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DEVELOPMENT OF THIRD GENERATION ADVANCED HIGH STRENGTH STEELS FOR AUTOMOTIVE APPLICATIONS

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Abstract

A weight reduction, driven by constantly more stringent CO_2 regulations along with an improved crashworthiness of modern vehicles, motivates the steel industry to develop novel advanced high strength steels (AHSS), which combine simultaneous increase of both rather contradictory properties: strength and ductility. The steel industry is currently associated with the development of the third generation (3-GEN) AHSS, where the excellent combination of strength and ductility can be achieved by an employment of the transformation-induced plasticity (TRIP) effect - the transformation of a large amount of metastable retained austenite to martensite during straining. This beneficial phenomenon takes place during forming operations with a global ductility such as deep drawing. Apart from that, an excellent performance of the steel during forming operations with a local ductility (e.g. sheet cutting, hole-expansion etc.) becomes also pre-requisite when forming complex components for the automotive industry. Furthermore, a component weight saving via steel density reduction may also come into play, while considering the development of these steel grades. The present paper reviews the above-mentioned aspects resulting in the development of TRIP bainitic ferrite (TBF), quenching and partitioning (Q&P), medium manganese and density reduced TRIP steels for the application in modern automotive platforms. These new developments are thoroughly described from their processing, microstructure and resulting mechanical properties point of view.

Keywords: TRIP effect, retained austenite, AHSS, mechanical properties, density reduction

1 Introduction

The persistent demand of the automotive industry on advanced high strength steels (AHSS) due to their lightweight potential, excellent crash behavior, outstanding formability and reasonable pricing compared to non-ferrous material solutions drives the development of new high strength steel grades. Challenging regulations within the EU legislation concerning CO_2 emissions of new passenger cars increase the research and development activities concerning high strength steel grades as potential lightweight materials. [1] Studies show that in the automotive industry the share of high strength steels (HSS) will increase up to 38 % in 2030, which means that the applications will more than double compared to 2010. [2] To meet the demanding requirements of the market, the improvement of existing concepts as well as the design of a new generation of AHSS will be necessary.

In general, a correlation between strength and ductility exists, especially for mild steels and conventional high strength steels (HSS). The increase of the strength typically leads to a decrease in total elongation, whereas the increase of ductility is accompanied by a loss in strength. While steels with high ductility are well suited for deep drawing applications, the steels having high strength resist higher loads during the crash event. In order to fulfill the requirements of the automotive industry the steel industry attempts to optimize the mechanical properties by developing AHSS having both high tensile strength and enhanced total elongation. [3]

The 1-GEN AHSS includes the DP (Dual Phase), TRIP (Transformation Induced Plasticity), CP (Complex Phase) and MS (Martensitic) steels. These steels mainly have a multiphase structure, which combines the advantages of the individual microstructural constituents, which are characterized by an excellent combination of strength and formability. [4]

The 2-GEN AHSS includes TWIP (Twinning Induced Plasticity), Nano-TWIP, Duplex and Triplex steels, which also possess an excellent combination of mechanical properties. [4, 5] However, their Mn content usually exceeds 15 wt-%, which makes them substantially expensive for the automotive industry and their application in car bodies is therefore rather limited.

The development of a steel family belonging to the 3-GEN AHSS is recently ongoing. The mechanical properties of Medium-Mn, TBF (TRIP-aided Bainitic Ferrite), Q&P (Quenching and Partitioning) and δ-TRIP steels close the gap between the 1-GEN and 2-GEN AHSS. [4, 6] This trend is schematically illustrated in Fig. 1, whereby the predicted region from the standpoint of above-mentioned mechanical properties for the 3-GEN AHSS is clearly envisaged.

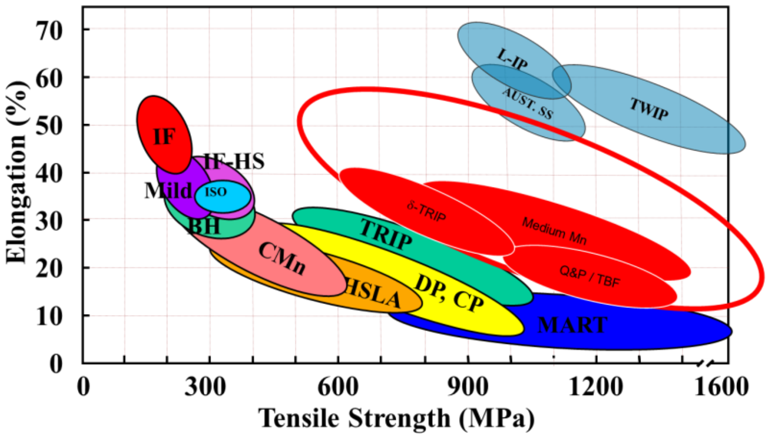

Fig. 1 *Mechanical properties of steels for automotive lightweight applications. The red-circled region represents the developments related to the 3-GEN AHSS. [7]*

The present contribution intends to give an explanatory insight into the development of the 3-GEN AHSS for the automotive industry in terms of their microstructure, mechanical properties and applied heat treatment. Furthermore, the TRIP effect, well known as the main strengthening mechanism operating during straining of this steel group, will be elucidated in detail. Finally, besides strength increase, an alternative mechanism for lightweighting of car structures, namely density reduction by remarkable Al additions, will also be thoroughly discussed.

2 TRIP effect

The retained austenite is considered harmful for the majority of steel applications (e.g. by tool steels or many applications in the defense industry) and its presence in the microstructure should therefore be avoided. It has been reported that the retained austenite can increase the internal stresses in the microstructure, leading to the subsequent crack initiation and thus vast in-service properties can be significantly impaired. Among others, machinability and durability of tool steels can be deteriorated, when the retained austenite is present in the microstructure. [8]

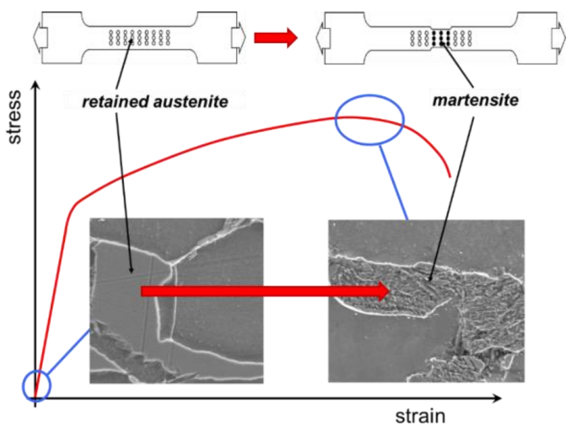

Fig. 2 *Strained-induced transformation of retained austenite to martensite during forming modern TRIP-assisted AHSS for automotive applications.*

On the contrary, in the body-in-white (BIW) applications for the automotive industry, AHSS containing the metastable retained austenite transforming to martensite during straining, represent the cutting-edge development, since rather contradictory mechanical properties, i.e. strength and ductility, can be simultaneously improved. It is aparent that the retained austenite must be well adjusted and controlled during steel processing in order to ensure the desirable improvement of mechanical properties.

As already mentioned above, the TRIP effect is based on the strain-induced austenite to martensite transformation during forming, resulting in an excellent combination of strength and ductility (Fig. 2). By enriching the retained austenite with C and by adjusting the optimal grain size, the martensite start temperature (M_s) is lowered below room temperature (RT), avoiding the formation of martensite during cooling (Fig. 3). For this reason a sufficient stability of retained austenite can be achieved at RT. [9]

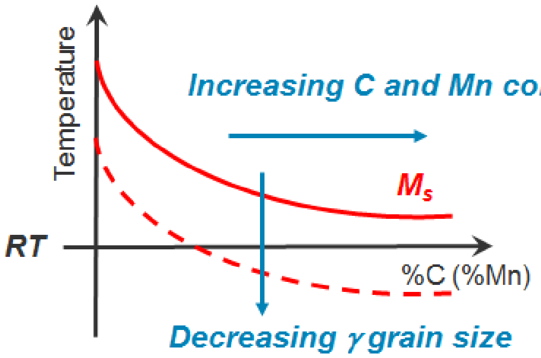

Fig. 3 *Schematic explanation of the retained austenite stabilization at RT.*

There exist more explanations how the TRIP effect improves the mechanical properties of a steel. [10] The most accepted one relates to the formation of additional mobile dislocations in ferrite in vicinity of strain-induced martensite. Fig. 4 represents a schematic, explaining the development of dislocation density in single phase, DP and conventional TRIP steel grades. In conventional single-phase mild steel grades, the overall dislocation density is primarily depending on the statistically stored dislocations. In DP steel grades with a soft ferritic matrix and hard martensite islands, a large number of geometrically necessary dislocations are mainly produced at the beginning of plastic deformation. The larger the difference in strength between the matrix and the second phase, the higher the number of geometrically necessary dislocations. Therefore, the work hardening behavior of DP steels is improved. In TRIP-assisted steels, further dislocations are generated during the strain-induced martensite transformation. Because of the accompanying volume change, the shear strain leads to an additional increase of the dislocation density. Further straining leads to the effect that the fresh formed strain-induced martensite generates more geometrically necessary dislocations. Since the amount of dislocations generated depends on the austenite fraction transformed to martensite, work hardening of TRIP steels can be adapted over a wide range of strains. [11] In analogy to this, the formation of additional mobile dislocations by the TRIP effect further improves work hardening of TRIP steels, which is essential for an improvement of both strength and ductility.

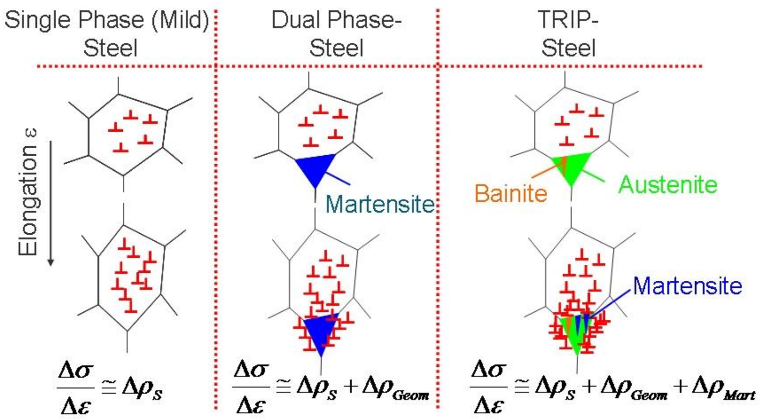

Fig. 4 *Schematic explaining the evolution of dislocation density in mild, DP and TRIP steels. [11]*

3 TBF and Q&P steels

The microstructure of conventional TRIP steels from the 1-GEN AHSS consists of a soft ferritic matrix and hard bainitic and retained austenitic inclusions, whereby the latter transforms to high C strain-induced martensite upon applied deformation. This leads to the formation of internal stresses at the interphases between soft matrix and hard inclusions, leading to the initiation and propagation of micro-cracks. In other words, such a material will benefit from an extensive work hardening and excellent performance during forming operations with global deformation such as deep-drawing, but it will crack severely when it comes to the forming operations with localized deformation such as cutting, hole expansion etc. One possible way, how to mitigate this issue, is to strengthen the ferritic matrix by micro-alloying, i.e. by the addition of Ti, Nb, V or their combination. [12-14] Another possibility, how to improve the steel performance during forming with localized deformation, is to replace the soft ferritic matrix with a harder one from either bainitic ferrite by TBF steels or tempered martensite/lower bainite by Q&P steels. [15, 16]

Fig. 5 presents the scanning electron microscopy (SEM) micrographs of the microstructure in case of TBF and Q&P steels. The matrix of TBF steel is of bainitic ferrite while the one of Q&P steel consists of large quantities of tempered martensite and some lower bainite, respectively. In both cases, the microstructure contains metastable retained austenite, which will transform to strain-induced martensite during sheet forming. Obviously, hard matrix will reduce the build-up of internal stresses in the microstructure and this phenomenon will be more pronounced in case of Q&P steels with a stronger matrix compared to TBF grades. In addition, the strain-induced martensitic transformation will enhance deep drawability in both steel grades by a sufficient exploitation of the TRIP effect.

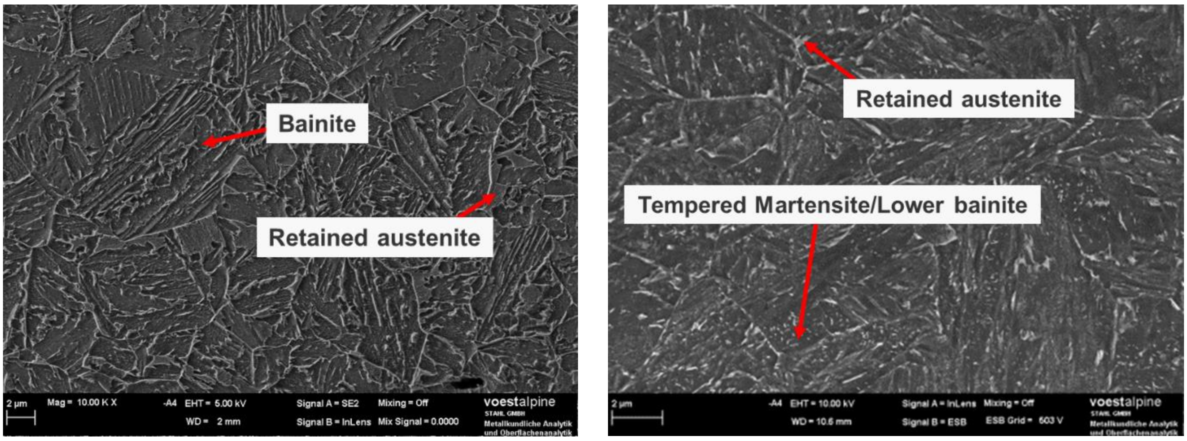

***Fig. 5** SEM micrographs of TBF steel (left) and Q&P steel (right).*

Fig. 6 shows the thermal cycles applied to adjust the final microstructure and mechanical properties of TBF and Q&P steels. In the heating stage of as cold rolled material, the static recrystallization of microstructure and carbide dissolution takes place. The steel is afterwards subjected to full austenitization, meaning that the material must be soaked at temperatures higher than Ac3. Intercritical annealing can also be applied to both steel concepts, but it shrinks their processing window and deteriorates the mechanical properties by a decrease of tensile strength and hole expansion. In the soaking region, the structure can be partially or fully homogenized with respect to its C and Mn content. In general manner, after the soaking stage the material is cooled down to a specific temperature. TBF steels are cooled to the overageing temperature (T_{OA} usually about 400°C), at which the isothermal bainitic transformation takes place. The bainite is supposed to be carbide-free and thus the steel composition should contain Si and/or Al in order to prevent the carbide formation in the retained austenite. By doing so, during the propagation of the bainitic transformation, the retained austenite can be enriched in C and its grain size should be decreased in such a matter that it can be sufficiently stabilized after final cooling to RT. In case of Q&P steels, the material is quenched under the Ms temperature to the well-defined quenching temperature T_Q, where an exact, rather high amount of martensite and retained austenite can be adjusted (Fig. 7). At too low T_Q, the formation of too high initial martensite content will decrease the final amount of retained austenite and at too high T_Q, the amount of initial austenite is too high, resulting in the insufficient stabilization of the retained austenite and martensite formation during final cooling. More detailed information about the model calculation for Q&P steels can be found in [17]. By heating the steel up to the partitioning temperature TP, the C partitioning from marteniste to austenite accompanied with martensite tempering and some minor bainite formation takes place. In analogy to TBF steels, a sufficient C partitioning from martensite to austenite will stabilize the retained austenite at RT.

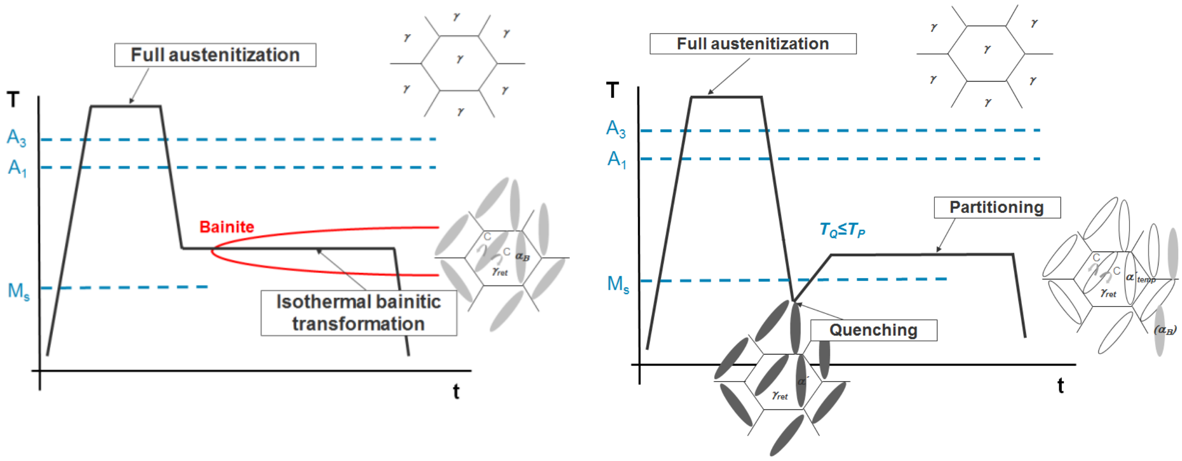

Fig. 6 *Thermal cycle applied to adjust microstructure and mechanical properties of TBF steels (left) and Q&P steels (right).*

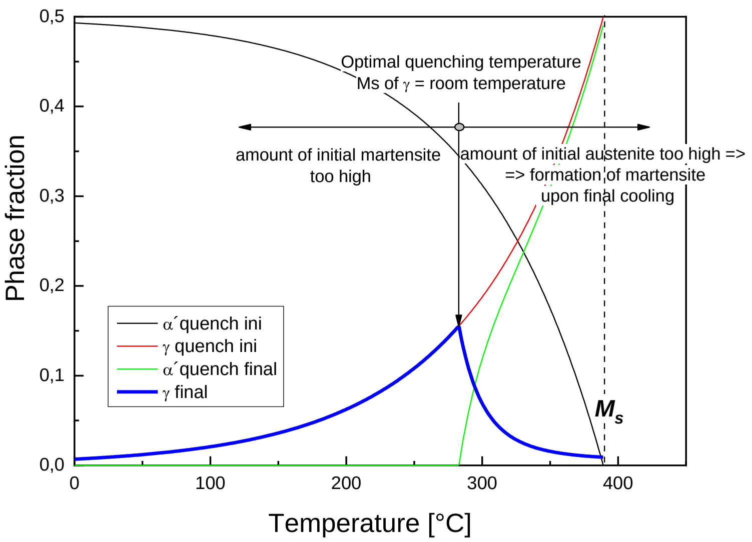

Fig. 7 *Model calculation to establish the optimal quenching temperature with the highest retained austenite content for an intercritically annealed Q&P steel.*

In general, the minimum tensile strength levels for both TBF and Q&P steels are 980MPa and 1180MPa, respectively. Fig. 8 depicts the spider diagrams of the mechanical properties for the industrially produced TBF and Q&P steels with a minimum tensile strength of 980MPa. The mechanical properties were obtained from tensile testing using the tensile specimens with a gauge length of 80mm measured in longitudinal and transversal direction. On the one hand, it can be seen that TBF steel of the same strength level is characterized by a lower yield strength and higher total elongation due to a softer matrix compared to Q&P steel. This means that TBF steel is expected to perform slightly better in the forming operations with global deformation. On the other hand, annealing at lower temperatures triggers the formation of harder microstructural compounds in the matrix, resulting in an improvement of hole expansion λ [18], one of the measures representing the performance of a steel in the forming operations with localized deformation (Fig. 9). In other words, it is expected that Q&P steels will have a slightly better performance in the forming operations with localized deformation. It has to be however pointed out that both steels have an extraordinary combination of formability upon global as well as local deformation, which makes them particularly suitable for forming of very complex structural parts as for example complicated seat structures. Nowadays, the first applications of TBF and Q&P steels for anti-intrusion structural parts and seat components can be found in emerging automobile platforms.

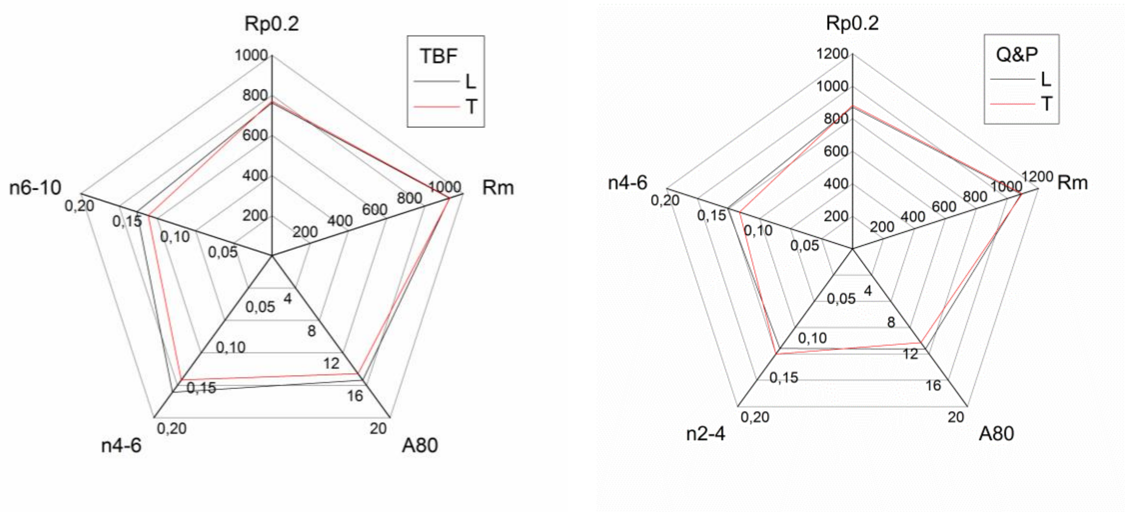

Fig. 8 *Mechanical properties of industrially annealed TBF and Q&P steels in longitudinal (L) and transversal (T) direction. [19]*

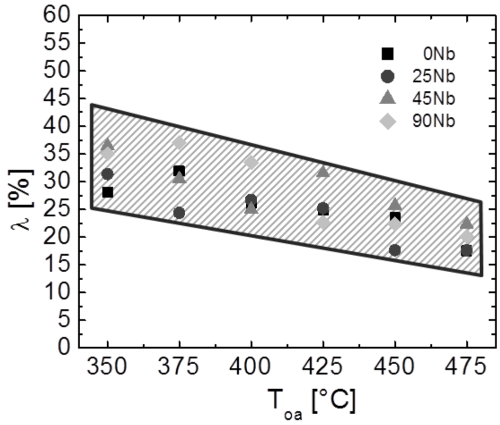

Fig. 9 *Hole expansion ratio λ as a function of T_{OA} for micro-alloyed TBF steels. [14]*

4 Medium Mn steels

First research work concerning the development of medium Mn steels returns to R. L. Miller [20] in 1972. In his work, he suggested intercritical annealing of high alloyed Ni (9-21 %) steels and a 6 %-Mn steel for holding times within the range of hours to obtain remarkable mechanical properties. At that time, this alloying concept did not meet the requests of the industry. In 2007 M. J. Merwin [21] readopted the alloying design and proposed a simple C-Mn chemical composition containing 0.1 %C and 5-7 %Mn for intercritical annealing in batch furnaces. Compared to the 1970's, the requirements of the automotive industry have radically changed and the alloying concept of medium Mn steels has currently found rising attention in both academic and industrial research. The alloying concept is proposed as one of the most promising candidates to fulfil the property requirements of the 3-GEN AHSS. Up to now, concepts for a production route via batch annealing (BA), which involves annealing an entire coil in a large furnace for several hours, and continuous annealing lines (CAL), where the typical annealing times is within the range of minutes, exist. Making use of the combination of both annealing processes has been recently gaining more importance.

Medium Mn TRIP steels contain from 3-12 wt-% Mn, but more typically from 5-7 wt-% Mn. Their microstructure consists of ultrafine-grained ferritic matrix with a grain size typically less than 1µm with a high volume fraction of retained austenite, usually up to 30 vol.-%. Tensile strength of these steels commonly exceeds 1000 MPa along with total elongations in a range of 25 – 40 %. Thus, the combination of tensile strength and total elongation in case of medium Mn steels can easily exceed the margin of 30000 MPa%. Medium Mn steels have received a lot of attention over the last few years not only due to their remarkable mechanical properties compared to the 1-GEN AHSS but also because of their reasonable alloying costs and less challenging production compared to the high Mn 2-GEN AHSS.

The main heat treatment step in the producing of medium Mn steels is intercritical annealing (soaking between A_{e1} and A_{e3}). Depending on which production route is chosen, the holding time at T_{IA} ranges between several minutes (continuous annealing) or several hours (batch annealing). In both cases, it has to be ensured that the intercritically formed austenite is stabilized to RT. Therefore, a sufficient chemical (satisfactory high amount of austenite former elements C and Mn in the retained austenite) as well as mechanical stabilization (sufficiently low grain size of the retained austenite) become essential.

Fig. 10 displays the time-temperature schedules recently used to anneal medium Mn steel, showing that besides the one-step intercritical annealing, there is one further existing heat treatment (HT) concept, which includes a full austenitization prior to the intercritical HT. This HT routine is commonly denominated as two-step HT. Furthermore, Fig. 10 reveals the microstructural development during the HT. The initial microstructure for the one-step HT consists of heavily cold worked martensite (and bainite). Depending on the selected T_{IA}, this deformed microstructure recrystallizes during heating or intercritical annealing. Corresponding to the applied T_{IA}, a certain amount of austenite is formed and enriched in C and Mn. By sufficient stabilization of the austenite, this HT results in a globular shaped ferritic/austenitic microstructure at RT. The two-step HT additionally consists of a full austenitization prior to intercritical annealing in order to establish an undeformed martensitic microstructure. The same effect can be achieved by providing a sufficient cooling rate after the hot rolling process followed by intercritical annealing of hot band in a batch annealing furnace. This martensitic microstructure is then substantially tempered during the heating and soaking stage. Alike to the one-step HT, austenite is formed and enriched in Mn and C. Finally, the resulting microstructure at RT differs mainly in its shape, which is predominantly lamellar.

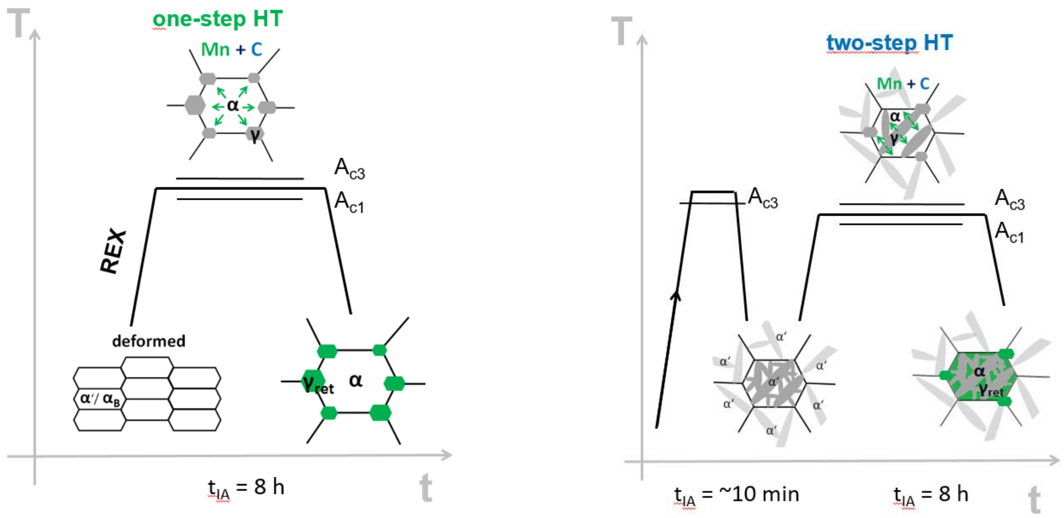

Fig. 10 *Time-temperature schedule and microstructural development during one-step HT (left) and two-step HT (right) of a medium Mn steel. [22]*

Fig. 11 shows the scanning transmission electron microscopy (STEM) micrographs of the ultrafine-grained microstructure in 0.1C6.4Mn steel subjected to the one-step and two-step HT, reflecting the fact that one-step HT leads to the globular morphology, whereas the microstructure after the two-step HT is dominantly lath-like. Moreover, from the Mn distribution within the microstructure can be clearly seen that Mn remarkably partitions from ferrite to austenite during the intercritical annealing step.

Fig. 12 presents the equilibrium thermodynamic calculation of the microstructural evolution in 0.1C6Mn steel as a function of T_{IA}. More detailed explanation of the model calculation can be found elsewhere. [23] It is evident that similarly to Q&P steels, there exists a specific theoretical temperature, where the maximum retained austenite can be persisted after cooling to RT. This temperature represents the status, at which the Ms-temperature of the intercritical austenite reaches RT. In other words, after cooling from this T_{IA}, no martensite will be form upon final cooling to RT. At T_{IA} higher than optimal one, the grain size of the intercritical austenite is too large and its C and Mn content too low, which causes its insufficient stabilization. As a result, martensite forms during final cooling, which impairs the mechanical properties of medium Mn steels. In opposite, at T_{IA} lower than the optimal one, the cementite is still undissolved (less C and Mn available for chemical stabilization of the retained austenite) and a lower amount of the intercritical austenite can be expected based on thermodynamic calculations. Thereby the combination of strength and ductility cannot reach the maximum value obtained at the predicted optimal T_{IA}. It should also be taken in consideration that the region, where the optimal microstructure and resulting mechanical properties can be achieved, is very narrow. Therefore, T_{IA} must be carefully adjusted and controlled during intercritical annealing, in order to guarantee the best combination of mechanical properties in the final product.

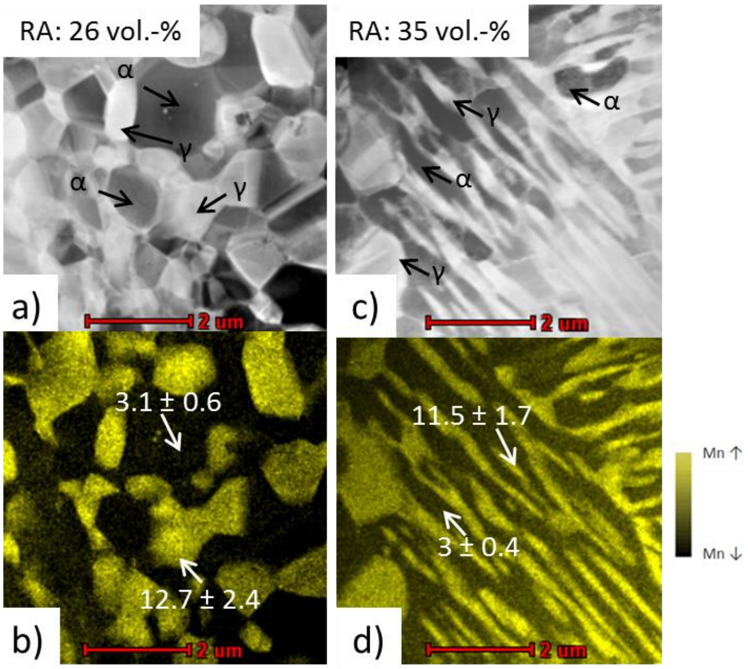

Fig. 11 *STEM micrographs and Mn distribution of a) and b) one-step heat-treated T_{IA} = 620 °C and c) and d) two-step heat-treated T_{IA} = 640 °C (numbers represent Mn-content (wt-%) determined by at least six spot analysis) (α = ferrite, γ = austenite). [24]*

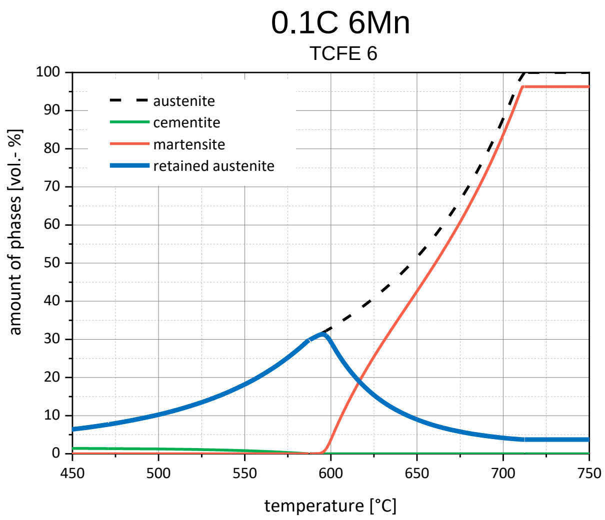

Fig. 12 *Thermodynamic calculation of microstructural evolution in medium Mn steel as a function of T_{IA}.[25]*

Fig. 13 shows the experimentally obtained volume fractions of the retained austenite for the one-step and two-step annealing measured by saturation magnetization. This magnetic method is thoroughly described in [26]. It can be claimed that the evolution of the retained austenite content is irrespective of the applied HT. It is also worthwhile to mention that the optimal temperature, at which the maximum retained austenite content is achieved, is shifted towards higher T_{IA} compared to the model calculations. This implies that the current model only serves for the first rough estimation of the microstructural evolution with T_{IA} and its further improvement is pre-requisite. Some research effort has already been carried out in order to solve this issue. This relates for example to the exact determination of the grain size of the intercritical austenite prior to final cooling in order to determine the M_s temperature with a higher accuracy. [27] In this context, a smaller grain size of the intercritical austenite will decrease the M_s temperature and therefore the optimal T_{IA} will be shifted to higher temperatures compared to the first model predictions. In analogy to this, the first attempts to optimize the M_s temperature formula suitable for medium Mn steels have also been put forward to the scientific community. [28]

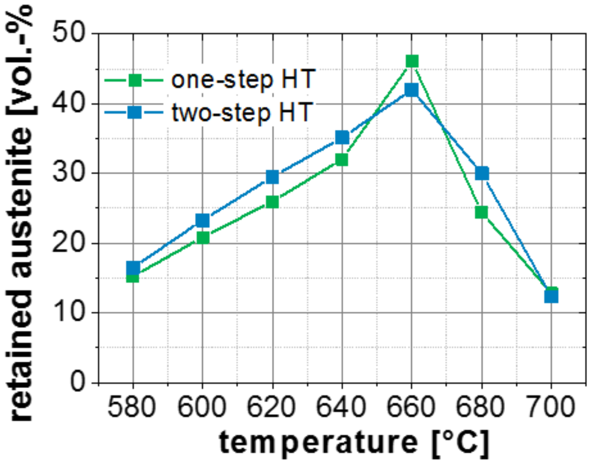

Fig. 13 *Retained austenite content as a function of T_{IA} for 0.1C6Mn steel measured by saturation magnetization. [29]*

Fig. 14 displays the mechanical properties of 0.1C6Mn steel, prepared by laboratory casting and processing, as a function of T_{IA}. After hot and cold rolling, the steel was subjected to final annealing using the one-step and two-step HT. It can be seen that the best combination of mechanical properties for both HT regimes were obtained at slightly lower temperature than the theoretical optimal T_{IA} with the highest retained austenite content. According to [29], this is due to the fact that the most optimal stability of retained austenite is in practice obtained slightly below this theoretically optimal temperature. Given that, the retained austenite is in reality slightly under-stabilized in the situation when the maximum amount of retained austenite is present in the microstructure. Overall evolution of mechanical properties is however in analogy to the prediction made by the model calculations. At the higher T_{IA}, the deterioration of mechanical properties comes into play for both HT routines due to the martensite formation upon final cooling. Accordingly, at the lower T_{IA}, lower amount of the retained austenite with a higher stability reduces the work hardening and therefore the loss of both strength and elongation can be expected. It is noticeable that in the lower T_{IA} range (T_{IA}=580-600°C) the more pronounced yield point elongation (YPE) can be observed and this is more remarkable for the steel subjected to the one-step HT compared to the two-step one. Evidently, at T_{IA} of 580°C, even an ideally plastic stress-strain curve was obtained. Such a behavior is unacceptable during forming operations of automotive parts, since the detrimental stretcher-strainer marks formation occurs at the surface of a component, deteriorating both its surface appearance and mechanical properties. Therefore, this phenomenon must be avoided, while forming the parts in the automotive industry. K. Steineder *et al.* [24] investigated this problem in detail. The pronounced YPE is a result of the lack of work hardening as a consequence of the hyper-stable retained austenite and ultra-fine grained ferritic matrix, which does not allow for the accommodation of dislocation cells and thus the plastic deformation starts to localize in the form of Lüders bands, traversing from one shoulder of tensile sample to another. This problem can be overcome by the employment of the two-step HT, resulting in the microstructure with lamellar morphology. In such a matrix, the accommodation of dislocation cells is allowed and mobile dislocations can move in the microstructure during straining. This promotes the significant work hardening of medium Mn steel and the pronounced YPE can be diminished or even completely eliminated. This effect is clearly observable in Fig. 14, where the stress-strain curves of the two-step heat-treated material possess much lower YPE compared to the one-step heat-treated samples. Moreover, retained austenite stability plays a crucial role in the reduction of YPE: the lower retained austenite stability the lower YPE. This is due to the formation of additional mobile dislocations in the system, promoting homogeneous deformation rather than its localization in the form of Lüders bands. [22]

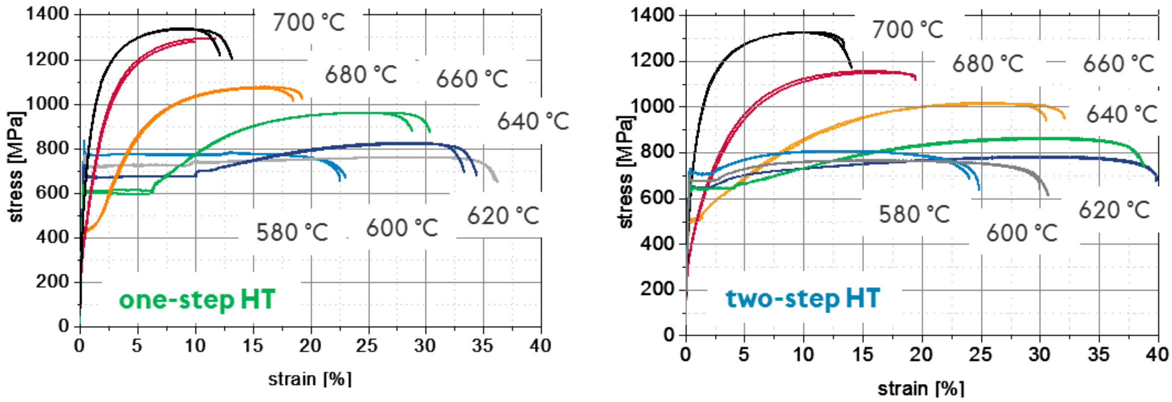

Fig. 14 *Mechanical properties of the 0.1C6Mn steel subjected to the one-step (left) and two-step (right) HT. [24, 29, 30]*

By further analysis of the stress-strain curves, it can be noticed that in both cases, the grade with a tensile strength level of 780 MPa with a total elongation exceeding 30 % could be achieved. However, a higher total elongation could be reached for the two-step HT material due to the better exploitation of the TRIP effect, as already predicted in one of the previous paragraphs. This is even more accentuated, while analyzing the product of tensile strength and total elongation for both HT regimes (Fig. 15). In case of one-step HT, the recommended margin of 30000 MPa% could not be achieved, whereas for the two-step HT, a value of almost 35000 MPa% could be easily tackled.

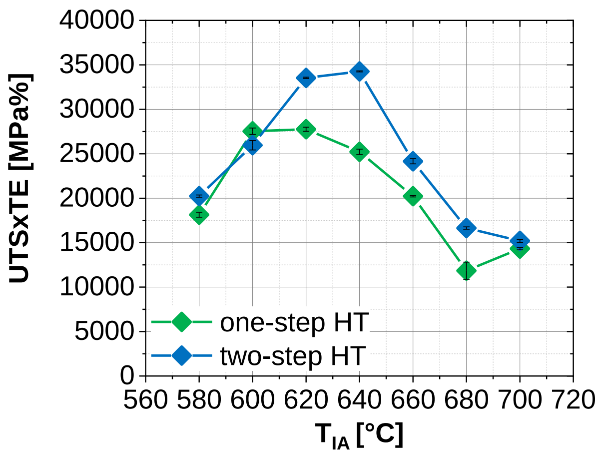

Fig. 15 *Product of tensile strength and total elongation for the 0.1C6Mn steel subjected to the one-step and two-step HT. [24, 29]*

Based on the findings from the laboratory development described above, the first industrial trials using a conventional LD convertor route available at the production site in Linz, could be performed. After industrial casting, hot and cold rolling, the two-step HT was applied to the 0.1C6Mn material, whereby the first HT step took place in an industrial CAL and the final second step was employed in an industrial BA furnace operating under hydrogen protective atmosphere. The stress-strain curve of the first industrially produced 0.1C6Mn steel is depicted in Fig. 16. Similarly to the laboratory results, the 780MPa steel grade with a total elongation of more than 35 %, measured using tensile specimen with a gauge length of 80mm in longitudinal direction, could be achieved on the industrial scale. The hole expansion ratio λ of approximately 30 % in punched condition is satisfactory for the BIW applications in the automotive industry. Moreover, almost YPE-free steel could be manufactured using the two-step HT even prior to final skin passing. The industrial material will be tested with respect to its formability, weldability and hydrogen embrittlement to evaluate its response in terms of in-service properties. In addition, further development will focus on the industrial trails of medium Mn steels with a tensile strength of 980MPa in order to satisfy the customer demands from the automotive sector.

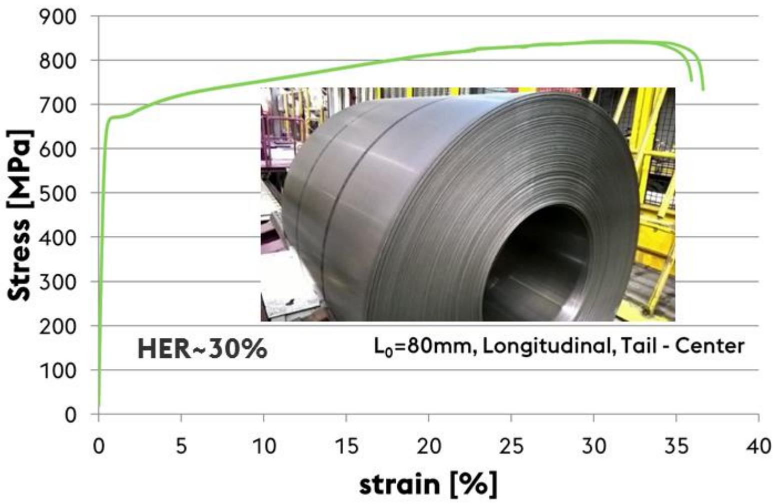

Fig. 16 *Stress-strain curve of the first industrially produced 0.1C6Mn steel using a conventional LD convertor-processing route available at the production site in Linz. [30]*

5 Delta TRIP steels

A further development of new steel grades are δ-TRIP steels, which refer to density-reduced steels due to their increased Al content. In comparison to conventional TRIP-assisted steels, the usual allotriomorphic ferrite grains are partially or fully replaced by dendritic δ-ferrite in δ-TRIP steels. The δ-ferrite grows from the liquid phase and persist at all temperatures down to RT. It cannot be eliminated by any HT because of its thermodynamic stability. The remaining microstructure of δ-TRIP steels is comparable with that of conventional TRIP-assisted steels. [31-33]

δ-TRIP steels have a typical chemical composition of 0.3 - 0.4 wt-% C, 2 - 6 wt-% Al, 0.2 - 0.8 wt-% Si and 0.5 - 1.6 wt-% Mn. [31] Due to their high Al content, a density reduction of up to approximately 1.5 % per wt-% Al becomes possible. Since these steels undergo the TRIP effect, they have excellent mechanical properties with tensile strengths of around 600 to 800 MPa at elongations above 30%. [34] Therefore, δ-TRIP steels are well suited for the lightweight application in the automotive industry, especially for safety related parts, where excellent deep drawability is required. [31] Nevertheless, their application for high strength density reduced exposed panels might also come into consideration in the future.

As already mentioned afore-hand, besides the opportunity to save weight due to the application of high strength steels, a weight reduction can also be achieved using density reduced steels. One possibility for this density reduction is the addition of alloying elements like Al. Al has impact on the steel's density because of two different factors. On the one hand, the weight reduction is achieved due to the replacement of pure Fe by this lighter substitutional element. On the other hand, the application of this element results in an additional lattice expansion. This is because the atomic radii of Al is larger than that of Fe, resulting in a significant increase of the lattice parameter. The larger lattice parameter leads to a decrease of the density because of an additional increase of the specific volume. G. Frommeyer and G. Brüx [35] described a linear relation between the Al content in steel and its density with the following formula:

$$\rho = -0.115 * Al[wt.\%] + 7.86 \qquad (1)$$

Fig. 17 represents the relation between the Al content and the density in steels, whereby with increasing Al content the density is linearly reduced. The arrows on the right-hand side indicate the two different factors influencing the steel's density – the lighter elements and the lattice expansion, respectively.

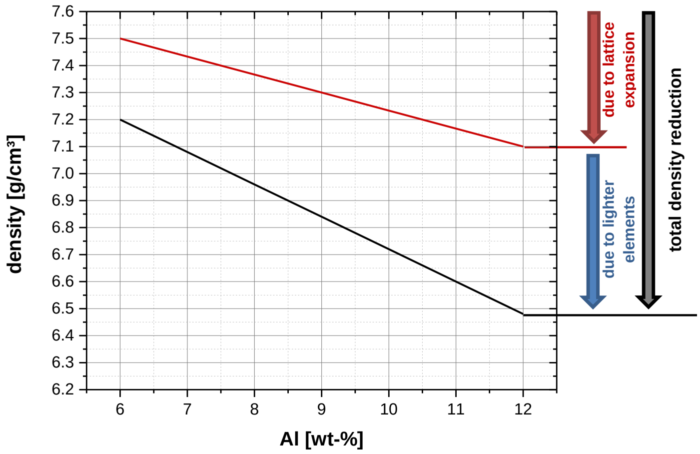

Fig. 17 *Density depending on the Al-content of FeMnAlC steel concepts. [31]*

As described in the literature [31], the Young's modulus is depending on the material's density. As the steel's density declines with increasing Al content, the Young's modulus decreases, as well. While the reduction in density is beneficial for automotive applications, the decrease of the Young's modulus is evidently adverse. The higher the Young's modulus, the higher is the material and component stiffness and crash resistance. For this reason, Si or Mn are usually added to Fe-Al alloys for increasing the Young's modulus.

The relation between the density and the Young's modulus is described with the following formula:

$$E = c_L^2 * \rho * \frac{(1-2\nu) * (1+\nu)}{1-\nu} \qquad (2)$$

Here, E is the Young's modulus in MPa, c_L is the sound velocity in m/s, ρ is the density in kg/m³ and ν is the Poisson's ratio.

The δ-TRIP alloy system is based on the use of a relatively large concentration of Al, which should lead to the formation of a substantial amount of ferrite dendrites at equilibrium. Large volume fractions of δ-ferrite, which forms during solidification, should resist the transformation into austenite at temperatures over approximately 900-1200°C (Fig. 18 left). Due to the final HT, austenite formed during intercritical annealing partially transforms to bainitic ferrite in the over-aging section, usually resulting in a typical final TRIP microstructure consisting of α- or δ-ferrite, bainitic ferrite and retained austenite (Fig. 18 right). In some cases, fresh martensite can be present in the microstructure as a consequence of insufficient stabilization of retained austenite. Since fresh martensite deteriorates hole expansion properties, such microstructures should be rather avoided.

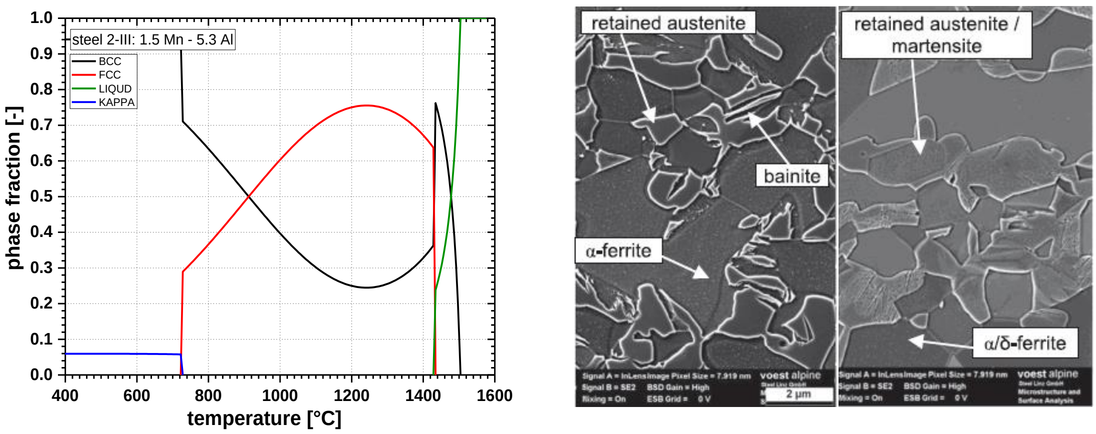

Fig. 18 *Thermodynamic calculations of phase evolution (left) and SEM micrograph of typical microstructures (right) of density reduced TRIP steels. [36]*

The temperature of intercritical annealing must be high enough in order to dissolve detrimental κ-carbides formed during previous processing. Keeping this important recommendation in mind, an excellent combination of mechanical properties of these steels is reachable. Whereas alloying elements like Si, Al and P retard the precipitation of cementite from untransformed austenite during bainite formation, the diffusion of C from the bainitic ferrite into untransformed austenite improves its stability and enables the retention of austenite at RT. [37] The typical heat treatment cycle consisting of intercritical annealing and the over-ageing step in a temperature range of 350°C - 450°C, which can be applied in a number of industrial CALs, is illustrated in Fig. 19.

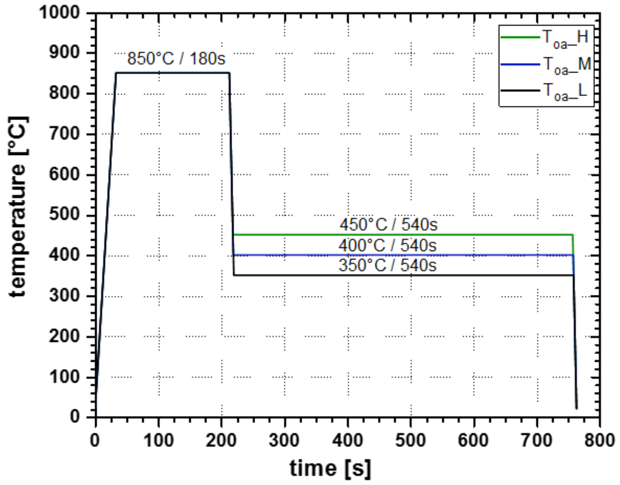

Fig. 19 *Thermal cycle for HT of density-reduced TRIP steels. [3]*

Fig. 20 represents the mechanical properties of the density-reduced TRIP steels with a varying chemical composition, especially Al. The best combination of mechanical properties was achieved for the steels with a lower Al content of approximately 3 - 3.5 wt-%. Tensile strength of above 700 MPa and total elongation higher than 40%, measured by means of small tensile samples with a gauge length of 25mm, are excellent for this steel grade. Bearing in mind a density reduction exceeding 4 %, these steels can be an attractive candidate for various automotive lightweight applications in the near future. First, to accomplish this, some issues related to the complicated castability and weldability of high Al containing steels must be however completely solved.

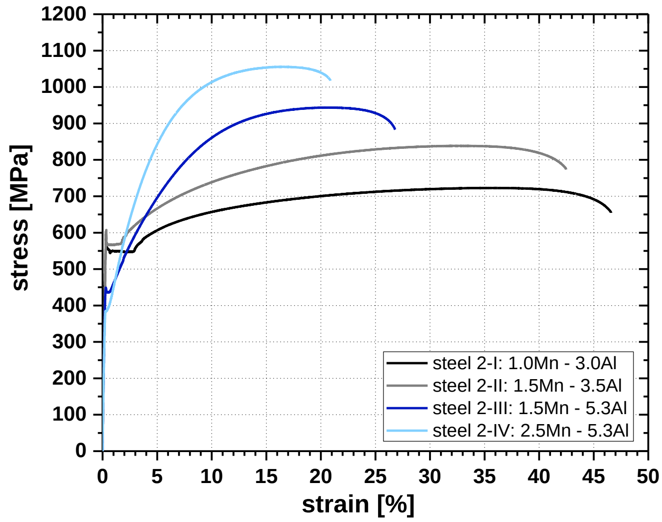

Fig. 20 *Mechanical properties of various density-reduced TRIP steel with different chemical composition. [36]*

6 Conclusions

The present paper was aimed to describe the state-of-the-research in the family of the 3-GEN AHSS for the automotive industry, namely TBF, Q&P, medium Mn and density-reduced TRIP steels. Vast developments have been carried out by both industry and academia in order to fulfil the challenging requirements of these steel grades with respect to their mechanical properties. Apart from strength increase, density reduction has also been introduced to reduce the overall weight of modern automotive platforms even further.

Based on the facts presented in this contribution, it is apparent that the microstructure of all candidates for this steel group must contain metastable retained austenite, transforming to martensite during straining. This is the easiest way to achieve the requested mechanical properties and therefore, among others (solid solution strengthening, precipitation strengthening, grain refinement and structural strengthening) the TRIP effect is the paramount mechanism activating the vital strengthening in the 3-GEN AHSS.

Whereas TBF and Q&P steels have already been incorporated in the recent car body structures, medium Mn steels and density-reduced TRIP steels are currently under development. Their application can be feasible in the near future, when some issues regarding their processability and in-service properties will be successfully overcome.

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