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Thermomechanically Hot Rolled High and Ultra High Strength Steel Grades – Processing, Properties and Application

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Abstract: In all application fields for hot rolled strip products for direct processing, e.g. construction and engineering but also crane and truck industry, there is a strong customers demand for grades with increased strength levels and well balanced formability. At voestalpine Stahl GmbH these requirements were met with the development of the high strength microalloyed steel grade ALFORM700M® and the ultra high strength steel grade ALFORM900M® with a minimum yield strength (YS) of 700 MPa and 900 MPa, respectively.

In the present paper investigations on the steel grades ALFORM700M® and ALFORM900M® are introduced. To explain the evolution of the obtained complex microstructures consisting of bainitic ferrite, bainite and martensite alloy design and industrial production process is discussed on the base of dilatometric experiments and TEM investigations. The formation of precipitates is studied by using a numerical model, chemical methods and mechanical testing after heat treatment. Mechanical and mechanic-technological properties of the two steel grades are compared. Furthermore, some processing aspects as weldability an bending behaviour are highlighted. Finally, some typical applications for this high and ultra high strength steel grades are presented.

Introduction

There is a substantial interest in the development and use of high-strength steels for e.g. construction and engineering but also crane and truck industry. The main goal for all applications is the reduction of weight of vehicles or any constructional element. Since the 1960ies low carbon steels microalloyed with Nb, Ti and V, produced via thermomechanical rolling, i.e. high-strength low-alloy (HSLA) steels, have become widely used in the aforementioned tasks. In a large amount of books and papers the metallurgical basics for thermomechanical processing as well as the various effects of microalloying (MA) elements as inhibition of grain coarsening, retardation of recrystallization and precipitation strengthening of the final microstructure are described and discussed (see for example [1], [2], [3]).

The nomenclature of HSLA steel grades is not consistent in literature, frequently the adjective *ultra high* is used to denote a step forward in the development of high strength steels. For some decades grades up to a yield strength of about 550 MPa have been produced by thermomechanical treatment using MA-additions, newer developed ones reach yield strengths of 760 MPa [4]. Steel grades with higher yield strength level could only be produced by application of a quench and temper process.

In this contribution two commercially available grades are presented which show minimum values for YS of 700 MPa and 900 MPa, respectively. The former one is a classical HSLA-steel, which is ready to use after hot rolling. The latter one needs a simple thermal treatment after hot rolling to reach its ultra high strength level. The two grades are denominated as ALFORM700M®

and ALFORM900M®. The brand name ALFORM indicates the thermomechanically hot rolled HSLA steels series with well balanced cold formability of voestalpine.

In the following chapters some features of production of these two grades are discussed and mechanical properties as YS, ultimate tensile strength (UTS), total elongation and charpy notch impact-toughness as well as processing aspects as weldability and bending behaviour are outlined.

Alloy design

To reach high strength levels it is necessary to optimize the contributions of each hardening mechanism with respect to the ancillary properties like weldability, toughness and bending behaviour. These widely known mechanisms can be briefly described with the following keywords:

(a) solid solution hardening by substitutional and interstitial elements as Mn, Si and C
(b) precipitation hardening which is in our case realized by the addition of Ti, Nb and V in combination with C and N
(c) transformation hardening by controlling the amount of bainite by the elements C, Mn, B, Ni
(d) grain refinement, which is on the one hand a consequence of (c) and on the other hand a consequence of thermomechanical treatment. From this point of view MA-elements help to get a fine microstructure
(e) dislocation hardening, which is enhanced by the suppression of austenite recrystallization and low transformation temperatures.

Table 1 shows the amount of MA-elements used in ALFORM700M® and ALFORM900M®. The typical content of C and N is about 0,1m% and 0,005m%, respectively, furthermore between 1,2 and 1,5m% Mn and from 0,2 to 0,4 m% Si are added in both grades. In the case of ALFORM900M® Ni and B are used to suppress the transformation into the lower bainite region.

Steel	Ti	Nb	V
ALFORM700M®	<0,18	<0,10	<0,08
ALFORM900M®	<0,25	<0,10	<0,10

Table 1: Amount of added MA-elements in the indicated steel grades in [m%]

Hot rolling and microstructure

Production process of hot rolling starts with soaking in a reheating furnace. In the case of MA steel types the sense of soaking is not only to reduce deformation forces, it is also required to dissolve Ti, Nb and V precipitates which have formed while cooling the slab after the casting process. At soaking temperature most of the carbides and nitrides should be dissolved, the rest of them has to prevent the austenite grains from uninhibited growth. As an example, the effect of temperature and time on the grain diameter can be seen from Figure 1. Specimens of ALFORM900M® were heated to different temperatures and held for times between 10 an 500 s. After quenching austenite grain sizes were determined by metallographic means. The strong influence of temperature on the grain size in Figure 1 can be interpreted as a measure for dissolution of carbonitrides. At technical relevant soaking times estimated slab temperatures between 1200°C in the core and 1220°C on the surface turn out to be convenient.

During roughening two essential effects can be assumed in the material, precipitation of carbonitrides stimulated by the temperature loss and recrystallization during and between the deformation passes. Precipitation will be discussed in a later section. To answer the question whether or not recrystallization takes place in the presented alloys at this stage of process, two evidences were used for ALFORM900M®.

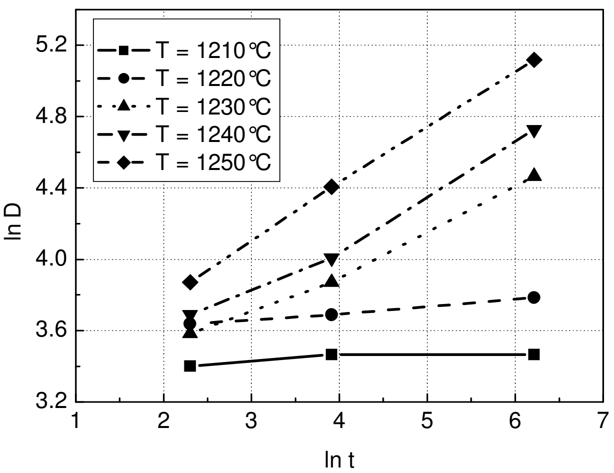

Figure1: Evolution of grain size of ALFORM900M® depending on time in the experiment. Axes are in a logarithmic scale, units: diameter D [µm], time t [s]

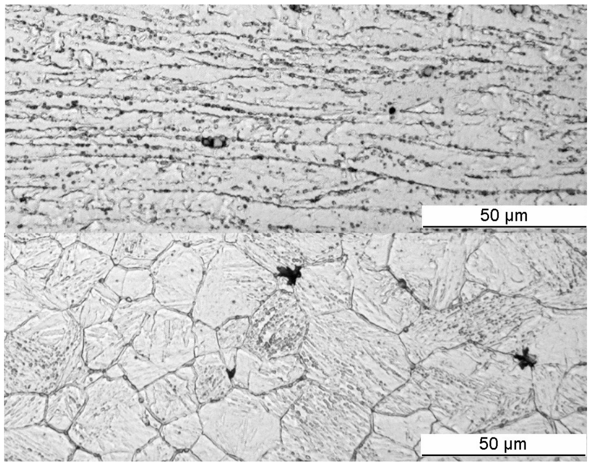

Figure 2: Austenitic microstructure of a crop of ALFORM900M®, elongated structure near the surface, equiaxed grains in the core. A picrinic acid based etchant was used for preparation.

The first was obtained by investigation of the microstructure. For this the crop was water quenched and metallographically analyzed. Figure 2 shows the recrystallized structure in the core of the strip and a thin layer of an elongated structure at the surface. The fact that the specimen was taken from the coldest spot of the bar leads to the assumption that the major part of it is recrystallized. Investigations using deformation experiments (hot compression tests in a deformation-dilatometer) support this assumption. Figure 3 and 4 show the stress-strain behaviour of ALFORM900M® at the highest and lowest possible temperature of roughening in a double-hit experiment with various interpass-times. Both specimens were heated to 1230°C, deformation was carried out at 1230°C (Fig.3) and after cooling in an appropriate manner at 1100°C (Fig.4), respectively. Deformation at the higher temperature with a strain rate of $0{,}1 s^{-1}$ triggers dynamic recrystallization at $\varphi \approx 0{,}2$. This can be supposed from the first hit curve and clearly seen from the second hit curves. Interpass times between 3 and 100 s don't change the behaviour at the second hit. Therefore it can be concluded that static recrystallization takes place within 3 s.

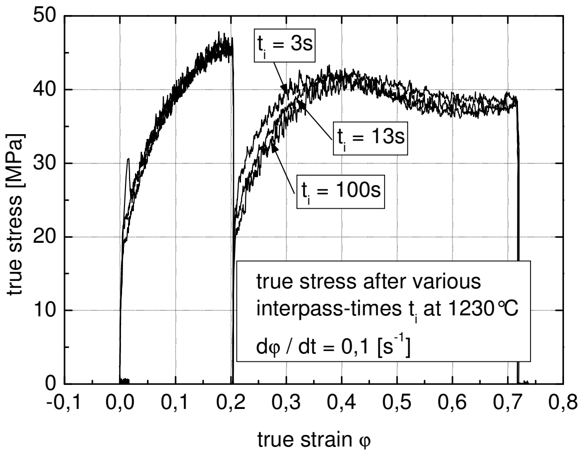

Figure 3: True stress-true strain curves of ALFORM900M® at 1230°C in double-hit experiments with different interpass-times

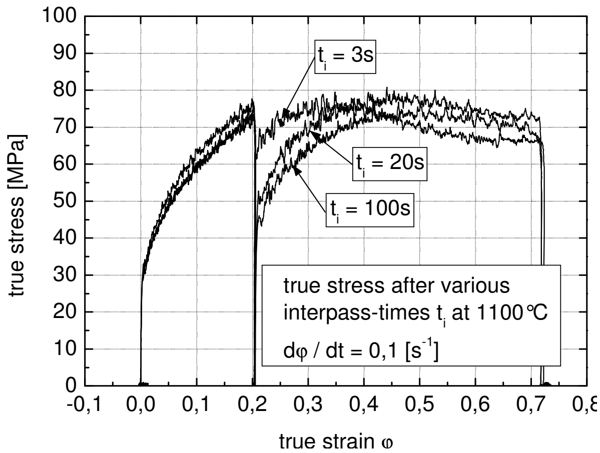

Figure 4: True stress-true strain curves of ALFORM900M® at 1100°C in double-hit experiments with different interpass-times

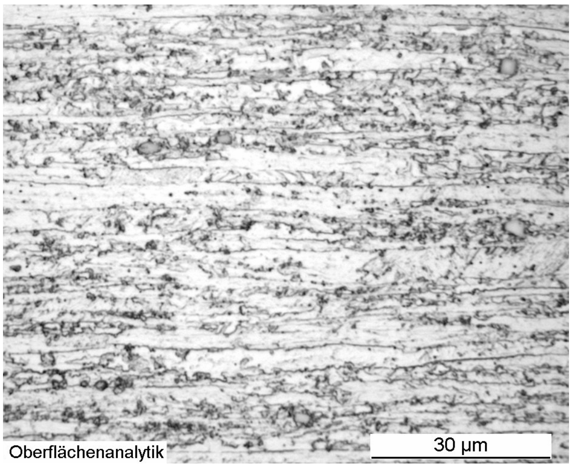

Figure 5a: Microstructure of ALFORM900M® in the light microscope after Nital etching

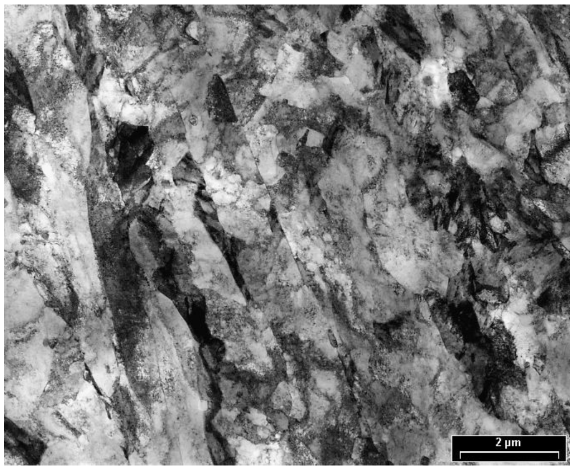

Figure 5b: Microstructure of ALFORM900M® in the transmission electron microscope

At 1100°C the stresses at the second hit are similar after 20s and 100 s but the stress strain curve after an interpass time of 3s shows a significant higher yield point, recrystallization is at least uncomplete. The real roughening process works with much higher strain rates and interpass times between 10s and 100s. It is well known that high strain rates and long times enhance the recrystallized fraction in a material, the transfer bar can be assumed as totally recrystallized.

Investigations of effects in microstructure while finishing are more complicated. Direct sampling from the finishing mill is impossible, deformation experiments do not allow to simulate real strain rates and interpass periods. Nevertheless, a hint for the assumption, that recrystallization is not completed during or between the final deformation passes is given by the elongated structure as can be seen in figure 5a. This micrograph was obtained by preparation the final microstructure of ALFORM900M® with Nital etching. The aspect ratio of the shown grains correlates with the elongation of the hot strip while finishing.

Cooling and phase transformation

As it was already stated, the thermomechanical treatment has a significant influence on the transformed microstructure. Beside of the alloying elements grain size and deformation state are essential variables which interact strongly with the cooling schedule. To describe the behaviour of the presented materials on the run out table dilatometer experiments where chosen. Figure 6 shows the relative dilatation of ALFORM700M® and ALFORM900M® while cooling from the austenite region to room temperature. Both specimens were heated to 900°C, cooled to 850°C with a cooling rate of $10Ks^{-1}$ and afterwards deformed to $\varphi \approx 0{,}5$ at 850°C. After holding at this temperature for 5 s cooling was carried out with $20Ks^{-1}$ to an intermediate temperature of 550°C. In the case of ALFORM700M® transformation starts immediately after the onset of cooling, the major fraction of austenite transforms between 850°C and 700°C. ALFORM900M® shows a quite different behaviour. Cooling to 550°C does not trigger any transformation until the isothermal period. Figure 7 shows dilatation in time space while isothermal holding. Transformation starts after an incubation period of a few seconds and comes to rest after about one minute. From Figure 6 it can be estimated, that the bcc phase amounts to the half of the sample. Subsequent fast cooling in the experiment with about 100 Ks^{-1} leads to super cooling of the fcc-bcc mixture until 400°C and a martensitic reaction at temperatures below. The obtained microstructure from industrial scale production of ALFORM700M® consists of granular bainitic ferrite. In contrast ALFORM900M® shows substructered lath-like bainite with high dislocation density and small martensitic islands [Fig. 5b].

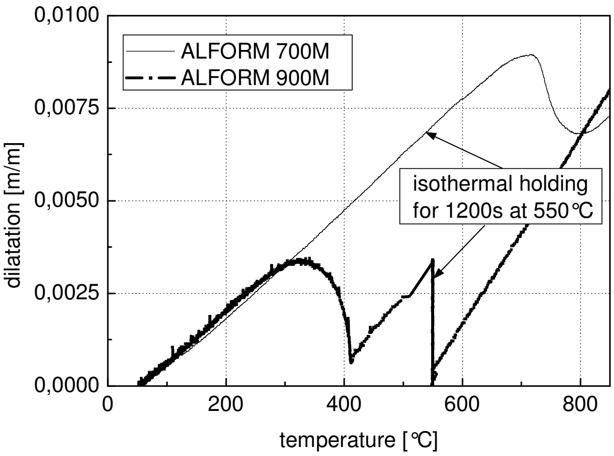

Figure 6: Dilatation of ALFORM700M®and ALFORM900M® during cooling from 850°C to 550° with 20 Ks^{-1} , subsequent cooling to room temperature with $100Ks^{-1}$

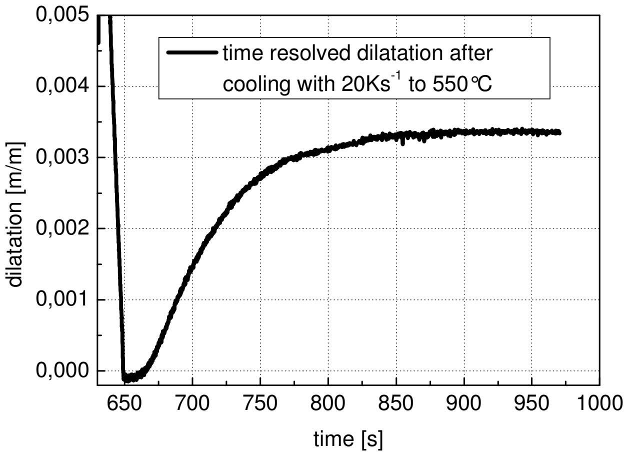

Figure 7: Dilatation of ALFORM900M® during holding period after cooling from 850°C to 550°C with 20 Ks^{-1}

Precipitation of microalloying elements

An experimental approach to nucleation and growth of precipitates is very difficult. To get an idea of the generation of Ti and Nb carbonitrides a theoretical model is used which is based on equilibrium thermodynamics taking also into account capillarity and multicomponent diffusional kinetics. Details of the model are described elsewhere [5].

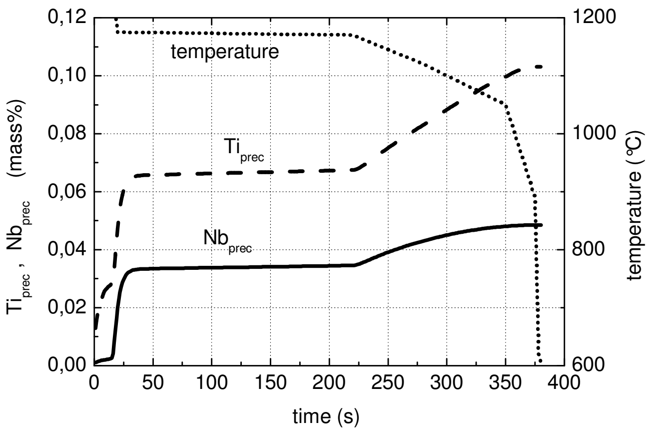

Figure 8: Calculated precipitated amounts of Ti and Nb in ALFORM700M® while following the shown temperature-time path

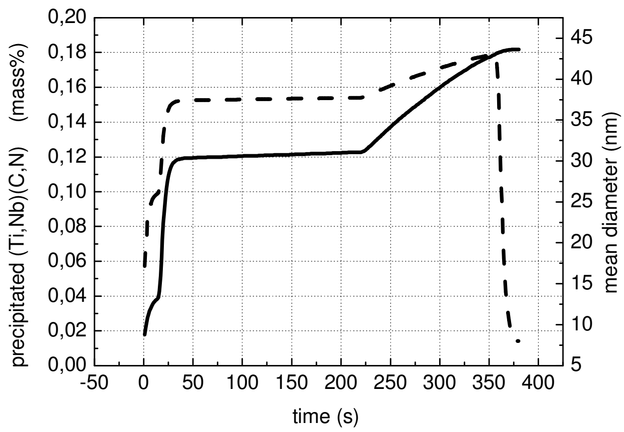

Figure 9: Calculated amounts of (Ti,Nb)(C,N) and mean diameter of the precipitates in ALFORM700M® while following the temperature-time path shown in Figure 8

In contradiction to the real process the calculation starts at a temperature near the solid to liquid transformation. This high start value and the selected subsequent cooling rate have the purpose to deliver a feasible number and density of precipitates for the moment of slab discharge out of the reheating furnace. From this time on the time-temperature curve approximates the real conditions in a section between center and surface of the bar or the strip, respectively. Figure 8 and 9 show the given temperature and as a result the amount of precipitated Ti and Nb (Fig.8) of ALFORM700M®

over time, also the total amount of (Ti,Nb)(C,N) and the mean diameter of the precipitates are shown (Fig.9). From the calculated mass fractions and diameters of the precipitates it can be concluded that during all process steps above 1050°C with slow or medium cooling rates the dominant effect is growing of already existing precipitates, nucleation plays an inferior role. In contrast higher cooling rates at lower temperatures do not lead to an increase in the precipitated mass but to a strongly negative slope of the diameter curve. This can be explained by the pronounced creation of nuclei and fine precipitates. In the case of ALFORM900M® the situation is similar. The differences are only of quantitative nature because of the different added amounts of MA elements. Comparison of the calculated with the measured amounts of precipitated Ti and Nb (see Table 2) indicates the qualitative nature of the investigation. It is not clear whether chemical analysis underestimates or calculation overestimates precipitated masses.

	Ti	Nb	V
ALFORM700M®	45	40	15
ALFORM900M®	65	80	8

Table 2: Precipitated amount of microalloyed elements after hot rolling from chemical analysis in [%] of the added amount

Until now the behaviour of V is not mentioned in this chapter. V is reported to have a higher solubility in iron than Ti and Nb [6]. It can be supposed that the high affinity to N plays no role in the case of the presented steel grades, because N is tied by Ti and Nb at high temperatures. From this, the presented values of the precipitated amount of V in Table 2 can be interpreted as V in VC (or V4C3). Seemingly the slow cooling in the coil plays an important role in V precipitation in ALFORM700M®. The outer winding of the coil shows only a quarter of the average amount of precipitated V. To elucidate the mechanisms and controlling parameters of V precipitation in both grades further investigations have to be done. An indicator for the contribution of V to strength is found in figure 10, where the behaviour of ALFORM900M® in the as hot rolled state and after various thermal treatments of the strip is shown. Specimens were heated to temperatures up to 700°C and held for about 30 minutes,

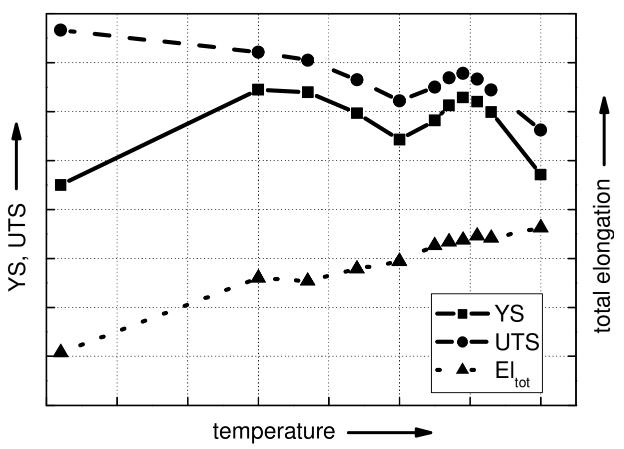

Figure 10: Annealing curves of ALFORM900M® after tempering for 30min, start points denote the as hot rolled state

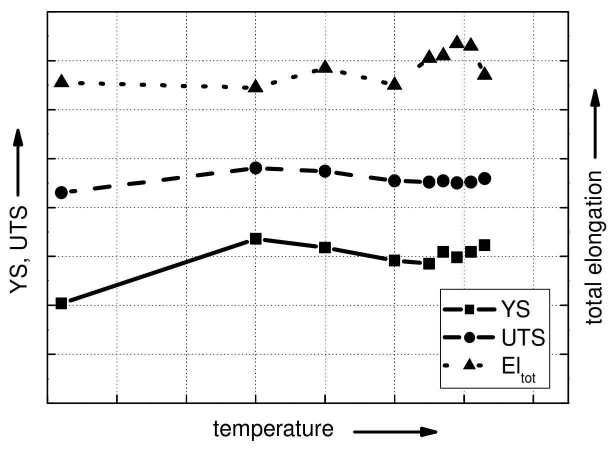

Figure 11: Annealing curves of ALFORM700M® after tempering for 30min, start points denote the as hot rolled state

cooling to room temperature was realized on air. UTS values decrease slightly until a local minimum is found. The succeeding local maximum can be interpreted as precipitation strengthening by VC formation. Chemical analysis at this point shows nearly twice the precipitated amount of V as in the as rolled state. The increase of YS at lower temperatures is assumed to result from ageing. At higher temperatures YS goes parallel to UTS. Detailed analysis of this effect is just in progress.

ALFORM700M® shows a quite different tempering curve (Fig.11). Similarities to ALFORM900M® can only be found for YS in the low temperature region. The lack of a pronounced minimum of UTS and YS at medium temperatures could be caused by the fact that coiling and annealing temperatures are in the same region and phase transformation is in contrast to ALFORM900M® completed at this time (Fig.6).

Mechanical properties and application

The key properties for the application in construction are of course YS, UTS and total elongation as well as low temperature toughness and formability. These properties should be convenient not only on the flat plate before but also after processing. Critical demands on high and ultra high steel grades are bendability, resistance against thermal treatments and weldability. Table 3 shows the guaranteed properties of ALFORM700M® und ALFORM900M® at the delivery state.

Concerning bending, inner radii of 1*s, where s is the strip thickness, are reached for both materials at bending angels of 180° if the cut edges of the bending specimens are grinded. From the practical point of view the cant test with unground but deburred cut edges is more significant. Specimens of a length of 500mm are canted into a die of 90°. For strip thicknesses over 6mm inner radii of 2s at ALFORM700M®and 3s at ALFORM900M®are guaranteed without crack formation, strip thicknesses below 6mm withstand radii of 1,5s and 2,5s, respectively.

Steel	YS [MPa]	UTS [MPa]	A5 [%]	Av(-20°C) [J/cm^2]
ALFORM700M®	≥700	750-930	≥15	≥50
ALFORM900M®	≥900	940-1100	≥11	≥50

Table 3: Guaranteed properties of ALFORM700M® and ALFORM900M®, A5 is the total elongation, Av specifies the charpy notch impact toughness at given temperatures in the as supplied condition

Stress relief heat treatment after processing at the customer can be done without problems up to 600°C. Both grades withstand these temperatures over hours without degradation of the mentioned properties.

For ALFORM700M®and ALFORM900M® fusion welding, flash butt welding, high frequency welding and resistance welding can be employed. Hardening in the heat-affected zone (HAZ) of welded joints is very low because of their low carbon equivalent. To avoid the occurrence of a so called *soft zone* in the area around the joint heat input has to be limited. From this the cooling time $t_{8/5}$ has to be restricted to 20 s in the case of ALFORM700M®and 10 s in the case of ALFORM900M®.

The presented spectrum of intrinsic as well as processing properties permits the application of ALFORM700M®and ALFORM900M® in a wide field of construction. Both grades have been used with success for more than ten years in the construction of cantilever beams for vehicle cranes and concrete pumps, in the domain of structural engineering, production of vehicles and agricultural implements. ALFORM700M® is also applied in the roll-forming industry for the production of tubes and profiles.

Summary

The high and ultra high strength steel grades ALFORM700M®and ALFORM900M®, which have been developed at voestalpine are introduced in the present paper. Both materials are produced by thermomechanical treatment in a hot rolling mill followed by accelerated cooling to coil temperatures. In the case of ALFORM900M®subsequent annealing is applied. To explain the evolution of the fine microstructure consisting of ferritic bainite, bainite and small amounts of martensite some features of hot rolling like recrystallisaton and phase transformation are investigated by dilatometric experiments. The formation of Ti- and Nb-rich precipitates is discussed by using a numerical model, the precipitated amounts of Ti, Nb and V were measured by chemical methods. The main difference between the two materials lies in the different amount of transformation retarding elements which causes in the case of ALFORM900M®a finer and more transformation hardened microstructure, the contribution of precipitation strengthening is also enhanced for this grade.

In spite of their high strength level ALFORM700M®und ALFORM900M®offer excellent technological and mechanical properties for application in a wide field of construction where weight reduction yields to economical benefits. Bending radii and low temperature toughness are comparable with grades with considerably lower strength levels.

References

[1] T. Gladman: *The physical metallurgy of microalloyed steels* (The institute of materials, 1997)

[2] A.J. DeArdo: Ironmaking and Steelmaking Vol.28 No.2 (2001), p. 138

[3] S.S. Campos, E.V. Morales, H.J. Kestenbach: Metall. Trans. A Vol.32A (2001), p. 1245

[4] R.D.K.Misra et al.: Mater. Sci. Tech. Vol.17 (2001), p.1119

[5] A. Samoilov, B. Buchmayr: Phase transformations during thermal/mechanical processing of steel, ed.E.B. Hawbolt and S.Yue, CIM Vancouver (1995), p.67

[6] J.H.Woodhead: *The physical Metallurgy of Vanadium Steels* (Vanadium Award Papers No.V0284, London 1983)

DOI References
[2] A.J. DeArdo: Ironmaking and Steelmaking Vol.28 No.2 (2001), p. 138
10.1179/030192301678055