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Reverse metallurgical engineering towards sustainable manufacturing of vehicles using Nb and Mo alloyed high performance steels

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Abstract Steel is the dominant construction material for most industrial goods such as equipments, structures, buildings or vehicles. Although there have been great advances in steel technology over the last 4 decades, the industry currently faces serious sustainability challenges with regard to energy conservation, reduction of CO_2 emission and a generally more efficient use of resources. The principal connotation in this respect is increasing steel strength allowing to reduce component weight. However, it is also necessary considering in how far the modified steel properties interfere with typical manufacturing techniques established in the processing chain. A feasible method in this sense is a reverse approach starting from detailed knowledge of the manufacturing process and translating it back into the most suitable metallurgical and microstructural design of steel. Modifying steels towards better manufacturing performance typically involves innovative alloying and metallurgical processing concepts. Niobium and molybdenum are two of the most powerful alloying elements in helping to adapt microstructures and properties with regard to downstream manufacturing processes. This paper will highlight several examples how a reverse metallurgical engineering approach can be successfully applied to optimize the efficiency of subsequent manufacturing processes with a focus on the automotive industry.

Keywords Automotive flat steel · Engineering steel · Stamping · Bending · Profiling · Press hardening · Welding · Cutting · Case carburizing

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1 Introduction

Increasing attention is being paid to the economic advantages that high-strength low-alloy steels have to offer. These advantages include lower structural weight (see Fig. 1), increased resistance to brittle failure, better economies during construction and transportation as a result of lower cost in handling lighter sections, fewer man-hours of welding and lower electrode consumption as a result of reduced gages (see Fig. 2). These perceived advantages are of primary interest to the transportation and materials handling industry where the ratio of payload to dead weight load is of paramount importance. Fringe benefits resulting from this are greater speeds, less fuel consumption per load and smaller sized propulsion units. In addition to mere strength, the selection of structural steel is concerned with ease of forming, welding and other fabrication procedures. Service conditions require that steel exhibit good toughness at the temperature of service. Thus the candidate steel should possess adequate impact resistance at the lowest temperatures anticipated in service.

No other industrial sector has pursued weight reduction as consequently as the automotive industry. This is motivated by the reduction of fuel consumption and emissions as well as by the improvement of crash safety. Both contradictory requirements continue to be major technical targets for future vehicles. The total vehicle weight has an important impact on fuel efficiency as indicated in Fig. 3. Indisputably, the fuel consumption and the CO_2 emission decrease with the vehicle weight. Detailed studies revealed that a weight reduction of 100 kg results in fuel savings of 0.15 L to 0.5 L per 100 km driven [1]. The corresponding reduction of CO_2 emissions for passenger cars ranges from 4 g/km to 12 g/km. The intensive use of stronger steels is effectively offering a significant weight reduction potential.

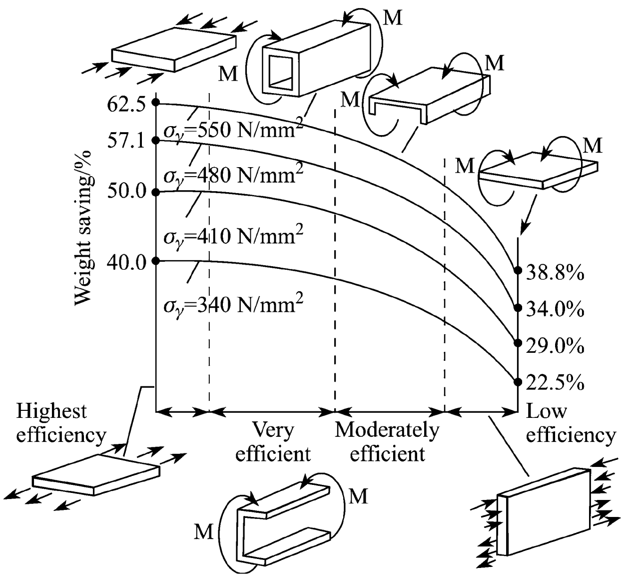


Fig. 1 Weight saving potential by substituting 200 MPa (YS) steel with high strength steels for various loading conditions

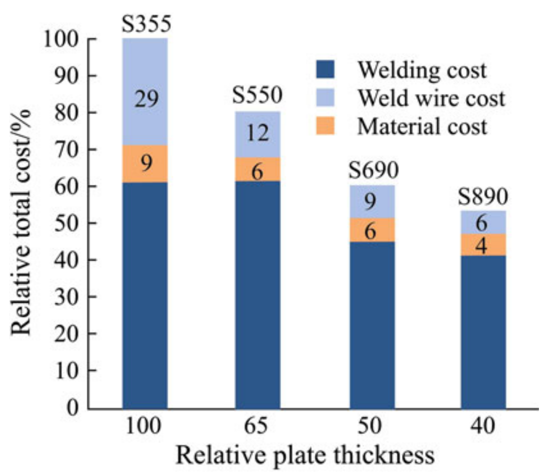


Fig. 2 Material and welding cost of higher strength steels relative to structural steel grade S355

The weight of state-of-the-art car bodies consists of up to 80 % higher strength steel having reduced the weight by 40–100 kg compared to a traditional car body that consisted mainly of mild steel [2–4]. Lighter vehicles not only reduce fuel consumption but also feature faster acceleration, shorter braking distance and an overall better drivability [5].

Very effective in terms of emission reduction is the so-called down sizing approach. In this case a smaller engine size typically with reduced number of cylinders is used at a similar power output than the previously bigger engine. Fuel consumption is lowered by means of reduced internal friction. Simultaneously the weight of the powertrain is lower. Thus, the specific load to powertrain components

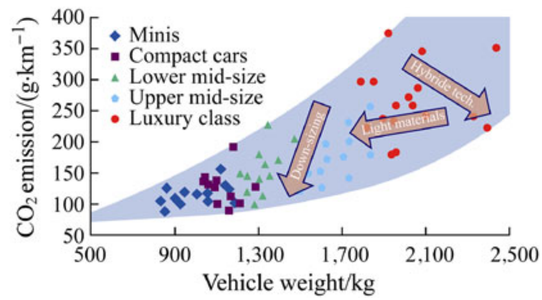


Fig. 3 Influence of vehicle (curb) weight on CO_2 emission of passenger vehicles and concepts for reducing weight and emission

increases under these conditions and hence upgraded steel and more elaborate processing is required.

Automotive production employs a broad variety of established manufacturing technologies. Key technologies are sheet metal forming, massive forming, heat treatment and welding. Steel has to comply with these technologies to allow a smooth and robust production. In this regard particular properties are typically being specified that steel suppliers have to fulfill. However, in practice there are more detailed characteristics making the difference in daily production even if the selected steel generally fulfills standard specifications. A successful metallurgical development in this respect requires detailed knowledge of the manufacturing process. Based on this knowledge steel's microstructure is being optimized by adjusting the alloy design and thermo- mechanical processing in the rollingmill.

2 HSLA steel for chassis and frame parts

Chassis and frame components are typically subjected to high static and dynamic loads. Application of high strength steel effectively allows reducing sheet gage by the principles shown in Fig. 1, thus lowering the component weight. Low carbon equivalent of such steel is mandatory for good weldability. Accordingly, high strength low alloy (HSLA) steels typically limit the maximum carbon content to 0.12 %. HSLA steel for automotive applications is currently applied with yield strength of up to 700 MPa. For applications up to 500 MPa yield strength such steels typically have a ferritic-pearlitic microstructure. Above that level ferritic-bainitic of fully bainitic microstructures are required. Frame and chassis components often have a moderate shape complexity (see Fig. 4). Hence the limited elongation of HSLA steel (see Fig. 5) is usually sufficient to form such components. At similar strength level, DP and TRIP steels offer better elongation than HSLA steel providing an alternative for manufacturing components of higher shape complexity.

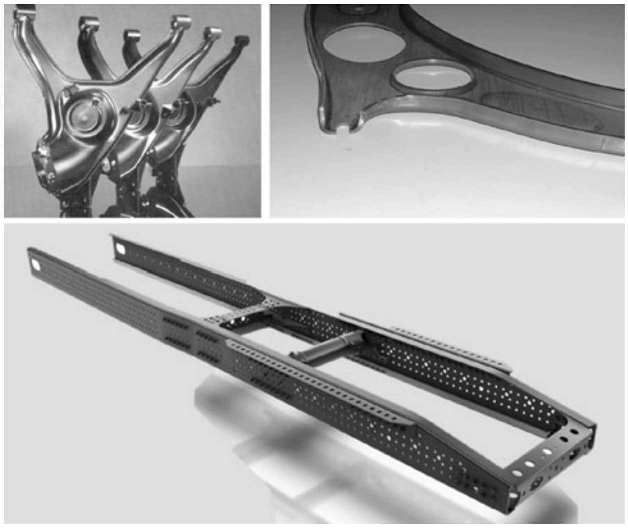


Fig. 4 Chassis and frame components produced from hot-rolled high strength steel comprising low-medium shape complexity and typical forming methods such as bending, stretch flanging, hole cutting and expansion

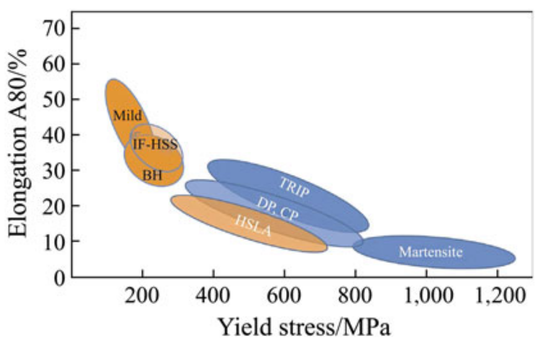


Fig. 5 Relationship between yield stress and total elongation for common automotive steel grades

Manufacturing of frame and chassis components significantly involves particular forming methods such as profiling, bending, stretch flanging as well as hole cutting and expansion. In this respect the so-called hole expansion ratio is a more suitable performance indicator than elongation. The hole expansion ratio (according to EN ISO-TS 16,630 or JIS Z 2,256:2,010) at a given strength level depends strongly on the steel microstructure (see Fig. 6). It is obvious that single-phase microstructures like ferrite, bainite and martensite are performing much better than microstructures containing phases of very different hardness such as ferritic-pearlitic or dual phase steels. Furthermore, the condition of the sheet edge after mechanical cutting is of high importance. It is essential to avoid pre-damage at the edge after mechanical cutting. Edge damage again is promoted by the presence of hard phases such as martensite or pearlite in the microstructure (see Fig. 7). The heavily localized deformation during mechanical cutting in combination with the largely different plasto-mechanical properties between the hard and soft phases leads to delamination at the phase boundary. This initial damage can grow into a macroscopic crack upon stretching the cut edge (see Fig. 7). Laser cutting is a suitable method to avoid damage initiation.

Taking the specific manufacturing details of chassis and frame parts into account, the tasks for reverse metallurgical engineering towards an optimum property profile of steel can be formulated as follows:

(i) single phase microstructure;
(ii) low carbon equivalent;
(iii) reduced scattering of mechanical properties;
(iv) high yield-to-tensile ratio;
(v) avoidance of hard particles in the microstructure.

Principally, these demands are best fulfilled by either a ferritic or bainitic microstructure as suggested by Fig. 6. The specific task is to adjust alloy composition and processing in the rolling mill to obtain the desired mechanical properties. The strength in ferritic microstructure has to be achieved by grain refinement and precipitation hardening since pearlite dispersed in ferrite is detrimental to some of the demands quoted above. Both methods have a potential of increasing the yield strength by up to 300 MPa each. Thereby the initial preference goes to grain refinement since it is the only strengthening mechanism that simultaneously increases toughness. Bainitic steels are combining a refined microstructure with dislocation strengthening whereas precipitation strengthening is less prominent due to specific processing conditions. Particular alloying elements are crucial in the metallurgical concept of such optimized steels. Niobium is the strongest grain refining element known and is applied in the range of 0.03 to 0.1 mass percent [6]. It also has the potential for precipitation hardening forming nano-sized NbC particles in the ferrite matrix [7]. Titanium can be used as an additional microalloying element for precipitation strengthening. Molybdenum is very effective in avoiding the formation of

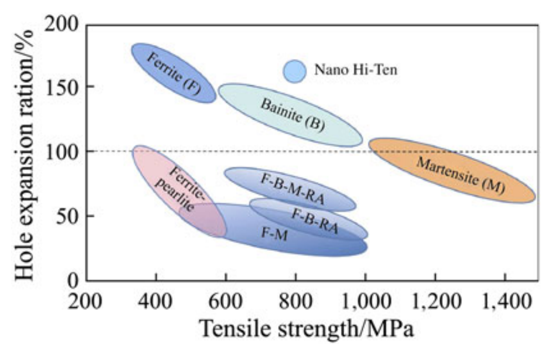


Fig. 6 Relationship between microstructure and hole expansion ratio for common automotive steel grades (RA = retained austenite)

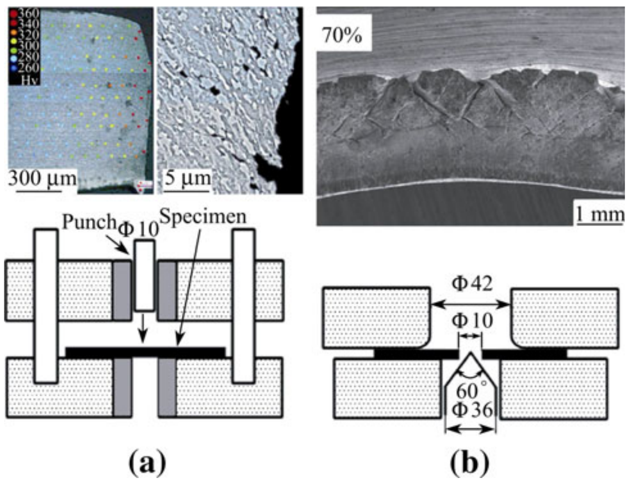


Fig. 7 Principle set-up of hole expansion test **a** damage initiation and hardening by cold working at the hole edge, **b** crack development at the edge after 70 % hole expansion

pearlite and promotes the transformation into bainite [8]. Furthermore, molybdenum limits the coarsening of carbide precipitates by Nb or Ti [9]. The strengthening effect of such precipitates is the best when their size is in the lower nanometer range [10].

The so-called nano Hi-Ten steel is a specific example for reverse metallurgical engineering with the target of designing steel that has very high strength and simultaneously an extremely high hole expansion ratio [9, 10]. Such a high hole expansion ratio necessitates a polygonal ferritic single phase microstructure, which however has intrinsically a relatively low strength. To reach a target minimum tensile strength of 780 MPa, the ferritic microstructure must be severely grain refined and precipitation hardened. For this purpose Nb and Ti are being alloyed. Furthermore, any formation of pearlite must be avoided and precipitate size must be limited. Therefore a low carbon content is chosen and Mo is added. The processing concept relies on the following steps according to Fig. 8:

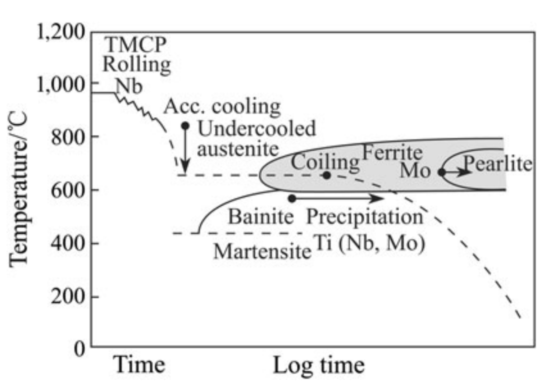


Fig. 8 Processing concept for hot rolled Nano Hi-Ten steel and effects of the alloying elements Nb, Ti, Mo

(i) Austenite pancaking by Nb microalloying in combination with thermo-mechanical rolling (TMCP).
(ii) Fast cooling after the last rolling pass into the range of ferrite formation providing additional grain refinement supported by the delay of austenite-to-ferrite transformation caused by solute Nb, Ti and Mo.
(iii) Quasi-isothermal holding in the ferrite range to obtain complete transformation and precipitation of the microalloying elements Ti and Nb.

Figure 9 indicates the tensile strength level that can be obtained by this process. An alloy concept based only on Ti addition does not provide sufficient strength as TiC precipitates coarsen during longer holding time at high temperature, thus losing their strengthening effect.

Combined alloying of Ti, Mo and Nb significantly enhances the strength by grain refinement and precipitate size control even after extended holding at high temperature. Steel produced with an optimized combination of alloy and process was found to have a hole expansion ratio of more than 150 % and total elongation of around 20 %. However, the processing window to reach such properties is narrow and applicable sheet gage is limited to thinner dimensions.

For the bulk of applications having less demands towards hole expansion ratio it is desirable to design a more robust process and metallurgical concept. Robust design leads to low scattering of mechanical properties, which is beneficial with regard to avoiding spring back in forming operations. Scattering of mechanical properties is majorly related to temperature control on the run-out table after finish rolling. Temperature variations along the coil and also across the strip commonly occur. Hence a robust alloy design is needed that is insensitive to temperature variations on the run-out table. An alloy design employing

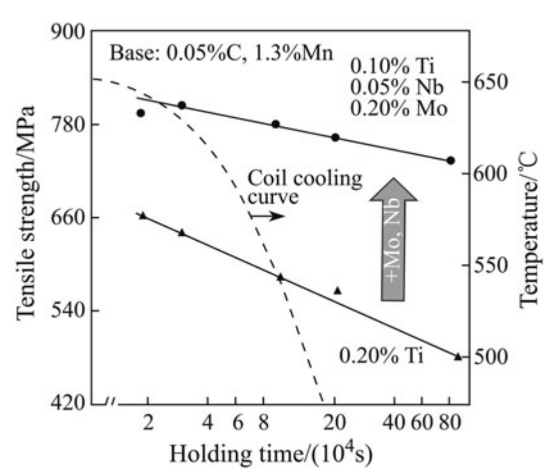


Fig. 9 Evolution of tensile strength as function of isothermal holding time above 600 °C (indication of real coil cooling curve)

competing metallurgical strengthening mechanisms is a suitable solution (see Fig. 10) [11]. Here the total strength is balanced by the degree of transformation and precipitation strengthening, respectively. Both effects depend on temperature and balance each other in magnitude so that the total strength being the some of both effects remains nearly constant below a coiling temperature of approximately 600 °C. For practical implementation of this strengthening strategy an alloy concept based on low carbon and a niobium addition in the range of 0.07 % to 0.1 % appears to be particularly suitable. Aided by a considerable amount of solute Nb this alloy concept efficiently transforms into non-polygonal ferritic or bainitic microstructure providing a high base strength. The solute amount of Nb is available for fine precipitation. Coiling temperature is preferably set to below 630 °C. An example of such an alloy concept is shown in Fig. 11 using a combination of 0.04 %C-1.4 %Mn-0.09 %Nb [12]. Yield and tensile strength are extremely stable over a wide range of coiling temperatures from 610 °C down to 480 °C qualifying for European grade S550MC or Chinese grade 610 L. Statistical evaluation of several hundred coils of this material during an industrial production campaign confirmed very narrow scattering of yield and tensile strength (see Fig. 12). For comparison, a more traditional alloy concept exhibits a much wider scattering of strength data indicating lower robustness against process variations. The new alloy concept also reveals excellent bending properties as well as high quality cutting edges (see Fig. 13). Steel processors perceived a significantly enhanced life of punching tools due to the absence of hard particles such as pearlite. The hole expansion ratio of this steel is the range of 120 % and elongation (A50) is between 20 and 25 %.

Production of European grade S700MC is possible by some modifications of the alloy concept still using the same

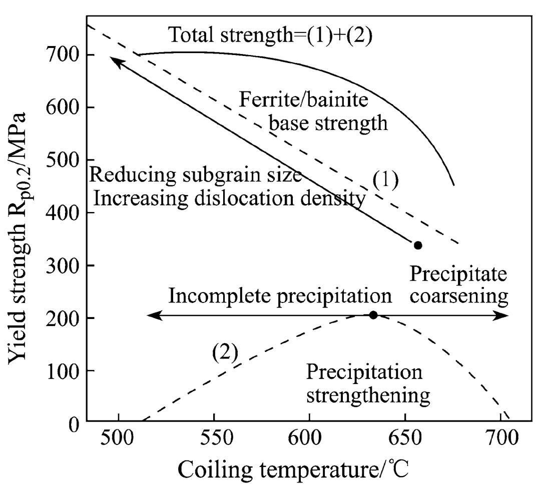


Fig. 10 Concept of strength balancing between transformation and precipitation hardening as function of coiling temperature

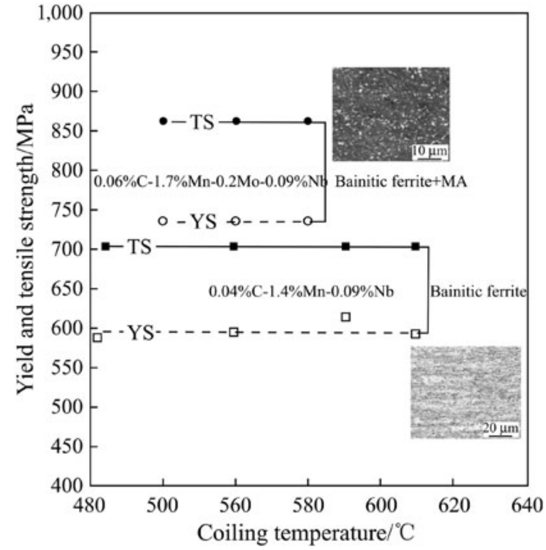


Fig. 11 Strength properties of two alloy concepts as function of coiling temperature and respective matrix microstructures

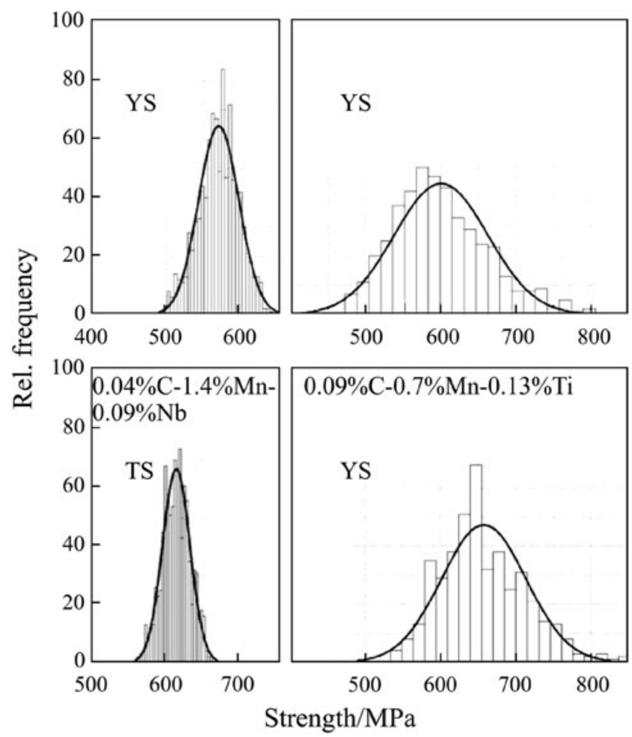


Fig. 12 Statistical evaluation of industrial coil production for target grade S550MC

strength balancing strategy. This was achieved by increasing the carbon and manganese content as well as adding molybdenum. In addition, this concept shows stable properties over coiling temperature varying in the range of 580 °C to 500 °C. The microstructure is basically bainitic containing a smaller fraction of MA phase (see Fig. 11).

Fig. 13 Processing properties of 0.04 %C-1.4%Mn-0.09 %Nb steel with regard to bendability and hole punching

3 Dual phase steel for structural body parts

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

Although DP steel allows forming of complex shapes, practical experience repeatedly revealed unexpected failure even when manufacturing quite simple geometries [13, 14]. Shear fracture at the bend radius was regularly observed in die bending operations (see Fig. 14). Another common problem was sheared edge splitting during flanging operations (see Fig. 15). Roll profiling leads to splitting in the outer bend when the bend radius is too tight causing severe downtime of production lines. All these situations have in common that strain is highly localized. Thus, severe dislocation pile-up is occurring in individual ferrite grains leading to delamination at the ferrite-marten site phase boundary or to cracking of martensite islands as demonstrated in Fig. 16. Such in situ induced micro damage can grow into a propagating crack. The larger the size of an initial damage site the smaller is the critical stress required for crack propagation. A crack typically propagates along ferrite-martensite interface.

Local dislocation pile-up causing micro damage under the condition of localized straining cannot be essentially avoided in DP steel. Nevertheless measures can be taken to reduce the criticality of this phenomenon. Most important in this respect is to refine the microstructure, i.e. reducing the size of the ferrite and martensite grains. Consequently the size of initial damage is reduced raising the critical stress for crack propagation to higher level. Furthermore it is important to homogenize the microstructure, i.e. to avoid the formation of martensite clusters as these provide an easy path for crack propagation. Both microstructural characteristics, a refined grain size as well as homogeneous phase distribution can be achieved by Nb microalloying in combination with appropriate hot rolling conditions. Application of this concept leads to a refined ferritic-pearlitic microstructure of the as-hot rolled strip. The final cold rolled annealed DP microstructure is in a significant way determined by the prior hot strip grain size although some coarsening usually occurs during annealing. A secondary effect of the refined hot strip microstructure is an accelerated nucleation rate of newly formed phases over the intercritical annealing cycle since the new phases preferably nucleate on the grain boundary. In the refined hot strip microstructure the total grain boundary area is significantly enhanced. As such the statistical probability of having martensite clusters is reduced.

The effectiveness of microstructural refinement by Nb microalloying is demonstrated at the example of DP780 in Fig. 17 [15]. Standard and grain refined DP steels are benchmarked in an instrumented 3-point bending test. It is

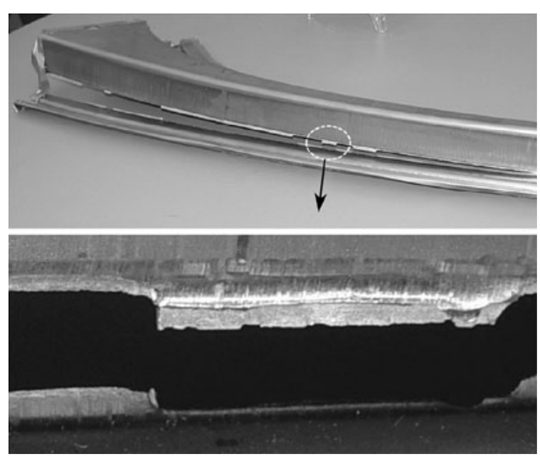


Fig. 14 Component made from DP780 exhibiting shear fracture on a bend radius

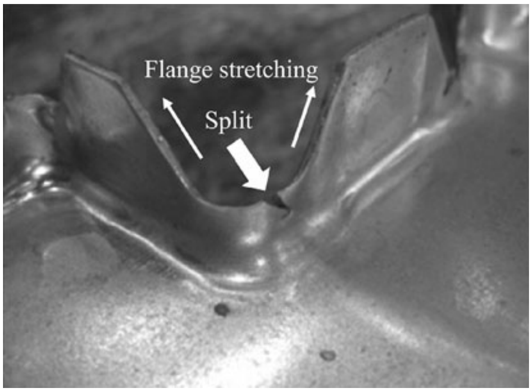


Fig. 15 Crack initiation at a sheared edge of a component made from DP780 after flanging

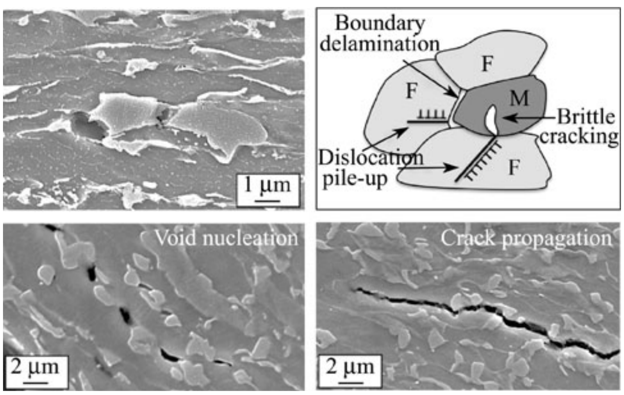


Fig. 16 Features and mechanisms of damage initiation and propagation in DP steel at the hard-soft interface as a consequence of localized deformation

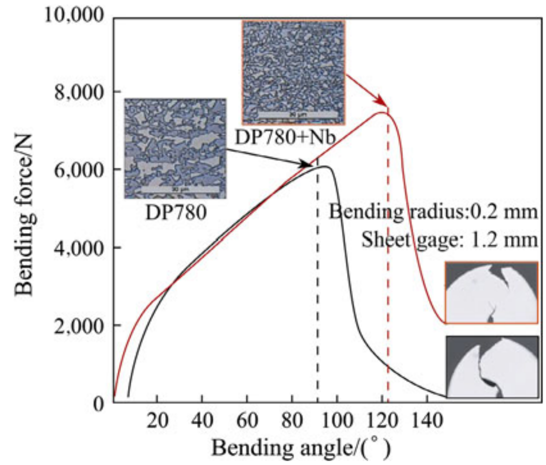


Fig. 17 Bending behavior under 3-point bending conditions of DP780 steel before and after microstructural refinement

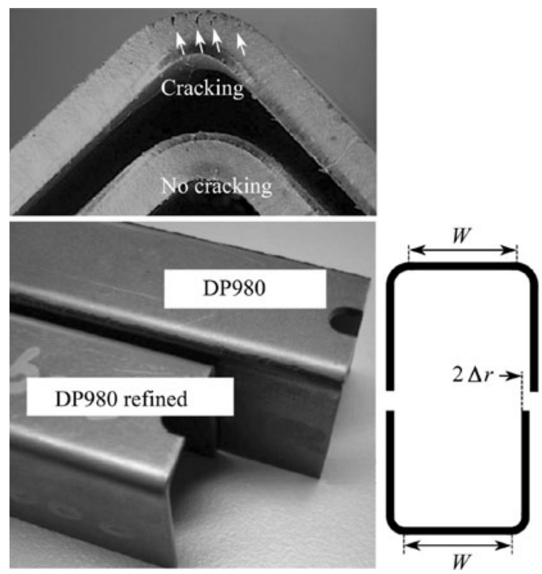


Fig. 18 Effect of microstructural refinement with regard to roll profiling of DP980 and forthcoming reduction of profile dimension

obvious that the grain refined steel supports a higher bending force at an increased bending angle. The standard steel fails at a bending angle of around 90° so that the production of typically U-shaped profiles is critically close to the limit. The grain refined steel on the contrary offers a sufficient margin for the forming process offering a clearly improved process robustness. Similar positive results were obtained with DP980 material where grain refinement allows tighter bending radii in roll profiling (see Fig. 18). In this case a profile with specified inner width W has a more compact outer dimension facilitating packaging issues and reducing total weight. Other experience with grain refined DP steels indicated a much improved hole expansion ratio reducing the risk of sheared edge cracking under flange stretching conditions.

Most of the standard DP steels produced by mills use over-peritectic alloy concepts, i.e., the carbon content is higher than 0.14 weight percent. The carbon content in conventional HSLA steels, on the contrary, is typically below 0.09 weight percent. The increased carbon content facilitates the formation of martensite for the DP microstructure but on the other hand it is detrimental with regard to weldability [16]. Typical low heat input welding processes used in automotive assembly line such as resistance spot welding and laser-welding lead to substantial hardening in the heat affected zone. Such welds can be brittle and sensitive to cold cracking if they are not post-weld heat-treated (see Fig. 19). For that reason there is growing interest on the manufacturing side to reduce the carbon content in DP steel to below 0.1 weight percent.

However, producing DP steel of increased strength level such as DP780 or DP980 with reduced carbon content is challenging, particularly in the hot-dip galvanizing line. The enrichment of austenite with carbon is relatively low in such an alloy concept and hence increases the risk of partial bainite formation during isothermal holding at zinc bath temperature (see Fig. 20). The presence of bainite in the

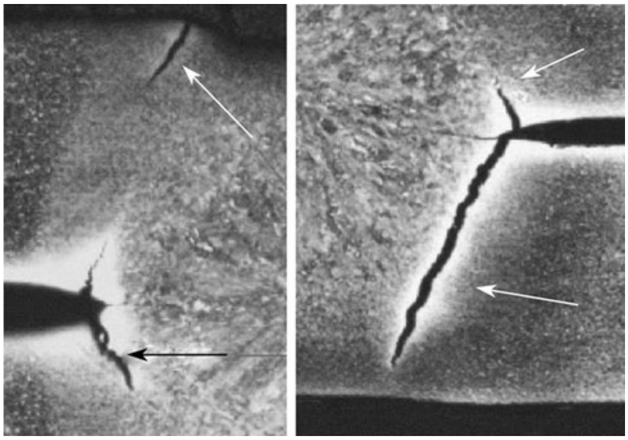


Fig. 19 Appearance of cold cracking in the heat affected zone of spot welded DP780 (0.15 %C)

microstructure reduces the tensile strength and raises the yield strength as compared to ideal ferritic-martensitic DP steel, thus deteriorating the key characteristics of low yield-to-tensile ratio and high *n*-value. Avoiding bainite formation during the isothermal holding phase requires sufficient shifting of the bainite phase field to longer time. This can be practically achieved by increasing the amount of alloying elements such as Mn, Cr, and Mo. The addition of molybdenum to low-carbon DP steel most effectively delays the formation of bainite (see Fig. 20) as its effect is approximately 2.6 and 1.3 times larger than that of Mn and Cr, respectively [17]. Therefore the addition of Mo is most appropriate for producing low-carbon DP steel with good weldability. Furthermore, it is beneficial to use a high cooling rate (CR) after intercritical annealing (ICA).

4 Press hardening steel for highly loaded crash parts

Press hardening also called "hot-stamping" steel has seen a remarkably increasing market in automotive body construction over the recent years. For the majority of applications the steel has a tensile strength of 1,500 MPa (22MnB5) allowing significant weight savings in the car body. Contrary to cold forming steels, press hardening steel is heated to temperatures around 950 °C before press forming. The microstructure is then austenitic providing low yield strength and high elongation during the forming stage. The quick removal of heat when closing the forming die results in a quenching effect, transforming the austenitic microstructure into a martensitic one. Martensite is the strongest microstructure of steel but also the most brittle. By applying press hardening technology a number of conflicting issues inherent to sheet metal forming of other high strength steels could be solved. Hot stamping provides the following advantages:

(i) good formability at low stamping force,
(ii) low wear of the forming die,
(iii) low elastic spring-back resulting in high shape accuracy.

Extensive use of press hardening steel, carmakers consider 45 % of the total structural weight as an upper limit, could bring the weight of a steel body structure to the same low level as achieved with a full aluminum body.

The strength of martensitic press hardening steel is practically entirely defined by the carbon content as shown in Fig. 21. The carbon range from 0.15 % to 0.35 % is covered by commercially available grades providing tensile strength from 1,300 to 1,900 MPa [18]. However, increasing the carbon in such steel has also negative consequences as toughness is being significantly reduced (see Fig. 22). Microstructural refinement on the contrary can considerably raise toughness at a given hardness (see Fig. 22). The microstructural refinement is practically achievable by adding Nb microalloying to the press hardening steel and optimizing the temperature schedule in the hot rolling mill [18]. It was found that the prior austenite grain size in final component made from standard press hardening steel (22MnB5) is often in the range of ASTM 5–6. Niobium microalloyed variants of that steel achieved typically ASTM 7–9 under the same process conditions (see Fig. 23). Austenite grain coarsening in the standard steel occurs especially when the furnace temperature before hot forming is too high or the residence time in the furnace is too long. Niobium microalloying provides temperature stable precipitates obstructing austenite grain growth under such overheating conditions (see Fig. 24). Accordingly the manufacturing process becomes more robust.

The productivity of a hot stamping line is primarily defined by the furnace throughput, the transfer time from furnace to die as well as the die closure time. A conventional heating cycle with a gradual temperature increase to

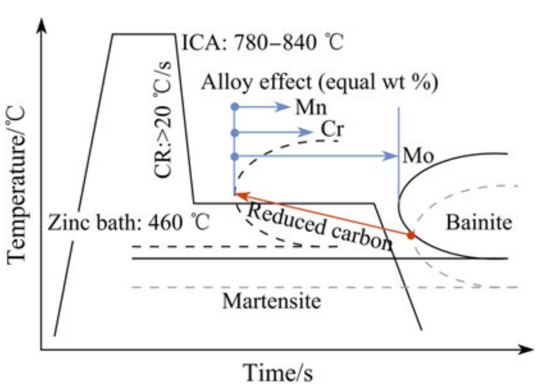


Fig. 20 Alloying strategy for avoiding bainite formation in low-carbon DP steel during a hot-dip galvanizing cycle

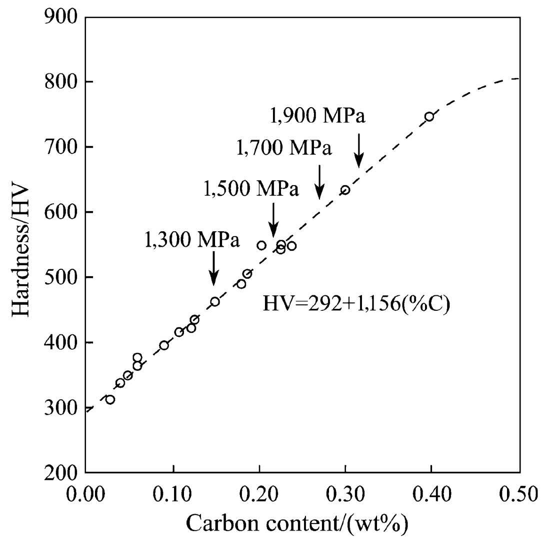


Fig. 21 Relationship between carbon content and hardness (UTS ~ 3xHV) in fully quenched martensite

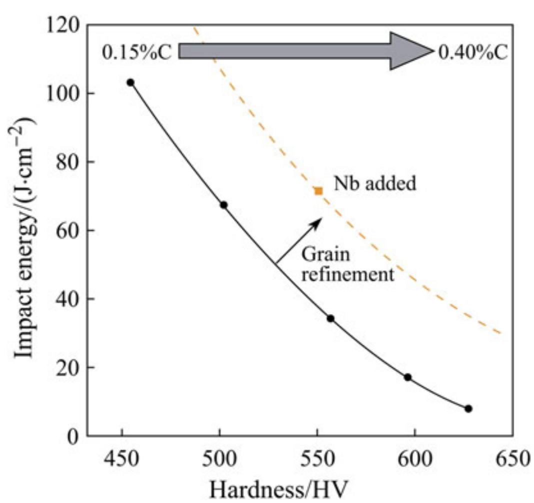


Fig. 22 Relationship between hardness and impact toughness of quenched standard C–Mn–B steel compared to Nb added grain refined variant

the reheating temperature of 930 °C has a dwell time of 5–10 min. Sufficient material throughput with this cycle requires a furnace length of 25 m to 30 m. A rapid heating cycle can be used alternatively to shorten the dwell time to 2–3 min. This heating cycle applies flash heating at temperatures much above 930 °C in the entry section of the furnace followed by a temperature drop to around 900 °C. This process requires a furnace length of only around 10 m. It has been estimated that the energy consumption of the shorter furnace is only 30 % of the conventional furnace.

An important requirement concerning the steel subjected to the rapid heating cycle is that the phase transformation from the as-delivered ferritic-pearlitic microstructure to

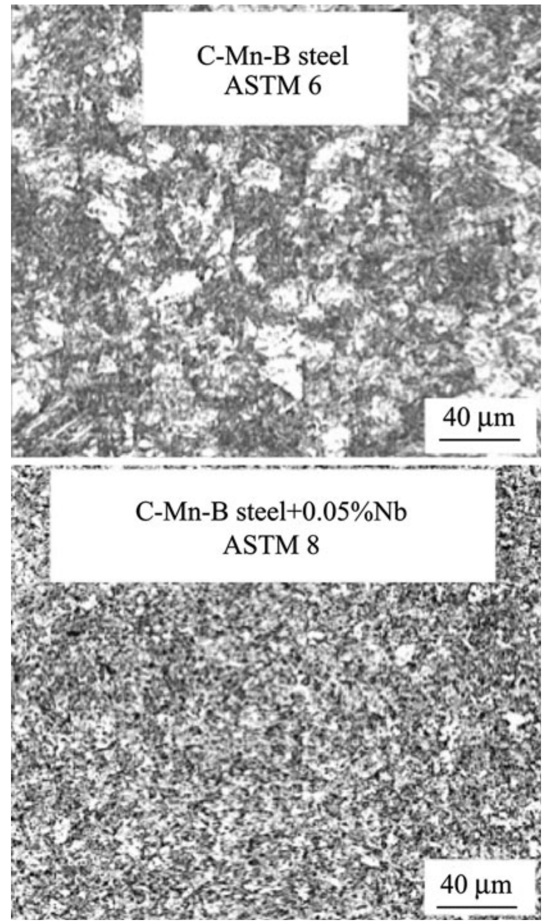


Fig. 23 Prior austenite grain structure of 22MnB5 after hot stamping (heating 950 °C, 6 min)

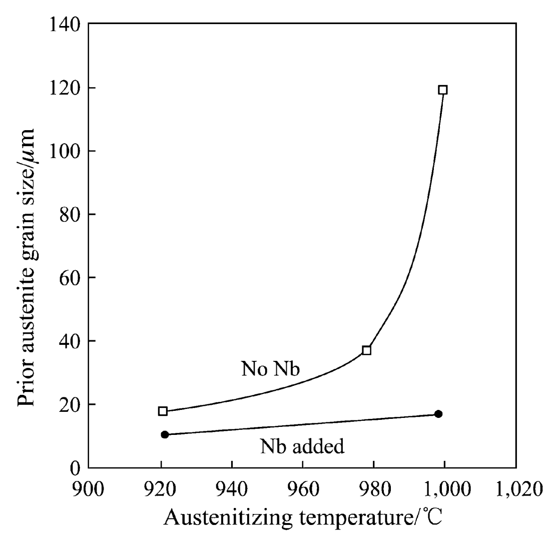


Fig. 24 Effect of Nb microalloying on prior austenite grain size under various austenitizing conditions

austenite occurs quickly and completely. Only then a fully martensitic microstructure providing the targeted strength can be obtained in the final component.

Niobium microalloyed press hardening steel produces a refined ferritic-pearlitic microstructure when hot rolled under suitable conditions. This usually requires a reduced finish rolling temperature. The grain refining effect of Nb is maximized at an addition of around 0.06 %. When reheating the ferritic-pearlitic sheet in the furnace line to austenitizing temperature, austenite primarily nucleates at grain boundaries and ferrite-pearlite triple points. In a grain-refined steel the number of nucleation sites is significantly enhanced due to the larger total grain boundary area and number of triple points. Therefore phase transformation kinetics is accelerated and more compatible with a rapid heating cycle.

5 Carburizing steel for gear and transmission parts

Engine, powertrain and axles account for up to half of the total vehicle cost (see Fig. 25). Many of the respective components are produced from special steel grades involving casting and massive forming (forging) operations. Those components exposed to contact loads such as gears and shafts often require additional heat treatment and surface hardening to achieve a suitable combination of strength, toughness and wear resistance.

Carburizing steels are used in the automotive industry to produce case hardened components for engine and powertrain applications. Table 1 lists different alloy and processing concepts for manufacturing of case hardened components. The selection of a suitable combination depends on the dimension and the operational performance requirements of the component to be manufactured.

In order to satisfy the requirements in terms of quality, performance and cost it is necessary considering the entire process chain from steelmaking through heat treatment to finish machining when planning and selecting a steel grade. The demands on steel materials for case carburized components can be categorized as follows:

(i) chemical composition/hardenability,
(ii) homogeneity/microscopic and macroscopic cleanness,
(iii) mechanical properties (tensile strength, fatigue strength and toughness),
(iv) wear resistance, rolling contact fatigue strength and vibration resistance,
(v) dimensional stability/low distortion after heat treatment.

The workability of steel is also of great economic significance, i.e., steel must be suitable for hot and cold forming as well as for machining. When selecting and developing materials for highly stressed gear components, three fundamental factors most prominent in determining the economic efficiency of a product must be taken into account. Consequently there should always be a balanced relationship between the following factors:

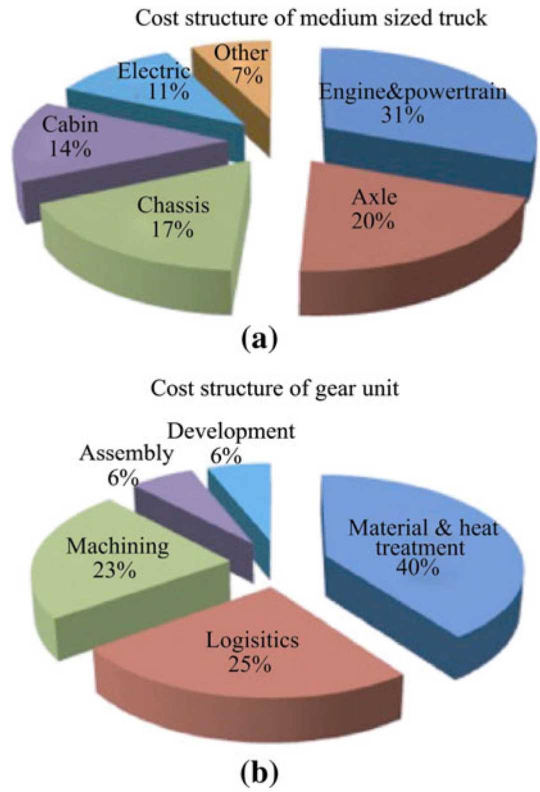
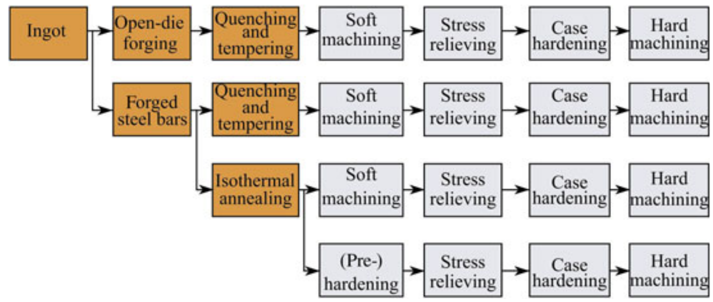


Fig. 25 Analysis of cost structure in a medium-sized truck

(i) component and service properties,
(ii) working properties,
(iii) material costs.

Carburizing steel is subjected to a series of heat treatments during processing. With increasing temperature during heat treatment in the austenite range coarsening of the grain structure can occur. End users typically specify limits with regard to grain size in the final component. Accordingly, 90 % of the grain should have a grain size smaller than ASTM 5 and no grain should be larger than ASTM 3. Coarse grain is detrimental to toughness and fatigue strength (see Fig. 26) [19]. A mixed microstructure consisting of fine and coarse grains leads to shape distortion (see Fig. 27) requiring additional straightening and re-machining efforts [20]. Carburizing treatment is typically performed at temperatures around 950 °C. The case depth is controlled by the carburizing time. Raising the carburizing temperature in the order of 100 °C can significantly reduce the carburizing time for reaching a desired case depth due to higher diffusion speed of carbon at the elevated temperature. Experience has demonstrated that carburizing at 1050 °C can reduce the treatment duration by as much as 60 % (see Fig. 28) [21]. However, the tendency to coarse grain formation is also increased at elevated carburizing temperature. Hence existing alloy concepts should be adapted to:

Table 1 Concepts and routes for manufacturing of carburized powertrain components

Alloy concept	Application	Possible processing routes
Cr	Low hardenability steel for smaller components	
Mn–Cr	Medium-hardenability steel used in passenger vehicles	
Cr–Mo, Mo–Cr	Medium hardenability steel used in passenger and commercial vehicles	
Cr–Ni–Mo	High-hardenability steel used in industrial and commercial vehicles	
Cr–Ni, Ni–Cr	High-hardenability steel used for high toughness requirements	

(i) generally refine the average grain size under standard heat treatment conditions,
(ii) restrict the grain size scattering around the refined mean value,
(iii) limit grain coarsening during carburizing treatments at elevated temperature.

This can be achieved by metallurgical re-engineering of existing case carburizing steels. Adding a combination of niobium and titanium results in the formation of titanium-niobium carbonitride particles that are stable at high temperature and have the capability of pinning the austenite grain boundary. In this case austenite grain coarsening even at increased carburizing temperature is impeded. Furthermore, the grain size is generally refined by the recrystallization retarding effect of niobium during hot deformation of the material prior to heat treatment. Vanadium microalloying does not offer a grain size controlling functionality under such treatment conditions as its respective carbides and nitrides dissolve already at relatively low temperature.

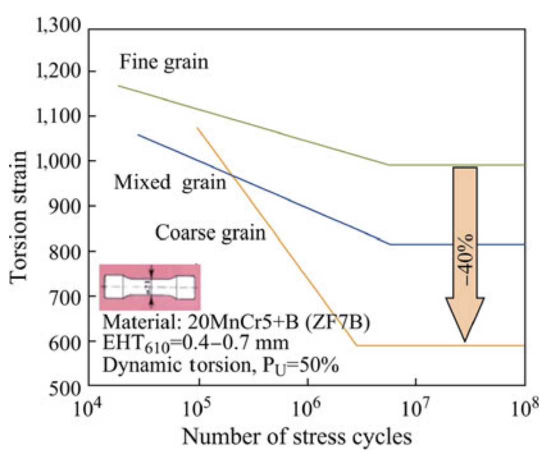


Fig. 26 Influence of grain size on fatigue performance of the carburizing steel 20MnCr5+B

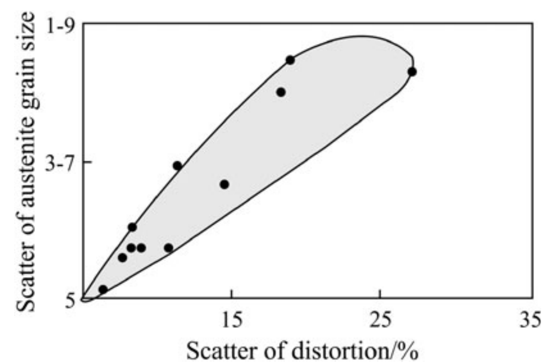


Fig. 27 Correlation between grain size scattering and dimensional distortion in case carburized steel

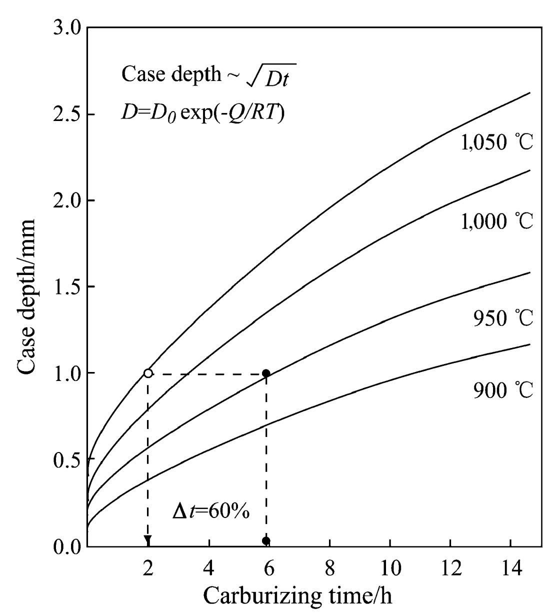


Fig. 28 Influence of carburizing temperature and time on case depth; effect of raising carburizing temperature by 100 °C on reduction of carburizing time for given case depth

Figure 29 demonstrates the effectiveness of the microalloying concept at the example of the high performance gear steel 18CrNiMo7-6 [22]. After carburizing at 1,030 °C for 25 h, the standard grade exhibits coarse grain components that are not accepted by specifications. A modified variant of this grade was microalloyed with Nb and Ti. This grade has a finer average grain size and does not show coarse grain formation after the carburizing treatment. The grain size distribution is also narrower preventing excessive component distortion and hence reducing hard machining efforts.

The benefit of a narrow grain size distribution becomes obvious from Fig. 30 showing the roundness deviation of a main gear shaft of grade 25MoCr4 after carburizing at 980 °C for 195 min. The modified variant that is microalloyed with Nb and Ti shows significantly less roundness deviation at any position along the shaft. The straightening effort is approximately 5 times larger for the standard grade at the position where the largest shape deviations occur (RL3).

The trend towards powertrain downsizing potentially increases the load on gears. Turbocharged engines develop high torque causing higher stress levels at the gear tooth root and flank (see Fig. 31). The stress is the highest at the gear tooth surface and decreases away from the surface. Likewise the hardened gear steel shows the highest strength at the surface. The drop of the material strength into the depth depends on the steel's hardenability response as well as the core strength. Depending on the latter the applied stress can locally exceed the strength of the material resulting in an increased probability of crack initiation. Taking the high performance gear steel 18CrNiMo7-6 as an example, two modified alloying concepts can be practiced to change the hardenability response (see Fig. 32). In one concept the nickel alloy content is

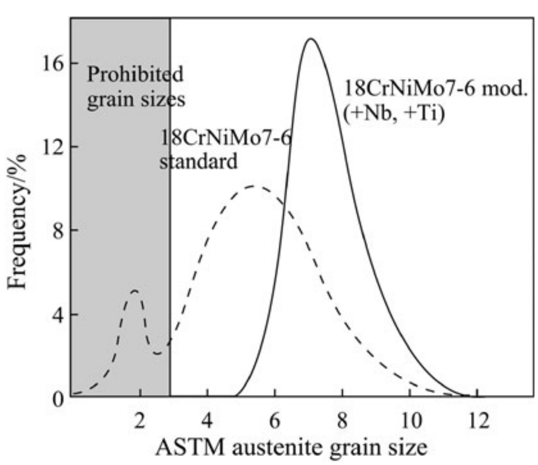


Fig. 29 Grain size distribution of a Nb-Ti modified 18CrNiMo7-6 steel compared to the standard grade after carburizing treatment at 1,030 °C for 25 h

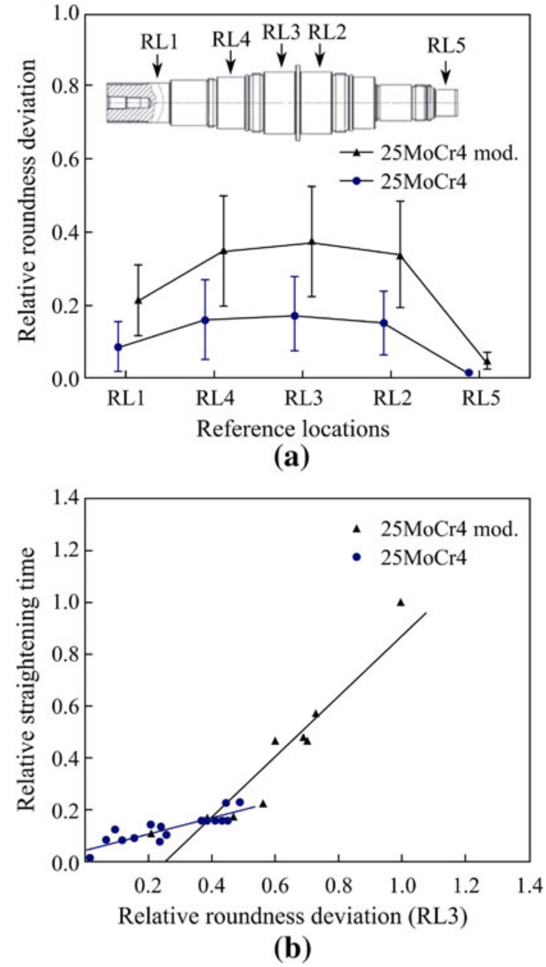


Fig. 30 Roundness deviation and straightening effort at various reference locations (RL) of a truck main gear shaft made from carburizing grade 25MoCr4 after heat treatment at 980 °C for 195 min (the modified grade is microalloyed with 0.03 %Nb and 0.01 %Ti)

increased leading to higher core strength. However in this approach the surface strength tends to be reduced since nickel is a strong austenite stabilizer. Increased amounts of soft retained austenite in a martensitic matrix reduce the strength. An alternative concept is to increase the molybdenum content and to add special carbide forming elements. Since molybdenum is not an austenite stabilizer, it allows achieving maximum martensitic hardness in the surface near area and simultaneously raising the core strength. Dispersed hard particles of carbide forming elements can further increase the shell hardness and thus the wear resistance. Toughness will decrease by this approach but can be at least partly regained by grain refinement using niobium microalloying. An increased Mo content also significantly raises the tempering resistance of the steel. Therefore additional treatments such as plasma nitriding or diamond-like carbon (DLC) coatings become applicable to the already quench hardened steel.

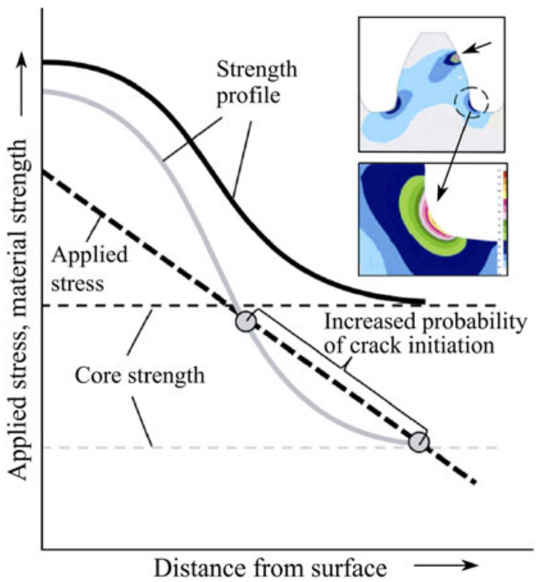


Fig. 31 Stress situation at a gear tooth in comparison with strength profiles with gear steels identifying the risk of crack initiation

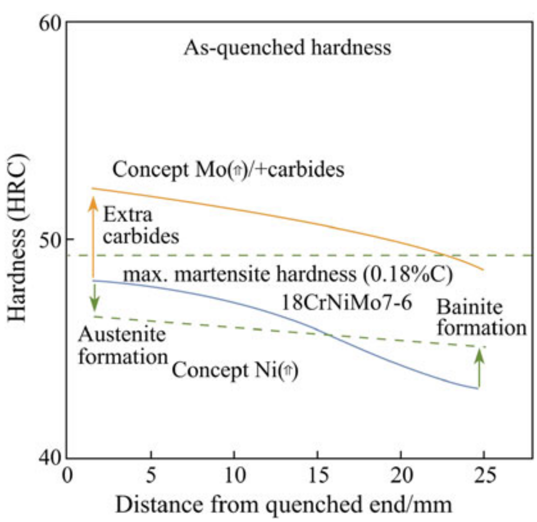


Fig. 32 Quench hardenability curves of possible alloy concept modifications based on the standard high performance gear steel (18CrNiMo7-6)

6 Conclusions

Sustainability in the mobility and transport sector requires weight reduction of vehicles, which can only be achieved by stronger materials such as high strength steel. A major challenge is processing of high strength steel in the manufacturing chain. It was shown that reverse metallurgical engineering could significantly facilitate high strength steel processing by optimizing particular key properties. It is thereby crucial to closely understand the processing demands and translating them into metallurgical and microstructural characteristics. Niobium and molybdenum are two powerful alloying elements in this metallurgical optimization effort. Several examples demonstrated that a combination of these elements with a suitable processing concept could provide steels that have superior performance in the manufacturing chain. The improved manufacturing performance additionally contributes to sustainability. Lower scrap rate during forming and welding means that less material must be produced. Improved machining and more efficient heat treatment require lower consumption of tools and energy. The achieved resource savings are clearly outweighing the moderate additional costs for alloying elements such as niobium and molybdenum.

References

1. Helms H (2005) Fuel saving by light-weighting for European articulated trucks. http://www.ifeu.de
2. Automotive Circle International-Euro Car Body Conference Series (2008–2010). In: Proceedings of global car body benchmark conference series, Vincentz, Hannover
3. Lüdke B, Pfestorf M (2006) International symposium on niobium microalloyed sheet steel for automotive application. In: Hashimoto S, Jansto S, Mohrbacher H, Siciliano F (eds) Proceedings of the international symposium on niobium microalloyed sheet steel for automotive application. TMS, Warrendale
4. Steel Market Development Institute (2002) ULSAB-AVC engineering report. http://www.autosteel.org
5. Hallfeldt T (2006) International symposium on Niobium microalloyed sheet steel for automotive application. In: Hashimoto S, Jansto S, Mohrbacher H, Siciliano F (eds) Proceedings of the international symposium on niobium microalloyed sheet steel for automotive application. TMS, Warrendale
6. Cuddy LJ, Raley JC (1983) Austenite grain coarsening in microalloyed steels. Metal Trans 14(10):1989–1995
7. Gray JM, Yeo RBG (1968) Columbium carbonitride precipitation in low-alloy steels with particular emphasis on precipitate row formation. Trans ASM 61:255–269
8. Wang S, Kao P (1993) The effect of alloying elements on the structure and mechanical properties of ultra low carbon bainitic steels. J Mater Sci 28(19):5175–5196
9. Funakawa Y, Seto K (2007) Stabilization in strength of hot-rolled: sheet steel strengthened by nanometer-sized carbides. Tetsu-to-Hagane 93:49
10. Funakawa Y, Shiozaki T, Tomita K et al (2004) Development of high strength hot-rolled sheet steel consisting of ferrite and nanometer-sized carbides. ISIJ Int 44(11):1945–1951
11. Coldren AP, Cryderman R, Semchyshen M (1969) Strength and impact properties of low carbon structrual steels containing molybdenum. In: Proceedings of steel strengthening mechanism, Zurich, Switzerland, 5–6 May 1969
12. Haensch W, Klinkenberg C (2004) Low carbon niobium-alloyed high strength steel for automotive hot strip. In: The 2nd international conference on thermomechanical rolling, Liège, Belgium, 15–17 June 2004
13. Fekete J (2006) International symposium on niobium microalloyed sheet steel for automotive application. In: Hashimoto S, Jansto S, Mohrbacher H, Siciliano F (eds) Proceedings of the international symposium on niobium microalloyed sheet steel for automotive application. TMS, Warrendale

14. Hebesberger T, Pichler A, Pauli H et al (2008) Dual-phase and complex-phase steels: AHSS material for a wide range of applications. In: Fuchsbauer B, Wiehland HJ (eds) Steels for cars and trucks 2008. Verlag Stahleisen GmbH, Düsseltal
15. Larour P, Pauli H, Kurz T et al (2010) Influence of post uniform tensile and bending properties on the crash behaviour of AHSS and press-hardening steel grades. In: IDDRG 2010 biennial conference, Graz, Australia, 31 May–2 June 2010
16. Takagi K, Yoshida T, Sato A (2006) In: Hashimoto S, Jansto S, Mohrbacher H, Siciliano F (eds) International symposium on Niobium microalloyed sheet steel for automotive applications, TMS 2006, Araxa, Brazil
17. Irie T, Satoh S, Hashiguchi K et al (1981) Kawasaki steel technical report 2
18. Mohrbacher H (2010) Metallurgical optimization of martensitic steel sheet for automotive applications. In: Proceedings of international conference on advanced steels, Guilin, China, 9–11 November 2010
19. Hock S, Kleff J, Kellermann I et al (2005) Proceedings of steels in cars and trucks, Wiesbaden, Germany, 5–9 June 2005
20. Randak A, Eberbach R (1969) Eberbach: Einfluss der Austenitkorngröβe auf einige Eigenschaften des stahles 16MnCr5. Härterei- Technische-Mitteilungen 24(3):201
21. Grabke H, Grassl D, Hoffmann F (1997) Die ProzessregelungbeimGasaufkohlen und Einsatzh?rten. Expert, Renningen
22. Hippenstiel F (2007) Tailored solutions in microalloyed engineering steels for the power transmission industry. Mater Sci Forum 539–543:4131–4136