



Advanced high strength steels for automotive industry

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The aim of this paper is to present the basic concepts of advanced high strength steels (AHSS) for use in the automobile industry, including chemical composition design, microstructure and mechanical properties development during thermomechanical processing, production technology characterisation, potential applications and performance in service. AHSS steels are considered to be the major materials for future applications in this production sector. As opposed to the cold formable single phase deep-drawable grades, the mechanical properties of AHSS steels are controlled by many factors, including: phase composition and distribution in the overall microstructure, volume fraction, size and morphology of phase constituents, as well as stability of metastable constituents. The main feature of these steels is that they do not permit to rely on the well-established traditional microstructure-properties relationships. Therefore, many different alloy concepts and alternative processing routes are still under development by different steel producers for comparable steel grades.

Keywords: advanced high strength steels, automobile industry, microstructure, mechanical properties, thermomechanical treatment, weldability, crash worthiness

1. Introduction

Over the last decade, a strong competition between steel and low density metal industries has been observed as a result of increasing requirements of passenger safety, vehicle performance and fuel economy. The response of steel industry to the new challenges is a rapid development of higher strength steels, named Advanced High Strength Steels (AHSS) [1]. These steels are characterised by improved formability and crash worthiness compared to conventional steel grades. The category of AHSS covers the following generic types: dual phase (DP), transformation induced plasticity (TRIP), complex phase (CP) and martensitic steels (MART).

The AHSS may be distinguished based upon the strength properties that roughly can be defined: yield strength > 300 MPa and tensile strength > 600 MPa. As opposed to the conventional high strength steels, in which ductility decreases with strength, modern AHSS steels combine high strength and formability/ductility. General classification of these steels is as follows:

- High strength steels with a high energy absorption potential (DP and TRIP steels with UTS < 1000 MPa), for dynamic loading occurring during car crashes or collisions.

- Extremely high strength steels, typically martensitic steels, with a very high UTS (>1200 MPa), providing high stiffness, anti-intrusion, load-transferring barriers for the protection of automotive passengers.

• The rationales for increased use of the AHSS in the automotive industry are as follows:

- The reduction of the car weight resulting from the use of high strength thinner gauge sheet steel, reducing the fuel consumption.
- Increased passenger safety by an improved crash worthiness.
- The strong competition from the light-weight materials, such as Al and Mg alloys and plastics.

AHSS derive their properties from multi-phase complex microstructure. Since these steels are relatively new, their classification differs from conventional high strength steels and was developed by Ultra-Light Steel Automotive Body – Advanced Vehicle Concept (ULSAB-AVC) Consortium [1]. The accepted practice involves specification of both yield strength (YS) and ultimate tensile strength (UTS) in the following way:

XX aaa/bbb,

where *XX* is type of steel, *aaa* is minimum YS in MPa and *bbb* is minimum UTS in MPa.

For example DP 280/600 designates dual phase steel with 280 MPa minimum yield strength and 600 MPa minimum ultimate tensile strength.

2. Microstructure – mechanical properties characterisation

2.1. Dual phase steels

Microstructure of dual phase steels is composed of soft ferrite matrix and 10–40% of hard martensite or matrensite-austenite (M–A) particles (Figure 1). This type of microstructure allows achieving the ultimate tensile strength in the range of 500–1200 MPa. When the volume fraction of martensite exceeds 20%, DP steels are often called partial martensitic. For some applications, also bainitic constituent may be desirable in the DP steel microstructure. This dual phase type of microstructure can exhibit the following advantageous features over the conventional high strength steels:

- The strength of the DP steel microstructure is controlled by the amount of martensite and ductility by the size and distribution of this phase,
- DP steels do not exhibit yield point elongation,
- DP steels posses low UTS/YS ratio (around 0.5) and high strain hardening characteristics (high *n* value), especially at the beginning of plastic deformation,
- They can be strengthened by static or dynamic strain ageing (BH effect),

- Grades containing low carbon content have been shown to exhibit excellent resistance to fatigue crack propagation at growth rates close to fatigue threshold intensity range ΔK_{th} .

Very important for the development of DP steel is the effect of carbon and alloying elements, which was summarized in Table 1.

Table 1. Effect of alloying elements in DP steels

Alloying element	Effect and reason of adding
C (0.06–0.15%)	<ul style="list-style-type: none"> ▪ Austenite stabilizer ▪ Strengthens martensite ▪ Determines the phase distribution
Mn (1.5–2.5%)	<ul style="list-style-type: none"> ▪ Austenite stabilizer ▪ Solid solution strengtheners of ferrite ▪ Retards ferrite formation
Si	<ul style="list-style-type: none"> ▪ Promotes ferritic transformation
Cr, Mo (up to 0.4%)	<ul style="list-style-type: none"> ▪ Austenite stabilizers ▪ Retards pearlite and bainite formation
V (up to 0.06%)	<ul style="list-style-type: none"> ▪ Austenite stabilizer ▪ Precipitation strengtheners ▪ Refines microstructure
Nb (up to 0.04%)	<ul style="list-style-type: none"> ▪ Austenite stabiliser ▪ Reduces M_s temperature ▪ Refines microstructure and promotes ferrite transformation from non-recrystallized austenite

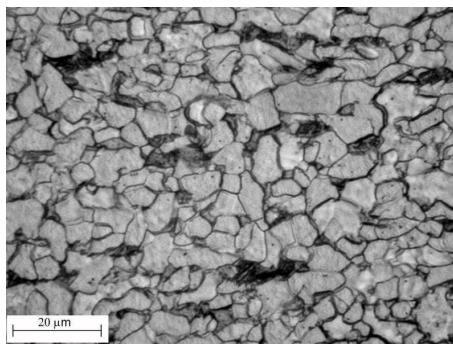


Fig. 1. Photomicrograph of the dual-phase steel microstructure

The microstructure of DP steels do not allow obtaining of high plastic strain ratio value (r_m), which means that these steels are not good candidates for applications that require high drawability. They usually exhibit poor hole expansion ratio values. This drawback, however, can be eliminated by adding Ti with the aim of inducing the precipitation strengthening in ferrite to reduce the differences in hardness between the two phases. Alternatively, M-A constituents may be replaced by bainitic phase. Dual

phase steels can be welded with all conventional welding methods currently used in the automotive industry (resistance spot welding, laser welding, arc welding).

The most important features influencing mechanical properties of the dual-phase microstructure comprise shape, size, amount and distribution of ferrite and martensite, the carbon content of martensite, and the volume fraction of retained austenite.

In DP steels, for a fixed volume fraction of M-A constituents, both applied stress and the work hardening rate at a given true strain can be related to the average size of M-A particles by the Hall-Petch type equation. Frequently cited expressions, for $f_{MA} = 0.2$, were developed by Lanzilotto and Pickering [2]:

$$\sigma_{f(\varepsilon=0.2)} = 350 + 18.1\lambda^{-0.5}, \quad (1)$$

$$\frac{d\sigma}{d\varepsilon_{(\varepsilon=0.2)}} = 40.1\lambda^{-0.5}, \quad (2)$$

where λ is the average size or diameter of M-A particles.

The coefficients of Equations (1) and (2) suggest that $d\sigma/d\varepsilon$ increases more rapidly than σ_f . This suggests that refinement of the M-A constituent in DP steels should result in an increase in uniform elongation. The variation of stress as a function of martensite content in DP steels has been frequently modelled on the basis of the rule of mixtures, applying the continuum mechanics models [3]:

$$\sigma = V_m \sigma_m + (1 - V_m) \sigma_f, \quad (3)$$

where σ_m , σ_f , σ are the stresses of the martensite, ferrite and composite structure respectively.

The dependence of stress on strain in composite and components is modelled with the following constitutive equation:

$$\sigma = k \varepsilon^n. \quad (4)$$

The rule of mixtures can be applied to either isostrain or isostress conditions [4]. Equation (3) is frequently applied with the assumption that σ_m and σ_f are invariant with respect to the volume fraction and morphology of the respective phases. However, most of the experimental results show that this approach is obeyed for martensite volume fraction of up to approximately 0.5. Chang and Preban proposed the model capable of explaining the variation between yield strength and volume fraction of martensite [5]. In their model, yield stress in DP steels is assumed to be influenced by the mean free path in ferrite, λ_f , through a Hall-Petch type relationship:

$$\sigma_y = \sigma_{0y} + K_y \lambda_f^{-0.5}, \quad (5)$$

where σ_{0y} is a frictional stress of the matrix, K_y is the dislocation-locking constant. In the model, both σ_{0y} and K_y are functions of V_m .

Bag et al. [6] proposed the new approach that allowed the prediction of yield stress using the following equation:

$$\log(\sigma_y) = 2.8565 - 0.25441 \log(\lambda_f). \quad (6)$$

Equation (6) can be used for the prediction of yield stress over a wide range of V_m parameter and is independent of this parameter. This is connected to the fact that the mean free path in ferrite is dependent on the martensite content.

2.2. TRIP steels

Advanced high-strength transformation-induced plasticity (TRIP) steels are well suited for light-weighting car body construction with added advantage to reduce the safety problems. The possible development of this type of steels was first discovered by Zackay et al. [7]. They proposed that the strain or stress induced transformation of retained austenite present in the microstructure in a sufficient amount can substantially harden the steel during deformation, and therefore results in a higher ductility. The mechanical properties of TRIP steels are derived from their disperse multi-phase microstructure which is composed of ferrite – α (0.50–0.55), bainite (0.30–0.35), retained austenite – γ_R (0.07–0.15), and possibly martensite (0.01–0.05). Retained austenite is the most important phase constituent of TRIP steels. An example of the EBSD map of this phase in TRIP steel is given in Figure 2. During deformation, retained austenite transforms to martensite. The TRIP mechanism in steels with dispersed austenite is shown schematically in Figure 3. The crucial technological challenge concerning TRIP steels lies in the necessity of producing in their structure of sufficiently high content of very stable austenite.

The TRIP steels are characterised by a relatively low content of alloying elements. For example, in current 800 MPa TRIP steels, the total content of alloying elements is about 3.5 wt.%. Thus, the selection of suitable alloying elements and the amount required to produce the intended properties is a critical stage in the alloy design stage. Conventional TRIP steel compositions concept proposed by Matsumura, is based upon the following C, Mn and Si content [8]:

C: 0.12 – 0.55 wt.%; Mn: 0.20 – 2.5 wt.%; Si: 0.40 – 1.8 wt%.

The carbon content plays a fundamental role in TRIP steel composition design, since its distribution among the main microstructural constituents is fundamental to

the material properties. To obtain the best mechanical properties in a product, first of all, carbon should distribute to austenite and enrich this phase as much as possible to preserve the M_s^σ of this constituent of 15–25°C below the room temperature. Carbon content in the current TRIP steels is limited to 0.20–0.25 wt.% due to the weldability concerns. For the hardenability reasons, the Mn content in low alloy TRIP steels is around 1.5 wt.%.

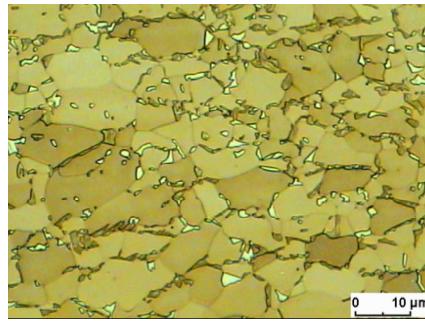


Fig. 2. EBSD map of retained austenite in TRIP steel microstructure

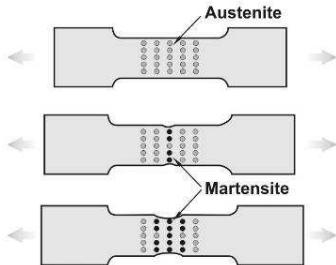


Fig. 3. Illustration of the TRIP effect during tensile test

Manganese is an austenite stabilizer and it lowers the cementite start temperature. It also lowers the carbon activity in austenite and ferrite and increases the carbon solubility in ferrite. Manganese is also soluble in cementite. On the contrary, silicon significantly increases the carbon activity in both ferrite and austenite and decreases its solubility in ferrite. As a result, Si inhibits the formation of cementite during the austempering stage. It has extremely low solubility in cementite, and thus, it affects the nucleation of this carbide. It is believed that the accumulation of Si around a cementite nucleus could considerably increase the C activity which prevents its diffusion to nucleus. The minimum level of Si needed to effectively suppress the cementite formation is believed to be around 0.8 wt.%. However, it is important to note that Si slows down the kinetics of bainitic transformation. As a result, cold rolled TRIP steels should be further annealed on a long line with long over-ageing section allowing the

long austempering stage to be conducted. The exploitation of Si to prevent against the carbides formation is currently being limited due to the requirement for continuous galvanizing of AHSS sheet steel. The high silicon content in TRIP steels results in film-forming surface oxides which prevent the formation of the inhibition layer during the hot dip galvanizing which prevents the wetting of the sheet surface by the liquid Zn.

The reasonable solution to handle the model production lines requirement is to limit the silicon content and use increased aluminium content in TRIP steels, preferably, apply the partial replacement of 1 wt.% of Si by 1 wt.% of Al. The CMnAl TRIP steels have received much attention over the last years, because, the high Al content in these steels results in a high C content in retained austenite. Quite similar to silicon, aluminium is insoluble in cementite and, generally, retards its formation, but more importantly, it accelerates the bainite transformation kinetics. This effect of aluminium is very important for the TRIP steel industrialization since it allows the adoption of currently available continuous lines dedicated to IF steels, that does not have long overageing sections. The disadvantages of using of Al are that this element lowers solid solution hardening effect as compared to silicon, and increases the M_s temperature.

The recent developments in the TRIP steel production concept includes only partial replacement of Si by a limited amount of Al and use P in the amount of 0.05–0.10 wt.%. Lower Si content makes the steel galvanizable. The use of P is intended to limit the Al content, since P also suppresses the cementite formation and is a very effective solid solution hardening element. It was shown for TRIP steels low in Si that P increases the amount of retained austenite. P also significantly increases the C activity in ferrite.

2.3. Complex phase steels with bainitic matrix

Complex phase (CP) steels belong to a group of steels with very high ultimate tensile strength of 800 MPa or even greater. The chemical composition of CP steels, and also their microstructure, is very similar to that of TRIP steels, but, additionally it contains some quantities of Nb, Ti and or V to cause the precipitation strengthening effect. Typically, CP steels have no retained austenite in the microstructure, but contain more hard phases like martensite and bainite. The microstructure of CP steels is composed of a very fine ferrite with the high volume fraction of hard phase, For cold shaped structures, a triple phase steel containing ferrite, bainite and martensite can be designed which are obviously more difficult to produce. Their mechanical properties are characterised by continuous yielding and high uniform elongation. CP steels with the bainitic matrix have superior formability because the difference between the hardness of bainite and martensite is relatively small. In bainitic CP steels polygonal ferrite is replaced by bainitic ferrite. The bainitic ferrite is strengthened by a high density of dislocations, i.e., $> 10^{12}/\text{cm}^2$, and by a fine dispersion of MA second phase and carbonitrides or carbides.

The bainitic CP microstructure exhibits better strain hardening and strain capacity than that for fully bainitic microstructure. This is illustrated in the schematic stress-strain diagram shown in Figure 4. It involves a strength-graded microstructure where the martensite and bainitic ferrite phases are separated by a third phase of intermediate strength.

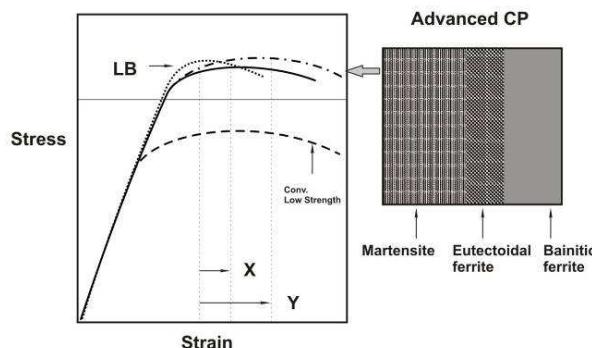


Fig. 4. Stress-strain schematic for advanced CP steels compared with lower bainite (LB) steel and conventional, low strength DP steels

2.4. Martensitic steels

Martensitic steels provide the highest ultimate strength in final products, up to 1500 MPa. Their concept is based upon a well established rules with respect to chemical composition and processing technology design. Microstructure of martensitic steels is mainly composed of lath martensite, which is developed by the transformation of austenite during quenching after hot rolling or annealing. They are often subjected to post-quench tempering with the aim of improving ductility and provide good formability even at very high ultimate strength. The concept of chemical composition of martensitic steels is based upon the proper carbon content adoption, since this element increases hardenability and strength. Manganese, chromium, silicon, molybdenum, boron, nickel and vanadium are also used in various combinations to increase hardenability.

3. Manufacturing technologies of AHSS steel strips

3.1. Dual phase steels

Modern AHSS are produced in complex processes involving thermomechanical processing followed controlled cooling. Alternatively continuous annealing following cold rolling is being used. Both types of processes will be characterised for DP and TRIP steels.

The cooling path in the rolling technology of DP steel strips is schematically shown in Figure 5. The last deformation in the DP strip rolling is conducted at approximately $Ar_3 + 40$ °C.

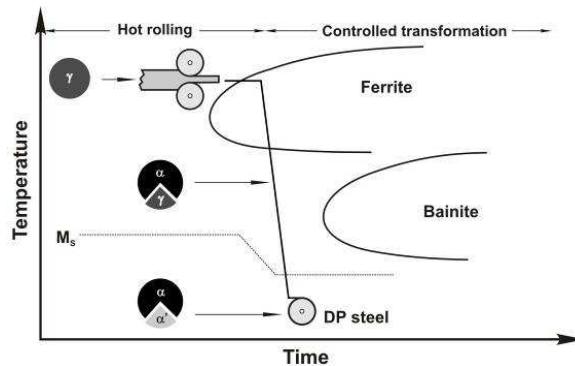


Fig. 5. Cooling schedule in the production of the DP strips

The cooling process after rolling starts with slow cooling stage on the run out table after rolling in which the desirable amount of ferrite is obtained as a result of the austenite transformation. Alternatively, the hot band can be subject to accelerated cooling to the temperature of the lowest austenite stability, followed by a slow cooling to allow the austenite decomposition into ferrite. The ferrite transformation allows the carbon content enrichment in the remaining austenite, which increases its hardenability and reduces M_s temperature. After that, the strip is accelerated cooled to the cooling temperature which must be below M_s . A critical cooling rate to obtain fully martensitic structure of the second phase is given by the following Equation [9]:

$$\log(C_R) > 5.36 - 2.36 Mn - 1.06 Si - 2.71 Cr - 4.72 P, \quad (7)$$

or [10]:

$$\begin{aligned} \log(C_R) &= 5.36 - 2.36 Mn_{eq} \\ Mn_{eq} &= Mn + 0.45 Si + 1.15 Cr + 2 P. \end{aligned} \quad (8)$$

Dual phase microstructure can also be developed in a cold rolled sheet material by the application of continuous annealing followed by hot dip galvanising. A schematic diagram of temperature changes during the annealing process is given in Figure 6.

During the annealing process, a sheet is heated to a temperature slightly above Ac_1 . During this stage, around 15 % of microstructure composed of ferrite and pearlite transforms to austenite. After this, the sheet is quenched causing the austenite transforma-

tion to martensite. As a result, the final strip's microstructure contains predominantly ferrite and martensite in proper proportions.

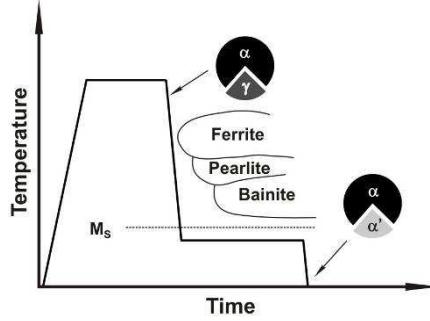


Fig. 6. Temperature changes during continuous annealing of DP steel sheets [11]

3.2. TRIP steels

The optimal combination of strength and ductility of TRIP steels is achieved by decreasing the mean free path in ferrite and by grain refinement and obtaining an uniform distribution of fine second phase particles. The cooling stage in the strip rolling process of TRIP steels, shown in Figure 7, is more complicated than that for DP strips. First, slow cooling is applied in the ferrite transformation temperature range. The transformation causes carbon content to increase in austenite to around 0.4%. After producing 50–60% of ferrite in the microstructure, accelerated cooling with cooling rate greater than 20 °C/s to the cooling temperature which lies in the bainitic transformation temperature range, is realised. During coil cooling, bainitic transformation proceeds, further increasing of carbon content in remaining austenite to around 1.2%. Part of this austenite, in the amount of 10–15%, remains untransformed accounting for the TRIP effect.

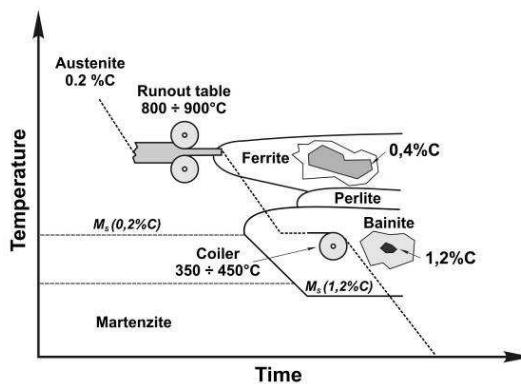


Fig. 7. Cooling schedule in the production of the TRIP strips [11]

TRIP strips can also be cold rolled and annealed. In such a case, a two-stage heat treatment with intercritical annealing in the temperature range 780–880 °C followed by cooling and isothermal annealing in the range 350–450 °C and cooling to ambient temperature is applied (Figure 8).

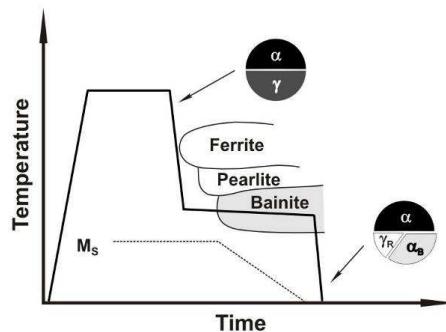


Fig. 8. Temperature changes during continuous annealing of TRIP steel sheets [11]

During the intercritical annealing, the microstructure composed of almost equal portions of ferrite and austenite is formed. However, as opposed to the case of hot rolled strips, during this process, a significant amount of ferrite exists in the initial microstructure prior to annealing. The holding time during the bainitic transformation equals to several minutes. After bainitic transformation, the microstructure contains around 50–60% of ferrite, 25–40% of bainite and 5–15% of retained austenite.

3.3. CP steels

The concept of CP steels is essentially similar to those of TRIP steels, however less stringent cooling practice can be imposed on the hot band during the last stage of processing. This is due to the fact that no presence of retained austenite is required in their microstructure.

3.4. MART steels

Martensitic steels are produced by applying rapid quenching from the austenitic phase to produce the laths martensite microstructure. Tempering after quenching may also be used to improve the strength/ductility relations.

4. Applications aspects of AHSS

The combination of high strength and ductility that provide modern AHSS can allow thinner components to be used in the cars construction and also to improve the safeness due to their high energy-absorption capabilities. The better formability of

AHSS, compared with conventional high strength steels of comparable strength give the automobile designer a high degree of flexibility to optimize the component geometry. Other component performance criteria comprise stiffness, durability, crash energy management. Below, a short characterization of AHSS with respect to these criteria is given.

Stiffness. The stiffness of a component is affected by material module of elasticity (E), as well as, its geometry. The module of elasticity is constant for steel, which means that changing steel grade does not affect the stiffness. Therefore, to improve stiffness, the component geometry must be changed. AHSS offer greater design flexibility to optimize the stiffness due to their enhanced formability. This can be done without increasing mass or decreasing strength. Reduction in gauge can be counterbalanced by changes in geometry or by using continuous joining techniques such as laser welding or adhesive bonding.

Strength. Strength of a component depends on its geometry and yield and/or tensile strength. AHSS provide an advantage in the design flexibility over conventional high strength steels due to their higher formability and work hardening characteristics. These grades also have good bake hardening ability. Therefore, it is important to account for this strength increase during the design process of car components in order to avoid the over design that may occurs when the design process is based upon as rolled mechanical properties specification. Both these features enable achieving high strength of as-manufactured components.

Fatigue. Fatigue properties of structural components depend on geometry, thickness, applied loads and material endurance limit. The endurance limit of a material increases with tensile strength. Thus, high strength combined with superior work hardening and bake hardening, resulting in a significant increase in the as-manufactured strength of AHSS components, also results in a better fatigue resistance.

Crashworthiness. Crashworthiness is an important characteristics that is currently becoming increasingly important. Recent trends require for a material to absorb more energy in crash scenario. The potential absorption energy can be assessed based upon the area under the stress-strain curves. Better performance in crash of AHSS compared to classical high strength steels is associated with higher work hardening rate and high flow stress. This feature account for a more uniform strain distribution in components in the crash event. Both, work hardening and bake hardening significantly improve the energy absorption characteristics due to the flow stress increase.

Formability. AHSS have many advantageous characteristics connected to formability compared to those of HSLA steels with comparable yield strength. AHSS in general have a higher initial work hardening rate, their higher ultimate tensile strength and, especially DP steels, have lower ratio of yield strength to tensile strength. All these advantages combined with an excellent elongation show, that AHSS combine high strength with good formability. High work hardening exponent accounts for the ability of a sheet metal to stretch and the ability of steel to distribute the strain more

uniformly in the presence of a stress gradient. An important advantage of DP steels is very high n -value at low strains which restricts the onset of strain localization and development of sharp gradients.

In the past, the selection process of materials for vehicle-body components was based solely upon their mechanical properties measured in the static tests. Members of ULSAB-AVC consortium adopted significantly different approach to the design of the body structure. They decided to use the established dependence of steel strength on strain rate. This is due to the fact that strain rates under crash conditions are much higher than during the static tests. Higher strain rates induce higher strength of the deformed material which, in turn, results in higher energy absorption. It was decided that the engineering experience of the vehicle designers would be supplemented with FEM simulations to assess forming behaviour. The particular attention in the course of the simulation is paid to:

- Assessment of the possibility of forming a car body component in the assumed tooling and making eventual design changes.
- Selection of the proper steel grades for components commonly considered of difficult or even impossible to form.
- Identification of alternatives to expensive materials or processes.

Table 2. Master materials list

Steel grade	YS (MPa)	UTS (MPa)	Total EL (%)	n-value (5–15%)	r-bar	Application code
Mild 140/270	140	270	38–44	0.23	1.8	A, C, F
BH 210/340	210	340	34–39	0.18	1.8	B
BH 260/370	260	370	29–34	0.13	1.6	B
IF 260/410	260	410	34–38	0.20	1.7	C
DP 280/600	280	600	30–34	0.21	1.0	B
IF 300/420	300	420	29–36	0.20	1.6	B
DP 300/500	300	500	30–34	0.16	1.0	B
HSLA 350/450	350	450	23–27	0.22	1.0	A, B, S
DP 350/600	350	600	24–30	0.14	1.1	A, B, C, W, S
DP 400/700	400	700	19–25	0.14	1.0	A, B
TRIP 450/800	450	800	26–32	0.24	0.9	A, B
HSLA 490/600	490	600	21–26	0.13	1.0	W
DP 500/800	500	800	14–20	0.14	1.0	A, B, C, W
SF 570/640	570	640	20–24	0.08	1.0	S
CP 700/800	700	800	10–15	0.13	1.0	B
DP 700/1000	700	1000	12–17	0.09	0.9	B
Mart 950/1200	950	1200	5–7	0.07	0.9	A, B
MnB	1200	1600	4–5	n/a	n/a	S
Mart 1250/1520	1250	1520	4–6	0.07	0.9	A

Application Code: A – Ancillary parts, B – Body structure, C – Closures, F – Fuel Tank, S – suspension/Chasis, W – Wheels

Steel producers, who participated in the ULSAB-AVC, have proposed various steel grades that should be used for manufacturing the vehicle body structure components. In the final concept design, specific steel grades were selected in a manner that best combine their unique mechanical properties, accounting for the structural demands of specific vehicle body component. The selection has been done for two body structure classes, for the C-class size vehicle and the PNGV-type vehicle.

The C-class is often called the VW Golf class while the PNGV (Partnership for a New Generation of Vehicles) class vehicle is defined as 5/6 passenger sedan with the interior volume of 3.115 m³, curb weight of 907 kg and fuel consumption of up to 3 l per 100 km. Table 2 contains the master materials for ULSAB-AVC. This Table gives the static mechanical properties of ULSAB-AVC body structure, closures, ancillary parts, suspension and wheels.

5. Summary

A substantial progress has been achieved during last years in the development of AHSS for applications in the auto body construction. The new strategy for introducing the technological advances into the mass production has been suggested. The most important contribution in this area should be ascribed to ULSAB and ULSAB-AVC research projects.

The approach adopted in these projects takes into account the diversification of material properties with respect to formability and energy absorption. To master the AHSS auto body components production technology, it is necessary to develop modern production lines that assure an accurate processing parameters control in the subsequent production stages. Temperature is the most important process parameter that must be precisely controlled after hot rolling and during continuous annealing of AHSS sheets.

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Wielofazowe stale wysokowytrzymałe dla przemysłu samochodowego

Celem artykułu jest przedstawienie najważniejszych koncepcji projektowania składu chemicznego oraz technologii obróbki cieplno-plastycznej i właściwości mechanicznych oraz użytkowych nowoczesnych stali wielofazowych dla przemysłu motoryzacyjnego. Stale te, do których zalicza się następujące gatunki; DP, TRIP, CP i MART, uważane są za najważniejsze materiały do zastosowań w konstrukcji samochodów w przyszłości. Ich cechą charakterystyczną są bardzo skomplikowane zależności między strukturą a właściwościami mechanicznymi.