, and the associated cost penalties are of great concern to all.

If there is an excess of Nb (solute) available then this may be beneficial in producing a good coating integrity in the final coated product if the solute Nb resides near or at grain boundaries. Dual stabilised Ti-Nb IF steels are also known to be less susceptible to cold work embrittlement and also exhibit better spot-weld characteristics then Ti stabilised only steels. If the steel is under-stabilised then excess carbon will exist in solid solution, and this will consequently generate Bake Hardening (BH) characteristics.

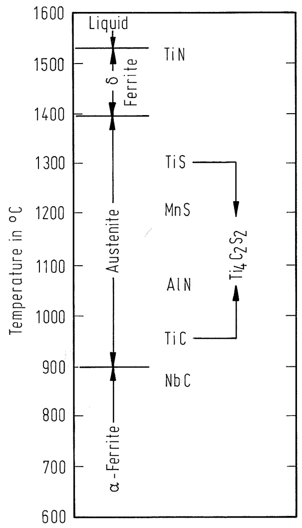

Fig. 3: Likely precipitation sequence for various compounds in a modern IF steel.

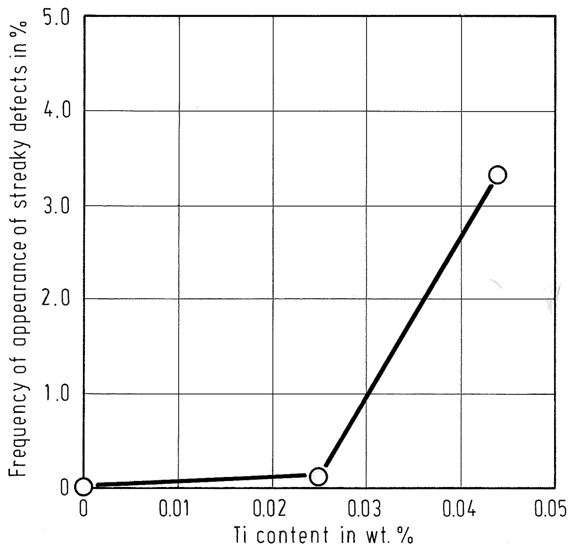

Fig. 4: The influence of Ti content on the frequency in appearance of streaky white defects in galvannealed IF-sheet

1.2. Hot Mill Processing

For ULC-IF type steels controlling the hot band microstructure is crucial in developing a high r-value in the subsequent cold rolled and annealed product. Typically, continuous cast slabs are reheated at temperatures ranging from 1250 to 1100°C. Although it has been reported by some researchers that a reduced reheat temperature (<1100°C) will result in an improvement in the r-bar value of cold rolled and annealed IF sheet steels for both Nb and Ti IF, care has to be taken such that the slab can still be successfully rolled, with respect to the rolling loads. Hence, in most cases the lowest practised reheat temperatures range from 1180 to 1200°C.

The roughing temperature and reduction is important in developing a high r-value. Research has demonstrated that a large uniform reheated grain size together with an initial heavy reduction and rolling below the recrystallisation temperature is beneficial. One of the crucial factors in hot rolling ULC and IF type steels is that the finishing mill exit temperature must remain above the austenite to ferrite transformation point. Typical aim finishing temperatures of 900°C or above are practised. However, this is dependent upon which alloying strategy is used. In general, it has been observed that Nb and Ti-Nb IF steels are more sensitive to the finishing temperature than Ti-IF steels, with both the r-value and ductility increasing with decreasing temperature. As would be expected, the Nb alloyed steels exhibit a finer hot band grain size due to solute drag effects. In line with this phenomenon, higher reductions and increased rolling speeds will also further enhance properties.

Although the hot rolling forces encountered are lower than that for high carbon or HSLA steel, by close attention, an increase in rolling load for the Nb-IF can be observed, indicating that no recrystallisation is occurring below a distinctive temperature. This is caused by the solute drag of Nb atoms on the austenite grain boundaries.

It has to be noted that the role of Nb here is different from the conventional method of refining the grain size in low carbon steels (<0.010wt.%), whereby precipitates of Nb(C,N) pin the austenite grain boundary and promote pancaking.

The solute drag mechanism can also be used to explain that the Nb-IF has a slightly lower austenite to ferrite transformation temperature than a Ti-IF. Furthermore, both the retardation of austenite recrystallisation and the lower transformation start temperature will result in greater numbers of nuclei for ferrite formation, hence a finer ferrite grain size in the hot band for a Nb-IF steel compared to a Ti variant. On exit of the finishing train, the strip is immediately cooled on a run-out-table to an aim coiling temperature ranging from 600 to 700°C. A higher coiling temperature is normally applied to the Ti-Nb and Nb-IF steels due to the development of Nb precipitates which require some degree of coarsening (Ostwald ripening), permitting recrystallisation to talk place during annealing following cold rolling. It is important to develop relatively coarse and widespread precipitates, as this will enhance the development of the required (111) texture necessary for a high r-value. Fine closely spaces precipitates will inhibit recrystallisation and grain growth during annealing and result in inferior properties. However, it must be noted that application of a very high coiling temperature (≥725°C) will results in the formation of an aggressive scale on the as-hot rolled strip, which will make it more expensive to remove during pickling.

1.3. Cold Rolling

Following pickling, the hot-rolled coil is cold reduced typically by 80-85% in a continuous rolling operation. The main purpose being to reach the desired final gauge and to optimise the r-value through the development of a high energy deformed structure from which recrystallisation of grains with favourable textures will nucleate and grow during the subsequent annealing process. Although investigations have revealed that the optimum cold reduction is near 90%, this is difficult to achieve in practise due to the development of excessive cold rolling forces and therefore reductions are limited to 80-85%. To reintroduce a high degree of elongation and formability, the cold-rolled strip is then annealed.

1.4. Annealing

The Batch Annealing (BA) route has traditionally been used to anneal cold rolled mild steel Al-killed steels to make highly formable grades. The necessary high r-value being obtained mainly as a result of controlling the aluminium and nitrogen during hot rolling and the BA route; the hot strip must be coiled at a low temperature (<600°C) to retain the AlN in solution, which will then precipitate during the slow heating up cycle in the BA route, thereby inhibiting recrystallisation and developing a satisfactory texture. During the slow cooling cycle (of several days) any liberated carbon is re-precipitated resulting in a completely non-ageing steel. However, as the BA route took several days to complete and could produce inconsistent final properties through the coil, the Continuous Annealing Line (CAL) was developed, permitting processing to be completed within a matter of minutes.

Due to the completely different processing route for the CAL, to develop an equivalent formable product requires the hot rolled Al-killed mild steel to be coiled at a high temperature, thereby reducing the solute interstitials through coarsening of cementite particles and the precipitation of AlN (another option is also the addition of boron to form stable boron nitrides). Additionally, the Al-killed steel must be overaged to remove any carbon that may have been liberated during the annealing cycle. For ULC steels, the problem was overcome by making additions of carbide forming elements such as Ti. However, for Al-killed steels this becomes more a problem as the rapid cooling from the annealing temperature for the CAL will promote the yield point phenomenon due to lack of an overageing sequence. The important difference between the annealing of conventional ULC–IF steels and Al-killed steels is that the IF steels are usually completely stabilised with either Ti and/or Nb. Therefore, they do not require an overageing section in their annealing cycle.

As a result they are ideal for processing on a hot dip coating line, which integrates annealing and coating into one process. It is this latter fact that established IF steels as the lead steel for the automotive industry. Nevertheless, and although not discussed here, both ULC-IF and BH grades can be produced via the BA route. Today, the BA route is predominately used for uncoated and electrolytically coated products.

1.5. Hot Dip Coating

As the majority of ULC and IF steels will be used for exposed automotive parts, they are often coated to protect the substrate from corrosion. The protective coating can be applied via hot dip galvanising or electrogalvanising. However, it is true to note that most auto-manufacturers will prefer a hot dipped product. In general there are two types of zinc coatings that can be applied; galvanising and galvannealing. In the latter process the hot dipped steel is annealed to permit a controlled diffusion between the zinc and the substrate to produce a Fe-Zn intermetallic compound.

There are many variables that will govern how well an applied coating will perform, and it is beyond the scope of this article to discuss it in any depth. Nevertheless, it has been reported that Ti-Nb IF steels provide a better substrate for hot dip coating and galvanising than Ti-IF steels i.e. better powdering resistance. This is due to a more uniform coating reaction taking place that yields greater consistency in coating structure. Furthermore, it has been observed that the presence of Nb at or near the grain boundaries enables the regulation of the Fe-Zn coating reaction by slowing it down. For Ti-IF steels it has been noted that the Fe-Zn reaction proceeds at an accelerated rate, which produces ‘outburst’, an uncontrolled reaction, which is highly undesirable. Additionally, the formation of the brittle Γ–phase may also result. Thus, the majority of steel-makers tend to use a Nb and/or a Ti-Nb IF steel for galvanising not only for better coating performance, but in keeping the Ti content below 0.020wt% the formation of an undesirable streaky surface is almost eliminated.

2. High strength ULC-IF steels

The increased strength compared to mild IF steel is achieved by adding solid solution hardening elements, and besides manganese of around 0.35 % the very effective element phosphorus is widely used, which will increase the strength by 100 MPa per 0.1wt.%. The strength increase by phosphorus addition is significantly higher in a niobium stabilized IF steel as compared to titanium stabilized one (Fig. 5), which can be explained by the fact, that titanium forms finally also titanium phosphides besides oxides, nitrides, sulphides and carbides[5)]. Consequently the amount of phosphorus in solid solution, the responsible fraction for the strength increase, is being reduced at higher titanium level.

The lower strengthening effect in the Ti-IF can be explained by the formation of a FeTiP phase, thus lowering the solution hardening effect of phosphorus. However, phosphorus can segregate to the grain boundary causing secondary cold work embrittlement (SCWE)[6)].

This embrittlement occurs especially in IF steels sheets during impact loading at lower temperatures following press forming and it has been generally observed that Ti-IF steels have higher SCWE ductile to brittle transition temperatures (DBTT) than Nb and Nb-Ti IF steels. This is because in a fully stabilised Ti-IF the grain boundaries are more depleted than in an Nb-based IF, as Ti has a greater affinity for C than Nb. The presence of solute Nb in IF steels has been rationalised to be the contributing factor for the more favourable SCWE resistance.

To overcome this problem a small quantity of boron (B) is usually added, whereby the boron occupies sites at the grain boundary thereby leaving the P in solution within the grains (Fig. 6). The addition of boron will also increase the hardenability and also delay the recrystallisation of austenite, the later occurring again via the mechanism of solute drag on the grain boundaries. The addition of boron in Ti-Nb IF steel will change the hot band microstructure by further grain refinement. The application of a low coiling temperature will also aid refinement and also introduce irregular shaped grains, reflecting the presence of a substructure. This can be related to the high dislocation densities introduced during the austenite to ferrite phase transformation.

The production of HS-IF re-phophorised grades is limited to continuous annealing or hot-dip galvanising lines. This is again due to the P atoms segregating to the grain boundaries during the slow cool after recrystallisation for the BA route. However, here the addition of B is not as effective, and so other alloying options have to be employed.

Compared to a titanium-stabilized grade, niobium stabilized IF steel also exhibits a finer grain size and thus higher yield strength. This derives from the already finer grain size in the hot strip material as niobium retards the austenite recrystallisation during the final rolling passes due to a solute drag effect of the large niobium atom[7)]. Furthermore, niobium in solid solution of the austenite also retards the transformation into ferrite, which has an additional grain refining effect. It should be noted however, that niobium stabilized IF steel requires a somewhat higher annealing temperature to achieve complete recrystallisation after cold rolling as compared to titanium stabilization[1)].

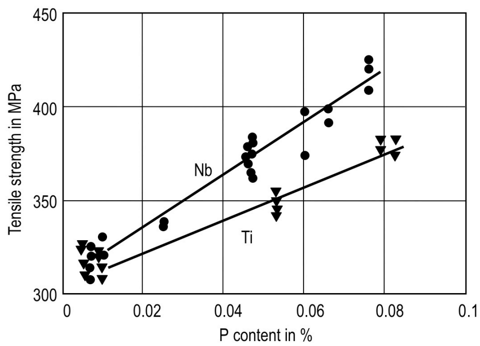

Fig. 5: Strengthening by phosphorus additions in Nb or Ti-IF sheet steels.

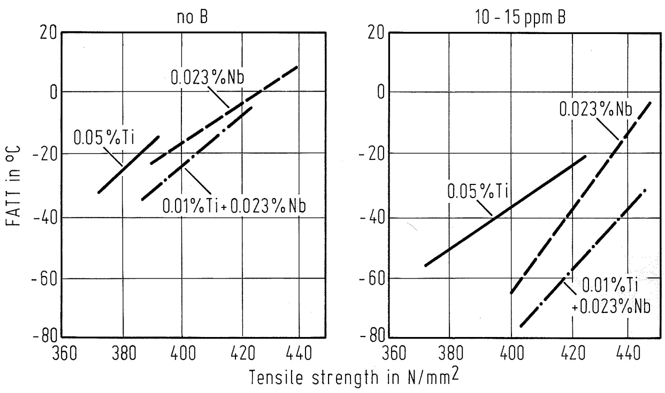

Fig. 6: Effect of boron on the ductile-to-brittle fracture transition temperature (FATT) of various high strength IF steel concepts.

An innovative approach for high strength IF steel alloy design is based on the dual addition of Ti and over-stoichiometric Nb[8)]. This alloy design leaves a large fraction of Nb in solid solution. As explained before, a non-polygonal ferritic structure is formed when such steel is coiled at low temperature after hot rolling. This hot band shows a quite strong fibre texture similar as in a cold rolled material. Thus the development of a recrystallised texture leading to a good r-value is enhanced. Moreover, due to the finer grain size there are more nucleation sites for <111>//ND recrystallised grains upon annealing. Solute Nb in such steel segregates to the grain boundary and delays recrystallisation even at a temperature close to A_{r1} compared to Ti-only stabilized steel. The solute drag effect of Nb raises the recrystallisation temperature during annealing by about 50°C so that an annealing temperature of minimum 800°C is required after cold rolling. On the other hand a beneficial practical effect of grain refinement and Nb segregation to grain boundaries is the strongly reduced temperature below which secondary work embrittlement occurs. Thus the addition of a small amount of boron to the alloy, which however is detrimental to the r-value, can be avoided. Another an important practical implication of Nb grain boundary segregation is the avoidance of grain coarsening and consequently softening in the heat affected zone of spot and laser welds (Fig. 7c).

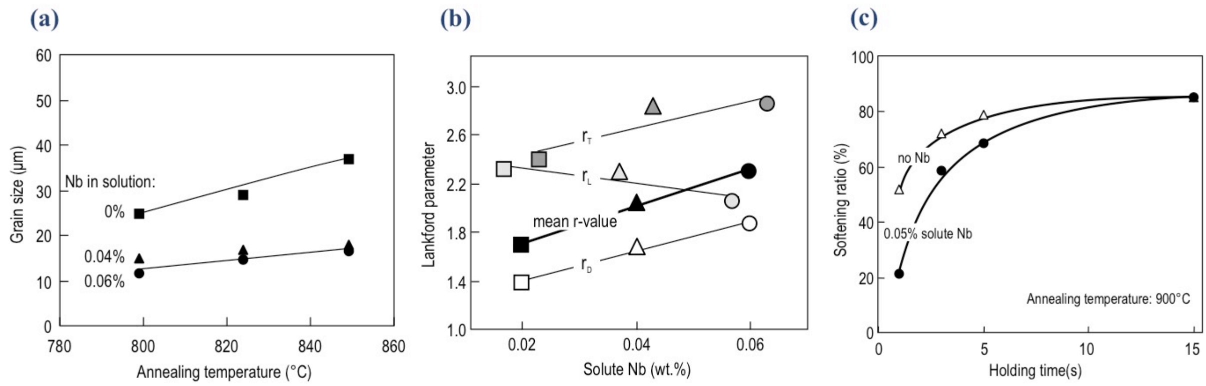

Fig. 7: Effects of adding Nb to a Ti-stabilized IF high strength steel (0.003%C-0.002%N-0.07%Ti-Mn-P); (a) Evolution of final grain size in function of annealing temperature at various solute Nb levels; (b) Evolution of r-value in function of the solute Nb level; (c) Softening behaviour for annealing close to the A_{r1} temperature.

An alternative metallurgical concept working using the effect of grain refinement as well as precipitation strengthening has been developed by Urabe et al.[9)]. The C content is 2 to 3 times higher than in a conventional IF steel. Accordingly the addition of Nb is also higher than usual. This approach not only leads to grain refinement but also to enhanced precipitation of fine Nb(C,N) particles increasing the strength. Therefore the amount of solid solution strengthening elements, particularly phosphorus, can be reduced which in combination with grain refinement is beneficial concerning secondary work embrittlement. The grain refinement leads to a tensile strength increase of around 30 MPa compared to the conventional concept. Interestingly, the yield ratio is decreased meaning that the yield strength is lower than in the conventional concept (Fig. 10). The lower yield stress has been related to the formation of precipitate-free zones after annealing. During recrystallisation the grain boundaries move and leave behind a zone of much less precipitates being coarsened by Ostwald ripening. This means the precipitation strengthening effect vanishes and yielding can occur at a lower stress level in that zone. To obtain effective Ostwald ripening, the heating rate during up-heating to soaking temperature must remain limited. With an increasing amount of precipitate-free material volume the bulk yield strength drops measurably. Since the tensile strength is maintained the n-value becomes larger with increasing precipitate free material volume. The effect appears to saturate above a volume fraction of around 8 percent precipitate free material (Fig. 11). This volume fraction can be controlled by the heating rate in the annealing line and the soaking temperature. The r-value and thus the deep-drawability of the new concept is particularly good. This is again related to the formation of a favourable fibre texture with <111>//ND. The key for forming this pronounced texture lies in the refined hot band structure providing more nucleation sites due to the enhanced grain boundary area.

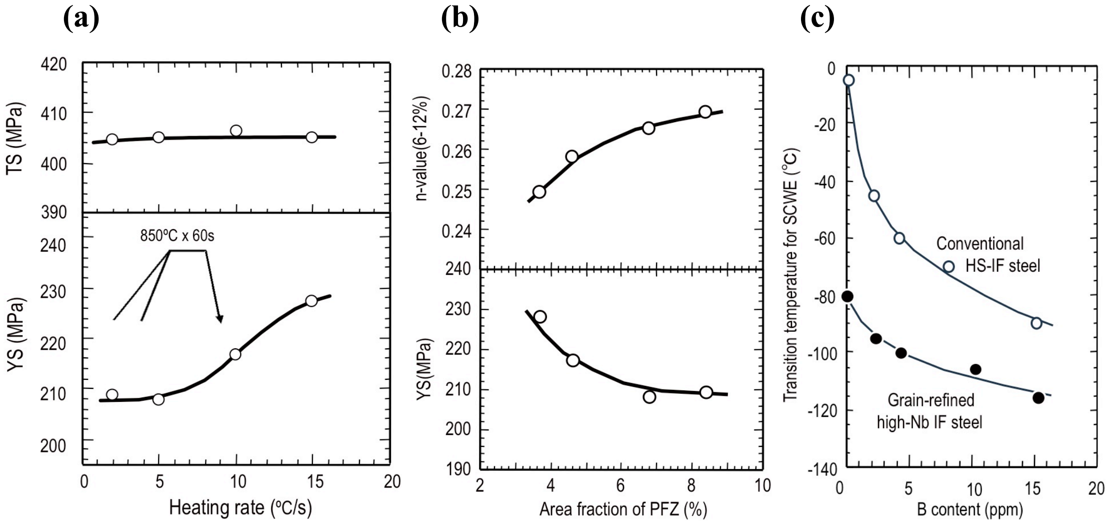

Fig. 8: Properties of a high-Nb grain refined and precipitation hardened IF steel (52 ppm C, 0.07% Nb, 0.6% Mn, 0.04% P); (a) Yield and tensile strength as a function of heating rate during annealing cycle; (b) Influence of precipitate-free zone area fraction on yield strength and n-value; (c) Effect of boron addition on SCWE transition temperature.

3. Bake Hardening Steel

Although the development of soft and highly formable extra deep drawable quality (EDDQ) steel grades meet the basic requirements of the automotive industry, it became evident that for exposed body panels, these steels exhibited poor dent resistant characteristics. Therefore, to avoid using higher strength steel, which would decrease the formability, bake hardenable steel was developed. This type of steel exhibits a strain aging phenomenon, which occurs at a temperature being equal to that of an automotive paint baking temperature at approximately 170°C for 20 minutes. The bake hardening response is related to the presence of solute carbon that prevents the movement of dislocations that result during press forming i.e. Cottrell atmospheres of carbon or fine carbide precipitates are formed around the dislocation cores, pining the dislocations. However, it is crucial to control the solute carbon level, as excess will lead to the appearance of stretcher strains in the pressing process.

Both conventional batch annealing and continuous annealing process routes are capable of producing BH grades. For IF-type steels, in essence there are three main routes to develop a BH response: (i) to have an understoichiometric with respect to the amount of Ti and/or Nb added, (ii) to employ a high annealing temperature to ensure partial dissolution of carbides, and (iii) to employ a mixture of (i) and (ii). In general the latter route is widely practised. During the annealing of the steel a degree of carbon will be liberated via dissolution of carbides of Nb and/or Ti. This carbon may then segregate to the grain boundary thus providing strengthening and/or immobilise the previously mobile dislocations formed in the material during press forming. Because the dissolution of carbides occurs after the completion of recrystallisation, a high r-value plus bake hardening characteristics are displayed.

Fig. 9a translates the BH2 requirements of carmakers into the range of solute carbon providing the bake hardening effect. Based on detailed laboratory studies solute carbon should be in the range of 5 to 10 ppm. Below 5 ppm the BH2-value is not meeting minimum specifications and beyond 10 ppm natural ageing becomes prominent leading to stretcher-strain defects during press forming. Thus, alloy design and processing conditions must be carefully adjusted to control the solute carbon within this range. Optimum control of the Nb/C atomic ratio close to stoichiometry can lead to a preferable 30-50 MPa BH response when coupled with annealing temperatures in the range of 820 to 850°C (Fig. 9b). To obtain equivalent values in a Ti-steel the Ti/C ratio has to exceed unity, and the annealing temperature necessary will be higher due to the greater stability of TiC. In addition, as Ti has a high affinity for S and N, it is difficult to accurately control the Ti/C ratio. Consequently Nb or Nb-Ti based steels are preferred for the BH steel grades[10,11]. Titanium-free ULC steel exhibits also a lower recrystallisation start temperature and by this means promotes a higher Lankford value in the final product, allowing compensation of the negative effect of solute carbon in the hot band.

For various reasons it becomes interesting to design bake hardening steels offering an increased resistance against natural ageing. For instance, extended storage of BH steel in geographical regions with very warm climate can lead to excessive yield point elongation. On the other hand, for practical reasons such as an increased dent resistance a higher BH2-value could be demanded. This would require a higher solute carbon content, which again increases the ageing sensitivity. Research has indicated that the addition of a small amount of molybdenum can solve this problem[12]. This has been reasoned by short-range atomic interaction between C and Mo reducing the mobility of carbon at room temperature. At increased temperature such as used for paint baking the effect disappears providing the full bake hardening effect. Fig. 9c indicates the effectiveness of a small Mo addition to a Nb-Ti partially stabilized ULC steel. The increased amount of solute carbon providing the extended bake hardening effect can be obtained by a sub-stoichiometric addition of Nb (Fig. 9b).

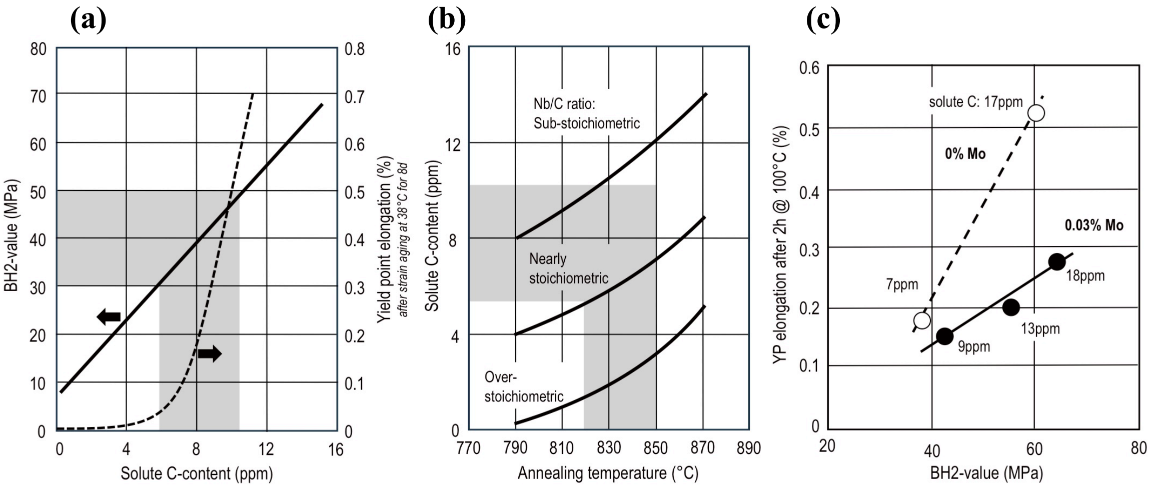

Fig. 9: (a) Relationship between solute carbon, BH2-value and yield point elongation in ULC steel; (b) Alloy design and annealing conditions to achieve the specified range of BH2-values; (c) Effect of Mo addition on yield point elongation in Nb-Ti ULC steel[12].

Sometimes it is a challenge for the steel shop to adjust a low and accurate carbon content in the melt, be it due to less capable equipment or lack of experience. It is also possible producing a bake hardening steel based on Nb stabilized ULC metallurgy[10,13]. This can be achieved by redissolving a fraction of the precipitated NbC in the annealing cycle requiring, however, an increased annealing temperature. Fig. 10 schematically describes the metallurgical mechanisms acting in this processing cycle. Based on solubility product calculations, Fig. 11 indicates the processing window for achieving a solute carbon content of 5-10 ppm necessary to obtain the specified BH2-value of 30-50 MPa. This is most effectively the case at annealing temperatures in the range of 850 to 870 °C. At annealing temperatures above 870°C heat buckling becomes a problem. A sufficiently high cooling speed after soaking is mandatory to avoid reprecipitation of NbC. It has been observed, that a lower Ti content in the ULC steel allows easier dissolution of NbC during the annealing process. A remarkable retardation of carbide dissolution occurs with the existence of rather complex precipitates observed in steel with an over-stoichiometric Ti/N ratio.

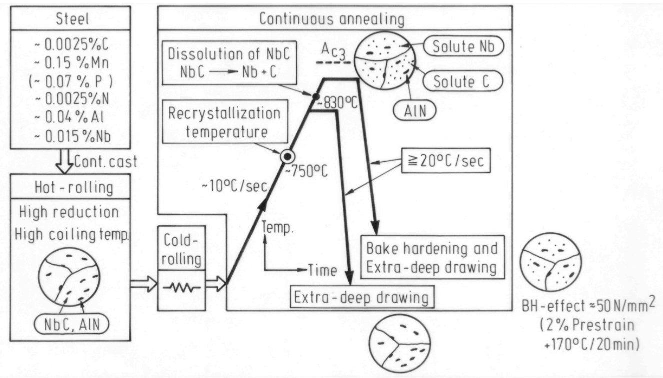

Fig. 10: Metallurgical mechanisms during high temperature annealing of a fully Nb-stabilized ULC steel[13].

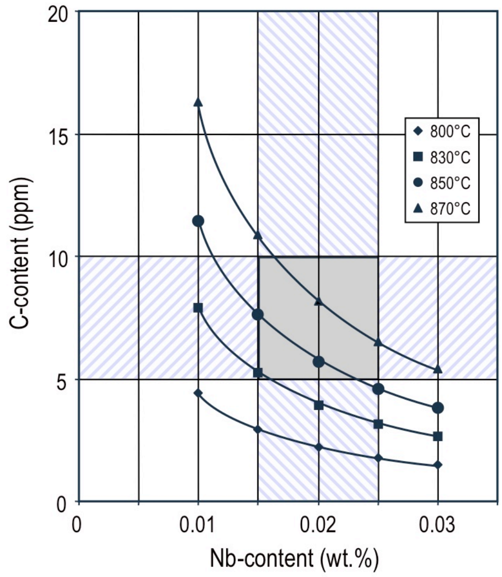

Fig. 11: Processing window for high temperature annealing (Nb-only stabilization).

Conclusions

In traditional soft IF steels, Ti and/or Nb additions have the function of stabilising carbon and nitrogen interstitials. For the production of such steel, Ti concepts are often preferred since they provide the lowest yield strength. However, especially for galvannealed coated sheet, Nb or Nb-Ti stabilised steel exhibits a much better surface quality and is hence the preferable concept.

In recent car body design substantial amounts of high strength IF steel and bake hardening steels are used. For these steel grades Nb based concepts offer intrinsic advantages by:

- Grain refinement achieved during hot rolling
- Precipitation hardening achieved during coiling or annealing
- More effective solid solution hardening of phosphorus
- Improved secondary cold work embrittlement behaviour even without boron addition
- Improved Lankford parameter, planar isotropy and strain hardening behaviour
- Improved welding behaviour
- Tighter solute carbon control in bake hardening grades
- Additional release of solute carbon during high temperature annealing

Innovative metallurgical concepts such as high-Nb or Nb-Mo based alloys in combination with adapted processing conditions have the potential of providing improved IF and BH steel grades with a superior property profile.

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