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(54) **HIGH STRENGTH COATED DUAL PHASE
STEEL STRIP AND METHOD TO PRODUCE
IT**

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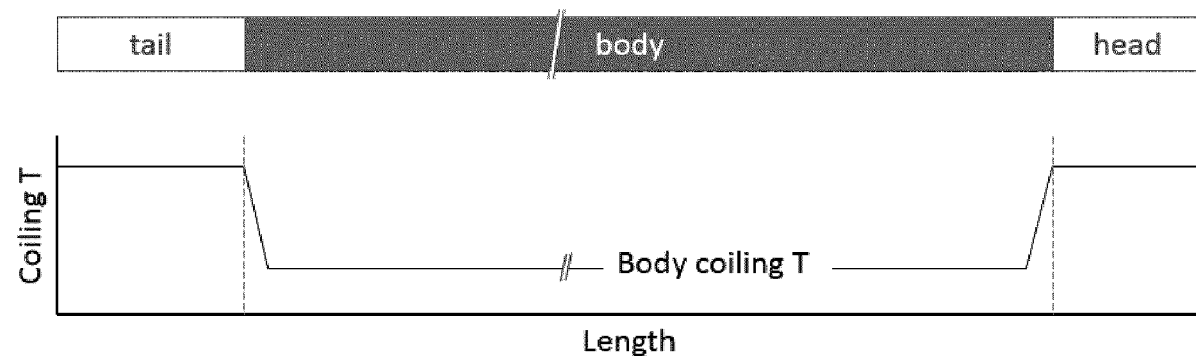
C21D 8/02 (2006.01)

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(57)

ABSTRACT

A coated and optionally temper-rolled dual phase steel strip with a tensile strength R_m of 580-720 MPa and a yield strength R_p of 310-430 MPa high strength having improved formability. The invention also relates to a method for producing such steel strip and the use thereof.



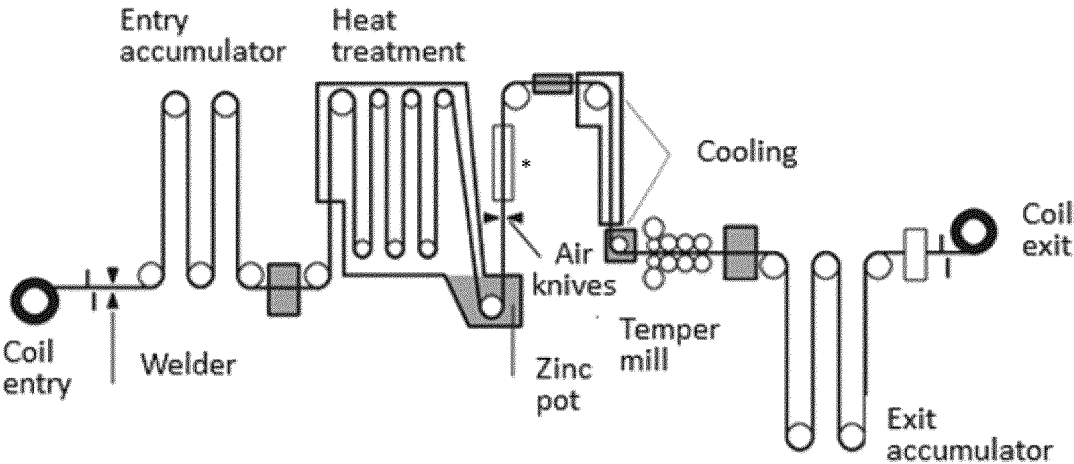


FIGURE 1

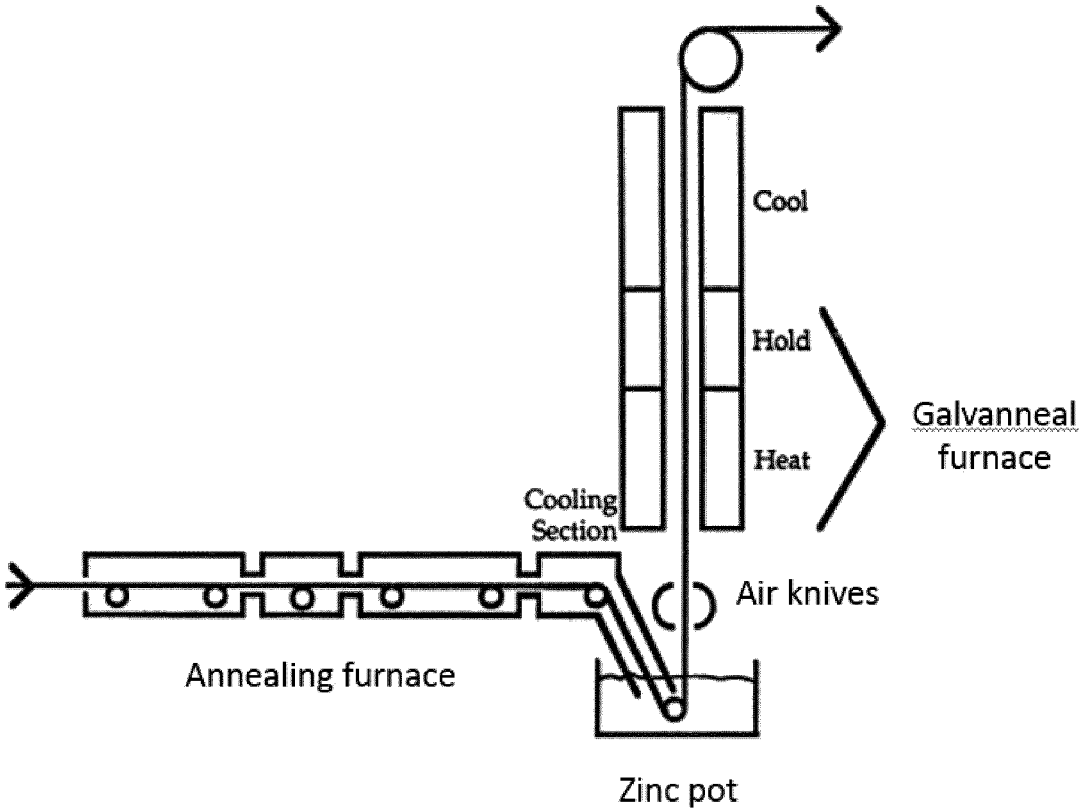


FIGURE 2

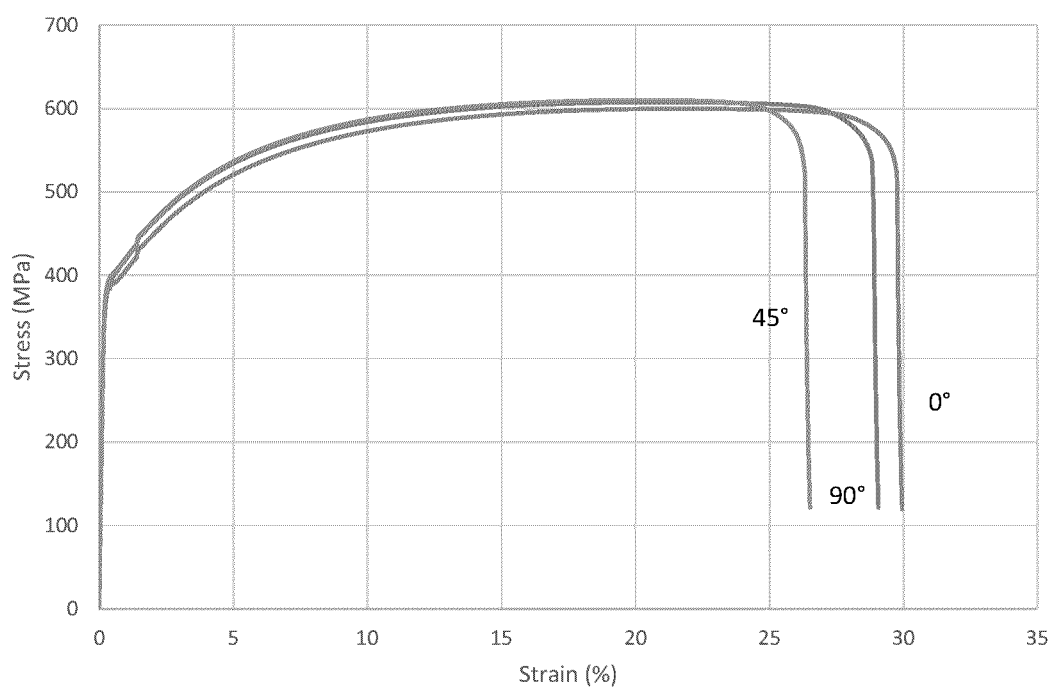


FIGURE 3

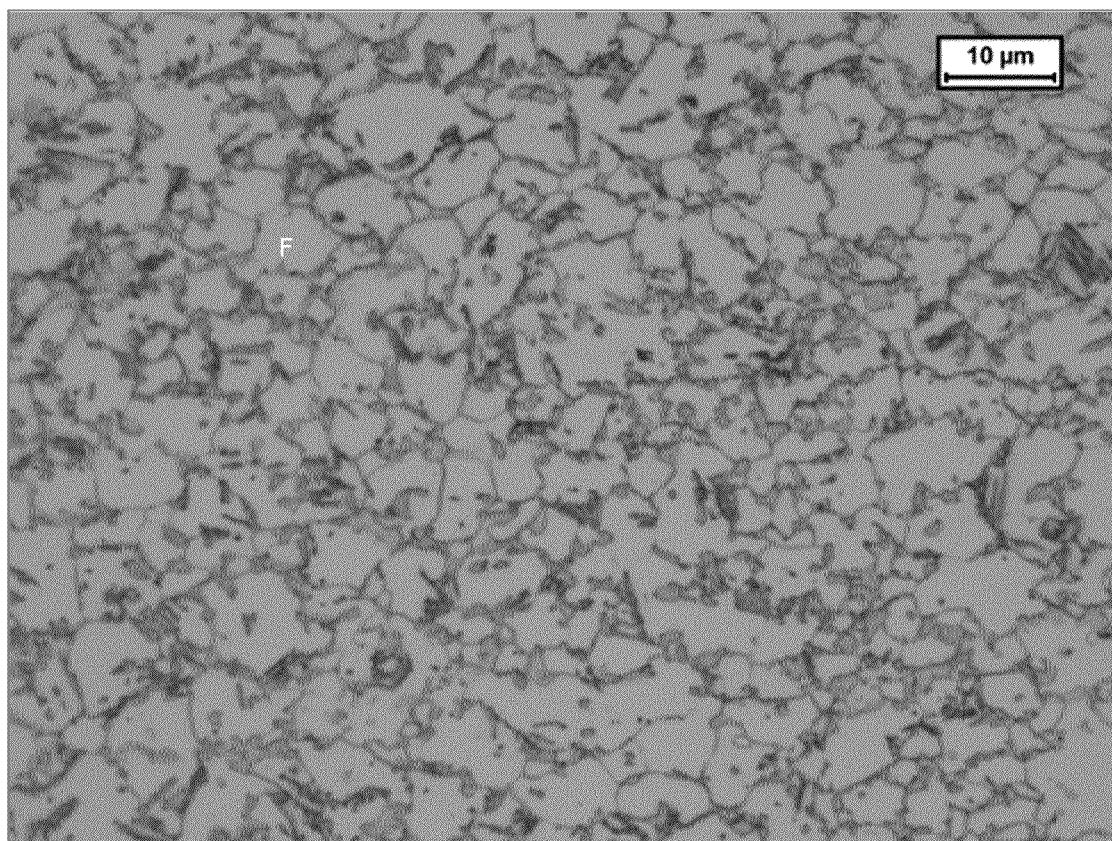


FIGURE 4

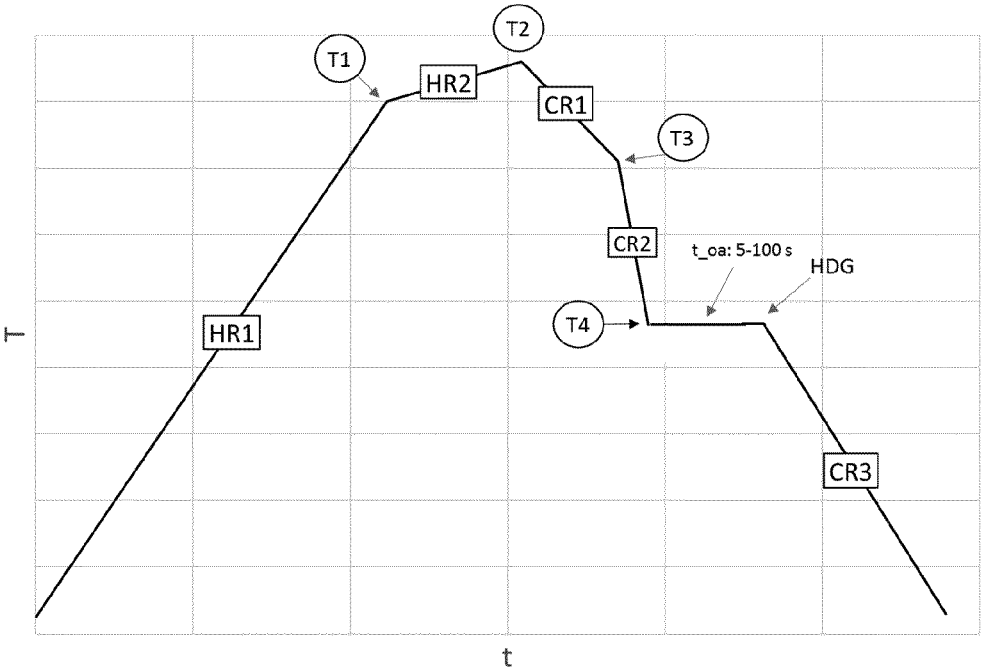


FIGURE 5

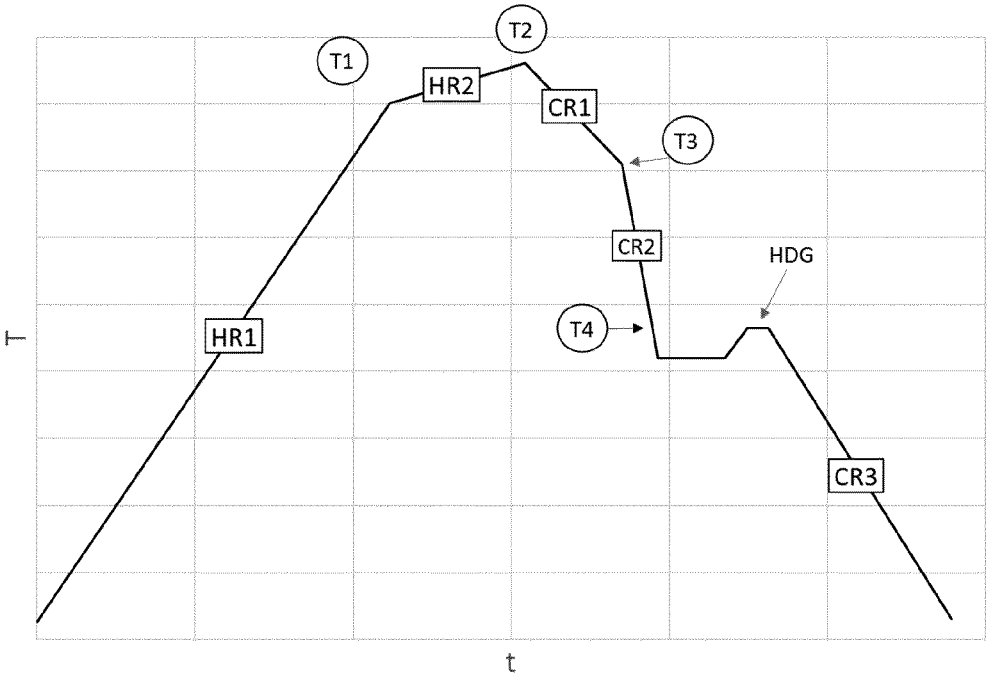


FIGURE 6

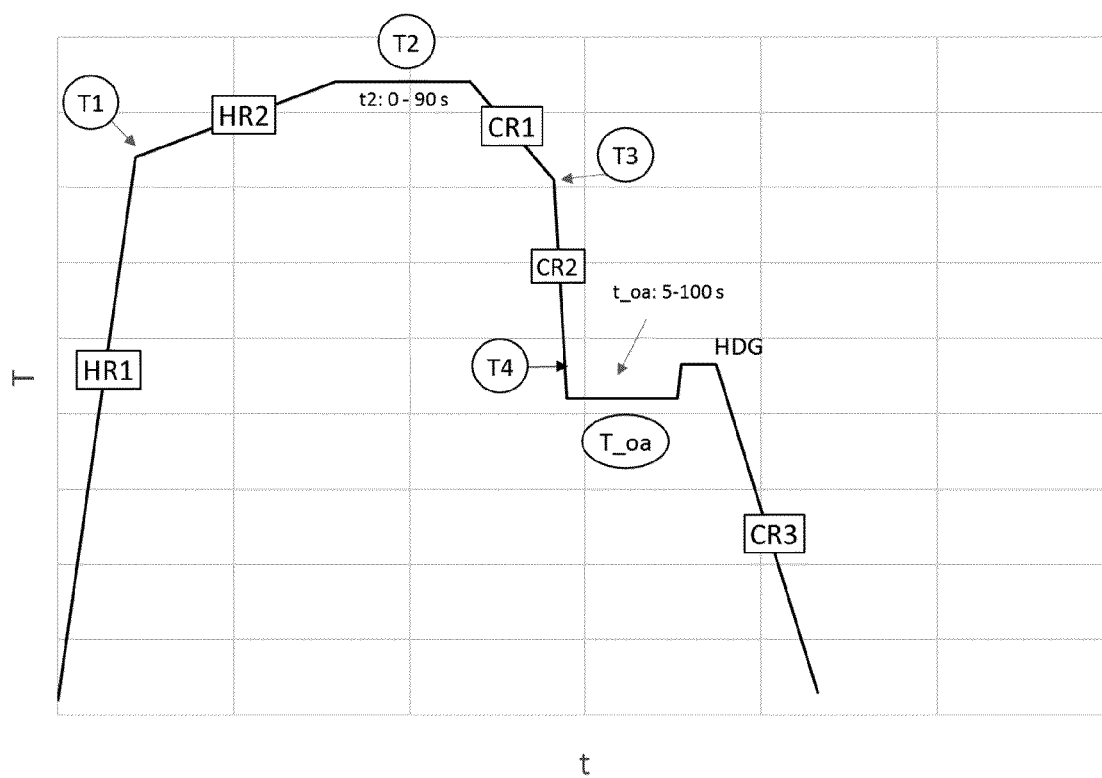


FIGURE 7

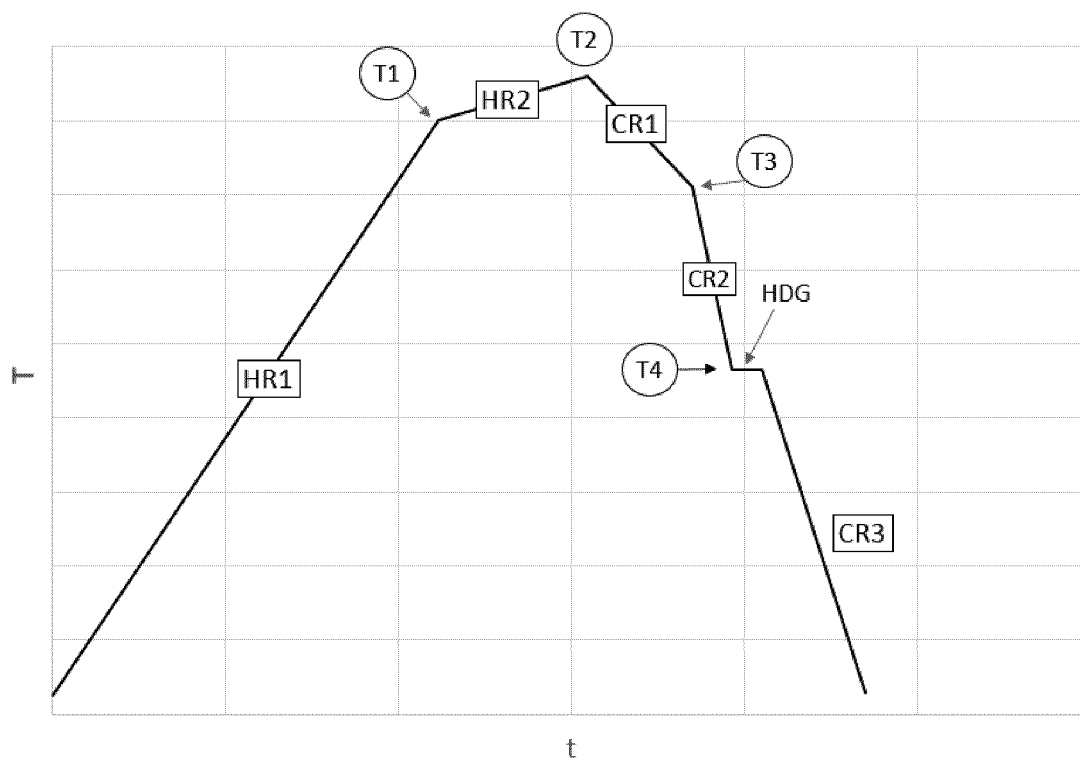


FIGURE 8

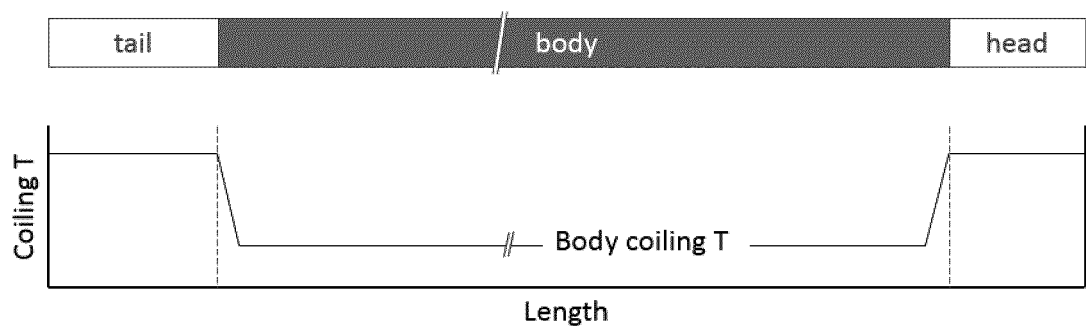


FIGURE 9

HIGH STRENGTH COATED DUAL PHASE
STEEL STRIP AND METHOD TO PRODUCE
IT

FIELD OF THE INVENTION

[0001] The invention relates to a high strength coated dual phase steel strip having improved formability, such as used in the automotive industry. The invention also relates to a method for producing such steel strip.

BACKGROUND OF THE INVENTION

[0002] In recent years, (advanced) high strength steel sheets, AHSS, are increasingly used in car components to reduce weight and fuel consumption. A series of (advanced) high strength steels, such as HSLA, Dual phase (DP), Ferritic-bainitic (FB) including stretch-flangeable (SF), Complex phase (CP), Transformation-induced plasticity

[0005] Dual Phase steels offer a good combination of strength and stampability (or pressability) as a result of their microstructure, in which main constituents of a (mixture of) hard martensitic and/or bainitic phase is dispersed in a ductile ferritic matrix. Small amounts of retained austenite (1-3%) and cementite may be present. These steels have high strain hardenability. This gives them not only good strain redistribution capability and thus stampability, but also finished part mechanical properties, including yield strength, that are far superior to those of the flat metal. The yield strength of Dual Phase steels is further increased by the paint baking (also called Bake Hardening, BH) process. These Dual-phase steels are referred to in VDA 239-100 as CR330Y590T-DP or DH, wherein 330 refers to the minimum yield (Y) strength ($R_p=R_{p_{0.2}}$ in the context of this invention) in MPa, and 590 to the minimum tensile (T) strength (R_m) in MPa. DP stands for Dual Phase and DH for DP with a higher formability (DH=Dehnung Hoch).

TABLE 1

Mechanical properties according to VDA 239-100.									
VDA 239-100		$R_{p_{0.2}}$ (MPa)	R_m (MPa)	A50 Type 1 (%)	A80 Type 2 (%)	A50 Type 3 (%)	n_{4-6} (—)	n_{10-20} (—)	BH2 (MPa)
CR330Y590-T-DP	RD	330-430	590-700	≥21	≥20	≥22	≥0.19	≥0.15	≥30
CR330Y590-T-DH	RD	330-430	590-700	≥28	≥26	≥28	≥0.21	≥0.16	≥30

A80 mm (%): Elongation after fracture using a specimen with gauge length $L_0 = 80$ mm (ISO 6892-1 type 2 (EN20 × 80)).
A50 mm (%): Elongation after fracture using a specimen with gauge length $L_0 = 50$ mm (ISO 6892-1 type 1 (ASTM12.5 × 50) or type 3 (JIS25 × 50)).
A (%) : Elongation after fracture using a proportional specimen with $L_0 = 5.65 \sqrt{S_0}$, wherein S_0 is the starting cross section of the specimen in mm².
Ag (%): Uniform elongation (plastic elongation at maximum force).
BH2 (MPa): Increase in yield strength between a reference condition after a 2% plastic pre-strain and the condition obtained after a heat treatment (170° C.-20 minutes).
 n_{10-20} (—): Strain hardening exponent determined between 10 and 20% plastic strain or Ag if Ag <20%.
 n_{4-6} (—): Strain hardening exponent determined between 4 and 6% plastic strain.
 R_m (MPa): Tensile strength.
 $R_{p_{0.2}}$ (MPa): Yield strength at 0.2% plastic extension (MPa). This is obtained after temper rolling.

(TRIP), Hot-formed, Twinning-induced plasticity (TWIP) has been developed to meet the growing requirements.

[0003] However, AHSS sheet steels cannot be applied easily to a wide variety of car components because their formability is relatively poor. The product is preferably supplied as a coil to an end customer who cuts blanks and shapes these to end parts.

[0004] As steels became increasingly stronger, they simultaneously became increasingly difficult to form into automotive parts. Actually, the real application of AHSS steels (DP, CP and TRIP) to car components is still limited by their formability. Therefore, improving formability and manufacturability become an important issue for AHSS application. On the other hand, affordability is also an important issue. This limits the ability to use expensive alloying elements to achieve the often contradictory requirements that the steels have to satisfy: corrosion protection, various mechanical properties, weldability, optical properties, processability and costs.

Objectives of the Invention

[0006] It is an object of the invention to find a composition of a high strength hot dip galvanised steel strip that strikes a balance between the formability, the homogeneity and the processability of the strip.

[0007] It is a further object of the invention to provide a high strength hot dip galvanised steel strip that has a good coatability by hot dip galvanising.

[0008] It is still a further object of the invention to provide a high strength hot dip galvanised steel strip that has a good weldability.

[0009] It is another object of the invention to provide a high strength hot dip galvanised steel strip that has a good surface quality.

[0010] It is still another object of the invention to provide a high strength hot dip galvanised steel strip having a cost price that is as low as possible.

[0011] It is still another object of the invention that CR330Y590T properties are obtained through coil or over the full coil width and length and coil-to-coil.

DESCRIPTION OF THE INVENTION

[0012] One or more of the objects is reached with the dual-phase steel strip consisting of, in wt. %:

C:	0.090-0.140	B:	0.0010-0.0050
Mn:	1.200-1.900	S:	at most 0.050
Si:	0.200-0.800	P:	at most 0.050
Al:	0.200-0.800	N:	at most 0.015

[0013] and optionally one or more of the elements selected from:

Cr:	at most 0.500	Nb:	at most 0.060
Mo:	at most 0.400	Ti:	at most 0.060
V:	at most 0.200	Ca:	at most 0.0050

[0014] the remainder being iron and unavoidable impurities,

[0015] wherein the steel has a tensile strength R_m of 580-720 MPa and a yield strength R_p of 310-430 MPa. The steel strip is optionally hot-dip galvanised, electrogalvanized or electroplated.

[0016] Preferred embodiments are provided by the dependent claims.

[0017] Dual-phase steels (DP) have of a multi-phase microstructure. Martensite islands are embedded in a ferritic matrix, the proportion of which is used to adjust the strength of the respective grade. In the case of dual-phase steels with improved formability (DH) the microstructure consists mainly of ferrite and martensite and small amounts of bainite and retained austenite. The microstructure and the desired mechanical properties are adjusted by means of the chemical composition and the annealing process optionally combined with hot-dip galvanizing. Dual-phase steels show a very low yield strength to tensile strength ratio (R_p/R_m) combined with high tensile strength and strong work hardening capacity. These dual phase steels are therefore well suited for forming operations with high stretching requirements. Cementite is preferably absent from the microstructure.

[0018] In the context of this invention hot-dip galvanising (hot-dip coating with zinc or a zinc alloy) includes hot-dip aluminizing (hot-dip coating with commercially pure aluminium or an aluminium alloy). Alternatively the annealed strip may be provided with a metallic coating by electrodeposition. In the context of this invention electroplating and electrodeposition are synonyms and are electrolytic processes by which a metallic coating is deposited on another metal. Electrogalvanizing is a species of electrodeposition generally associated with electrodeposition of a zinc or zinc alloy coating on another metal.

[0019] The use of the main alloying elements C—Mn—Si—Al—B, with the optional addition of optional elements, is a cost effective solution for achieving the grade CR330Y590T-DH as prescribed in the European norm VDA 239-100. Some requirements for this grade according to VDA 239-100 are shown in table 1. The steel according to the invention complies with VDA 239-100. The enhanced formability gives it added value in the case of forming critical parts during cold forming operations such as drawing, while retaining the welding performance equivalent to the conventional dual phase grades. Thanks to their

mechanical properties, these grades are suitable for automotive safety parts for crash resistance. By limiting the amount of substitutional elements (such as Mn) good Hole Expansion Coefficient (HEC) values could be obtained. This limitation of Mn also has a beneficial effect on the amount of the post-uniform elongation. Strengthening of the steel is further achieved by the addition of boron and silicon. A disadvantage of using silicon is the risk of cracks in spot welds like for example through liquid metal embrittlement (LME) or though thermal shrink pressure and hence the addition of aluminium is necessary to suppress cracking during spot welding. This is quite surprising as aluminium in general is believed to contribute to LME, but in this carefully balanced chemistry the beneficial effect of the aluminium as a ferrite former on the phase transformation prevails.

[0020] The inventors have found that by a careful selection of the amounts of the main constituting elements of the steel, being carbon, manganese, silicon, aluminium, chromium and boron, a high strength hot dip galvanised steel strip can be produced that has the required formability, homogeneity, processability, strength and elongation, while at the same time providing a sufficient weldability, coatability and surface quality.

[0021] The reason for the amounts of the main constituting elements in the steel according to the invention is as explained below (all compositions throughout are given in weight percent (wt. %) unless indicated otherwise). It is noted that, although the alloying elements as given below must be applied in a balanced manner to achieve the desired results, the elements may be varied independently of each other within the boundaries as described for the individual elements herein below.

[0022] C: 0.090-0.140 wt. %. Carbon is an essential cost-effective alloying element in the steel grade to achieve the strength levels in a conventional continuous annealing/galvanising line. Carbon has to be present in an amount that ensures the required strength and elongation. To ensure hardenability and the resulting formation of martensite at the cooling rates available in a conventional annealing/galvanising line was found to require at least 0.090 wt. %. A maximum level of 0.140 wt. % has been found to be essential to secure sufficient weldability and elongation. Preferably the C is at least 0.094 wt. %, more preferably 0.098 wt. %, more preferably 0.104 wt. % and even more preferably at least 0.110 wt. %. The carbon content is preferably at most 0.130 wt. %, and more preferably at most 0.125 wt. %. In a preferable embodiment C is from 0.090 wt. % to 0.130 wt. %.

[0023] Mn: 1.200-1.900 wt. %. Manganese improves the hardenability of the steel so that it promotes the formation of bainite and martensite. Manganese strengthens the ferritic matrix by solid solution strengthening and lowers the transformation temperature range of the steel, thus lowering the required annealing temperature. As an austenite stabilising element, it promotes retained austenite formation. Mn is known to adversely affect the HEC-values and through coil ductility variation as well as the rolling forces during hot- and cold-rolling and therefore Mn-values should be below 1.900 wt. % and preferably below 1.800 wt. %. A suitable maximum value for Mn is 1.750 wt. %. Mn values below 1.700 wt. % or even below 1.600 wt. % further improve HEC-values and ductility. These maximum levels are also given in view of the stronger segregation during casting and

centre line segregation in the strip at higher values. The product should further be processable in continuous annealing lines with both RTF heating and faster flame heating. An excess of 1.7 wt. % critically delays cold rolling recrystallisation and subsequent ferrite grain growth during continuous annealing, especially if faster flame heating like in a DFF (Direct Firing Furnace) or NOF (Non-Oxidising Furnace) or any other fast heating method like induction heating. Given the unavoidable variations in heating over the coil width and length this may adversely affect ductility over the full coil width and length. The minimum Mn value is 1.200 wt. %, preferably at least 1.300 wt. %, and more preferably at least 1.400 wt. % and most preferably at least 1.450 wt. %. If the Mn-level is too low then the required strength level is not achieved while DP properties with low Rp/Rm ratio are not achieved at lower values. This lower limit is possible in view of the addition of other elements, such as boron. These values are reached more easily at higher values of Mn, but the costs and elongation values deteriorate, so that a proper balance must be chosen. In a preferable embodiment Mn is from 1.400 wt. % to 1.750 wt. %.

[0024] Si: 0.200-0.800 wt. %. Silicon suppresses cementite formation in the end product and stabilizes retained austenite formation thereby improving the ductility of the product. Silicon effectively reinforces the steel matrix by solid solution to achieve the required strength level and increases the work-hardening rate of the ferrite component, which improves local formability (HEC) and a high total elongation. The maximum Si-content in the steel according to the invention is 0.800 wt. %. It was however observed that Si adversely affects the spot-weldability and may be prone to liquid metal embrittlement of the coated strip and therefore the maximum is preferably limited to 0.750 or even 0.700 wt. % or most preferably up to 0.650 wt. %. In addition, an excess silicon may lead to a too high degree of internal and grain boundary oxidation during coil cooling. This may be circumvented by using a very low coiling temperature and batch annealing but it is outside the scope of a cost-effective solution in terms of process steps.

[0025] To achieve a minimum strengthening effect at least 0.200 wt. % is required, and preferably at least 0.250 wt. %. By balancing Si with the other elements, such as e.g. Al) a minimum Si-level of at least 0.400 wt. % and even more preferably at least 0.450 wt. % may be preferred. In a preferable embodiment Si is from 0.500 wt. % to 0.700 wt. %.

[0026] Al: 0.200-0.800 wt. %. Aluminium is primarily added to liquid steel for the purpose of de-oxidation in relatively small amounts. The excess Al also reacts with nitrogen and forms AlN precipitates which hence prevents boron from reacting with nitrogen. As a result B segregates on the grain boundaries and hence acts as a hardenability element. To achieve at least the binding of free oxygen and nitrogen, a lower level of Al of at least 0.060 wt. % is proposed.

[0027] At higher contents Al slows bainite transformation, thus determining the optimum bainite formation within the time constraints imposed by the annealing section of a continuous annealing/galvanising line. The Al is given as the total aluminium content. Aluminium also retards the formation of carbides thus keeping carbon in solution, which in turn causes partitioning of the carbon to austenite during overaging and promotes the stability of austenite. Unlike

Silicon, Al hardly increases the partitioning time for austenite carbon enrichment in the continuous annealing line overage section. This improves the elongation values and the hardenability and also allows a minimum or no overage time in the continuous annealing line while still achieving the required. A maximum level of 0.800 wt. % is imposed for castability, since high aluminium contents lead to poisoning of the casting mould slag and consequently an increase in mould slag viscosity, leading to incorrect heat transfer and lubrication during casting.

[0028] During casting, aluminium segregates resulting in carbon, silicon and manganese enriched segregation bands with intermediary lower Mn high aluminium rich segregation bands. Both effects increase the elongation and the hardenability value ($n_{4.6}$ (which is the strain-hardening exponent in the strain range of 4 to 6%) and n -values (which is the strain-hardening coefficient)) of the end product while keeping the yields stress low. In addition, as Al causes a segregation of Mn and Si it also increases strength. A further beneficial effect of aluminium is that it is less sensitive to crack formation during welding of zinc (alloyed) coated strip like and therefore reduces the sensitivity for liquid metal embrittlement. It can therefore to a limited extent replace Si.

[0029] The amount of aluminium affects intercritical annealing and the phase distribution during intercritical annealing. This can be understood by aluminium being a ferrite former. For the above reasons, Al is preferably above 0.200 wt. % and more preferably a minimum Al content of 0.25 wt. %. Preferably the aluminium content is at most 0.750 wt. %, more preferably at most 0.650 wt. %, and even more preferably at most 0.400 wt. %. In a preferable embodiment Al is from 0.200 wt. % to 0.400 wt. %.

[0030] The sum of aluminium and silicon is preferably between Al+Si=0.350-1.200 wt. % and more preferably between 0.600-1.200 wt. %. The combination of both elements suppresses carbide formation in the end product and stabilisation of sufficient austenite. Silicon is known to increase the optimum time for retained austenite formation in the overage section while Aluminium hardly affects the partitioning time for retained austenite formation. Using the correct compositions the sum of Al and Si allows obtaining the desired extension of formability. Its use should be balanced as it also determines the hot-rolling non-recrystallisation temperature (TNR). Silicon reduces the TNR while Al increases the TNR. TNR is preferably as low as possible to allow thermo-mechanical rolling to ensure a further desirable extension in formability through improved r -values. Preferably Al+Si is between 0.650 and 1.100 more preferably 0.700 and 1.100, more preferably between 0.750 and 1.050 wt. % and most preferably between 0.800 and 0.950 wt. %.

[0031] B: 0.0010-0.0050 wt. % (5-50 ppm by weight). Boron is added as a hardenability element to achieve the required strength level. The addition of boron allows a saving on the use of more expensive alloying elements. Compared to more expensive solid solution alloying elements, boron affects the ability to recrystallisation much less. This allows for addition static or dynamic recrystallisation in the hot rolling line, which reduces mean flow stress build up. It also increases the cold rolling recrystallisation capability during the heating section of the continuous annealing thus allowing sufficient grain growth and achieving the DH600 properties both in a direct firing and slow

tube heating annealing line. Since Boron affects the properties of the final product but not the rolling forces, the use of boron instead of solid solution hardening elements like Mn and Si this results in an improved dimensional window for the steel strip, meaning a higher width to thickness ratio while the mechanical properties of the steel over the width of the strip remain suitable. Boron refines retained austenite, thus stabilising formed retained austenite. Boron improves weldability because it segregates to the grain boundaries and partially replaces phosphorus, which makes it possible to allow a somewhat higher P amount in the steel without sacrificing weldability.

[0032] The boron amount is 0.0010-0.0050 wt. % (10-50 ppm). A suitable lower boundary level is 0.0012 wt. % or even 0.0015 wt. %. Above a certain boron level zinc coating is adversely affected while hardenability reduces. A suitable upper boundary level is 0.0040 wt. %, preferably 0.0035 wt. %, more preferably 0.0030 or even 0.0025 wt. %. In a preferable embodiment B is from 0.0015 to 0.0025 wt. %.

[0033] N: at most 0.015 wt. % (150 ppm). Nitrogen is present during casting. Free nitrogen in the steel should be avoided as it can lead to severe ageing and affects the capability of boron to function as a hardenability element. However, in the chemical composition of the steel according to the invention there is a surplus of aluminium and boron and consequently all nitrogen is bound to Al or B as AlN or BN. The N-content is limited to 0.015 wt. %, typical for BOS-steelmaking and continuous casting plants. However, it is preferably to reduce the nitrogen to a value of at most 0.012 wt. %, preferably to at most 0.010 wt. %. A more preferred maximum level is 0.008 wt. % or even 0.006 wt. %. It is not economic to remove all nitrogen from the steel and therefore in a preferable embodiment N is from 0.0001 wt. % to 0.006 wt. % (10 to 60 ppm).

[0034] P: at most 0.050. Phosphorus interferes with the formation of carbides, and therefore some phosphorus in the steel can be advantageous. In addition, phosphorus positively affects picklability and strengthens the steel. A maximum of 0.050 wt. % is allowable. However, phosphorus can make steel brittle upon welding, so the amount of phosphorus is preferably limited to 0.040 wt. % to maintain sufficient hot ductility and to avoid failure by peel during tension-shear tests carried out on spot-welded assemblies. More preferably the maximum level is of P is 0.030 wt. %, even more preferably 0.020 mass %. In a preferable embodiment P is at most 0.015 wt. %.

[0035] S: at most 0.050 wt. %. Sulphur (S) is maximised at a value of 0.050 wt. % and preferably maximised at a value of 0.020 wt. % or even at a value of 0.010 wt. %. Even though sulphur may somewhat improve picklability, sulphur is preferably avoided completely but is inevitably present as a result of the steelmaking process and the raw materials used therein. Sulphur precipitates in the form of manganese and/or calcium sulphides that greatly reduce the formability. Preferably the amount of sulphur is 0.0001-0.005 wt. %, more preferably at most 0.003 wt. % or more preferably at most 0.002 wt. %. A suitable minimum amount is 0.0002 wt. %. In a preferable embodiment S is from 0.0001 wt. % to 0.002 wt. %.

[0036] Optionally one or more elements selected from Ti, V, Cr, Mo, Nb, Ni, Cu, Ca, may be added to the steel composition or be present as impurities.

[0037] The addition of Calcium (Ca) modifies the morphology of manganese sulphide inclusions. When calcium is

added the inclusions get a globular rather than an elongated shape. Elongated inclusions, also called stringers, may act as planes of weakness along which lamellar tearing and delamination fracture can occur. The avoidance of stringers is beneficial for forming processes of steel sheets which entail the expansion of holes or the stretching of flanges and promotes isotropic forming behaviour. Calcium treatment also prevents the formation of hard, angular, abrasive alumina inclusions in aluminium deoxidised steel types, forming instead calcium aluminate inclusions which are softer and globular at rolling temperatures, thereby improving the material's processing characteristics. In continuous casting machines, some inclusions occurring in molten steel have a tendency to block the nozzle, resulting in lost output and increased costs. Calcium treatment reduces the propensity for blockage by promoting the formation of low melting point species which will not clog the caster nozzles. It is also possible to add no calcium when the sulphur content is very low. The maximum for Ca elements is set at 0.0050 wt. % (=50 ppm) and preferably at 0.0030 wt. % and more preferably at 0.0020 wt. % or even at 0.0010 wt. %. In a preferable embodiment Ca is at most 0.0010 wt. %. Magnesium or Rare Earth Metals (REM) may be added for similar reasons as Ca. A maximum for these elements is set at 500 ppm for Mg and 50 ppm for REM.

[0038] Chromium (Cr) is added to increase hardenability. It promotes formation of ferrite. A maximum level of 0.500 wt. % is imposed to ensure that not too much martensite forms at the cost of retained austenite. It is also possible to add no Cr, in which case it is absent completely or present at most at a residual element or impurity level. Cr is quite expensive. From that perspective the amount of Cr is preferably at most 0.300 wt. %, more preferably at most 0.250 wt. % and even more preferably at most 0.200 wt. %. From a mechanical property point of view, if Cr is added as an alloying element, the preferable range is between 0.020 and 0.400 wt. %. A good balance between obtaining the desired microstructure, the desired mechanical properties and costs is obtained when Cr is between 0.020 and 0.150 wt. %, and preferably between 0.020 and 0.120 wt. %. If added as an alloying element, a suitable minimum amount of Cr is 0.030 wt. %, or even 0.040 wt. %. Values below 0.020 are deemed to be at residual element level. When an element is at residual element level, the elements are deemed not to affect the properties or processability of the steel in a significant way, and in these cases the costs of removing these elements further exceed the expected benefits of a further reduction, even if it were technically feasible to reduce the level further.

[0039] Titanium (Ti) is mainly added to strengthen the steel, and the allowable level is at most 0.060 wt. %. It is also possible to add no Ti, in which case it is absent completely or present at most at a residual element or impurity level. If added, then preferably the amount of Ti is between 0.010 and 0.060 wt. %, more preferably between 0.02 and 0.050 wt. %. Values below 0.010 wt. % are deemed to be at residual element level.

[0040] Vanadium may be added up to 0.200 wt. %. If added a preferred range is 0.005-0.200 wt. % When the vanadium content is less than 0.005% by weight vanadium precipitation strengthening is insufficient. When the vanadium content is greater than 0.200 wt. % the precipitation occurs at an early stage during or after the hot-rolling in the form of fine precipitates, which reduces the dimensional

window or coarsen during continuous annealing which reduces precipitate strengthening. The vanadium amount is 0.005-0.200 wt. %, preferably 0.010-0.200 wt. %, more preferably 0.030-0.200 wt. %, most preferably 0.040-0.150 wt. %. It is also possible to add no V, in which case it is absent completely or present at most at a residual element or impurity level. Values below 0.010 wt. % are deemed to be at residual element level.

[0041] Optionally, elements such as molybdenum, which retards the bainitic transformation and promotes solid-solution hardening, may be added in amounts not exceeding 0.40 wt. %. The amount of Mo is preferably 0.005-0.20 wt. %, more preferably 0.005-0.10 wt. %, so as to limit the cost of the steel and keep the dimensional window as large as possible. Mo can be added to improve to strength and improves the quality of a zinc-based coating. Mo also helps to strengthen the steel through precipitate formation. However, it is also possible to add no Mo, in which case it is absent completely or present at most at a residual element or impurity level. Values below 0.015 wt. % are deemed to be at residual element level.

[0042] Niobium (Nb) may be added in an amount of preferably 0.001-0.060 wt. %, more preferably 0.001-0.050 wt. %, and most preferably 0.001-0.030 wt. %. The addition of Nb increases the strength by complementary carbonitride precipitation but adds to hot-rolling forces which reduce the dimensional window. If added, then preferably the amount of Nb is between 0.010 and 0.060 wt. %, more at most 0.050, or at most 0.030 wt. %. Values below 0.010 wt. % are deemed to be at residual element level.

[0043] Consequently, in an embodiment a dual phase steel strip according to any one of the preceding claims, wherein one or more or all of Cr, Mo, V, Nb and Ti is present only as an impurity this means that, if applicable, the levels of one or more or all of Cr is below 0.020 wt. %, Ti is below 0.010 wt. %, V is below 0.010 wt. %, Mo is below 0.015 wt. % and Nb is below 0.010 wt. % respectively.

[0044] Nickel and copper are not added to the steel according to the invention and these elements are either absent completely or present at most at a residual element or impurity level. Values below 0.060 wt. % are considered to be at residual element level.

[0045] In an embodiment the dual phase steel strip according to the invention has a tensile strength R_m of 590 to 720 MPa and/or a yield strength R_p of 320 to 430 MPa. Preferably the steel according to the invention has a yield strength R_p of at least 330 MPa. Preferably the steel according to the invention has a tensile strength R_m of at most 700 MPa and a yield strength R_p of at most 430 MPa.

[0046] According to a second aspect of the invention there is provided a method for producing a coated and optionally temper-rolled dual phase steel strip with a tensile strength R_m of 580-720 MPa and a yield strength R_p of 310-430 MPa according to any one of claims 1 to 11, comprising the following steps:

[0047] providing a hot-rolled steel strip by hot rolling a continuously cast slab with a composition consisting of, in wt. %:

C:	0.090-0.140	B:	0.0010-0.0050
Mn:	1.200-1.900	S:	at most 0.050
Si:	0.200-0.800	P:	at most 0.050
Al:	0.200-0.800	N:	at most 0.015

[0048] and optionally one or more of the elements selected from:

Cr:	at most 0.500	Nb:	at most 0.060
Mo:	at most 0.400	Ti:	at most 0.060
V:	at most 0.200	Ca:	at most 0.0050

[0049] the remainder being iron and unavoidable impurities,

[0050] to a hot-rolled strip having a thickness of 2.0-4.5 mm, wherein finish-rolling is performed while the strip has an austenitic microstructure;

[0051] Cooling the hot-rolled strip after finish-rolling with a cooling rate of at least 30° C./s;

[0052] Coiling the cooled strip at a coiling temperature CT in the range of 500 to 650° C. wherein the amount of austenite is more than 60% and the coiled strip is subsequently allowed to cool to ambient temperatures;

[0053] Uncoiling the coiled hot-rolled strip, followed by pickling and cold-rolling with a reduction of 40-80%;

[0054] Continuous annealing of the cold-rolled strip by

[0055] i. heating the strip to a temperature T₁ in the range of 550-710° C. with a heating rate HR₁ of at least 3° C./s to allow the recrystallisation of the cold-rolled strip to proceed for at least 60% prior to the start of the austenite formation;

[0056] ii. further heating of the strip with an average heating rate HR₂ to a temperature T₂ in the range of (Ac₁+50° C.) to (Ac₃-30° C.) to form a partially austenitic microstructure, preferably wherein HR₂ is in the range of 2-20° C./s;

[0057] iii. subsequently either

[0058] a. keeping the strip at T₂ for a time period t₂ of at most 90s, followed by slow cooling of the strip with a cooling rate CR₁ in the range of 0.5-12° C./s to a temperature T₃ in the range of 600-790° C., preferably 620-790° C. so that 16-30% austenite is present in the slow cooled strip, or

[0059] b. immediately slowly cooling the strip from T₂ with a cooling rate CR₁ in the range of 0.5-12° C./s to a temperature T₃ in the range of 600-790° C., preferably 620-790° C. so that 16-30% austenite is present in the slow cooled strip;

[0060] iv. rapidly cooling the slowly cooled strip comprising 16-30% austenite with an average cooling rate CR₂ in the range of 5-70° C./s either

[0061] a. to a temperature T₄ in the range of 330-470° C.;

[0062] b. to an overaging temperature T_{oa} for a time t_{oa} in the range of 5 to 100 s, wherein T_{oa} is between 390 and 465° C.;

[0063] v. cooling the steel strip at a cooling rate CR₃ of at least 4° C./s to a temperature below 300° C.;

[0064] wherein the strip is a). provided with a metallic coating by hot-dip coating between step iv. and v.,

optionally followed by galvannealing the hot-dip coated strip, or b). with a metallic coating by electrodeposition after step v.;

- [0065] Optionally temper rolling or tension levelling of the coated steel strip with a reduction of at most 0.70%;
 - [0066] Wherein the coated and optionally temper-rolled steel strip has a tensile strength R_m of 580-720 MPa and a yield strength R_p of 310-430 MPa;
 - [0067] Coiling the coated steel strip or cutting the coated steel strip into sheets or blanks;
 - [0068] Optionally shaping the coated steel strip, sheet or blank via cold-forming operations like stamping, bending, deep drawing or via warm press forming or hot press forming.
- [0069] Preferred embodiments are provided by the dependent claims.

Steelmaking and Casting

[0070] The steel melt is preferably produced in a BOS-process (Basic Oxygen Steelmaking). This process is preferable over the Electric Arc (EA)-process because of its ability to control the chemistry and the lower levels of unavoidable impurities that can be reached in the BOS-process compared to the EA-process. In the context of this invention unavoidable impurities and residual elements mean the same in that the level of unavoidable impurities and the level of residual elements are determined by the technical or economic incapability to lower the level of an element below these levels.

[0071] The steel melt is continuously cast into slabs on a conventional caster. After casting the strip cools in the continuous caster. During cooling of the continuous cast slab goes to a specific temperature range while being bent in which the ductility of the steel is reduced which is known as the temperature ductility trough. In this temperature range precipitates may form. One of the precipitate forming element is Al, which forms AlN at grain boundaries. This should partly be suppressed as it may lead to sliver formation later in the process. By adding B AlN is partially suppressed as BN forms preferentially.

[0072] The cast slab may be a thick slab (150-350 mm thick) or a thin slab (50-150 mm). Inhomogeneous excessive grain growth can be suppressed by the formation of BN and/or AlN (mixed) precipitates. Part of the BN precipitates are dissolved while AlN precipitates form.

Hot-Rolling

[0073] Subsequently the slab is hot-rolled. The thick slabs are hot-rolled in a conventional hot-strip-mill (HSM). The thin slabs are usually directly hot-rolled after casting in a thin slab caster and (direct) rolling facility (TSCR). Before hot-rolling the thick cast slabs in a HSM the slabs have to be reheated to a temperature of 1150° C. or higher. Before hot-rolling the thin cast slabs in a TSCR the slabs have to be homogenised in temperature at a temperature of about 1150° C. Both types of hot-rolling process are equally applicable to hot-roll the steel according to the invention.

[0074] Although it is not essential, the inventors found that it is preferable that the majority of the hot-rolling reduction, starting from the cast slab, is applied above TNR.

[0075] It is preferable that, when starting from cast thick slab (150-350 mm), at least 50%, more preferably at least 75% and even more preferably at least 80% or even 85% of

the total hot-rolling reduction (slab thickness to hot-rolled strip thickness) is performed above the non-recrystallisation temperature (TNR). Starting from cast thin slab (50-150 mm) it is preferable that at least 50%, more preferably at least 60% and even more preferably at least 70% or even 75% of the total hot-rolling reduction is performed above the non-recrystallisation temperature (TNR) and that full recrystallisation occurs in the first stand of the finishing mill. Preferably at least 40% and more preferably at least 50% of the hot-rolling reduction (transfer bar thickness to hot-rolled strip thickness) in the tandem finishing mill is performed above TNR. A higher reduction above TNR leads to a finer austenite structure due to repeated static or dynamic recrystallisation and thereby to a finer final structure of the hot-rolled coil with a more evenly distributed fraction of retained austenite in the ferrite matrix. For the invented alloy the equation used is (compositions in wt. %):

$$TNR (^{\circ}C.) = 887 + 464 * [C] + 363 * [Al] - 357 * [Si]$$

[0076] TNR is preferably lower than 1050° C. and the finish rolling temperature (FRT) is preferably at least 880° C. The steel must still be in its austenitic state (i.e. above A_{c3}) during the last hot-rolling reduction step to promote the r-value and hence local ductility by optimizing the final ferrite crystallographic grain orientation.

[0077] After finishing the hot-rolling the strip is cooled on the run-out table at an average cooling rate which is preferably above 50° C./s to the coiling temperature. The coiling temperature specified in this invention relates to the aimed coiling temperature of the body of the strip. Due to the effects of cooling of the coiled hot-rolled strip the head and tail of the strip cools faster on the coil than the body of the strip. The head and tail form the inner and outer wraps of the coil. To compensate for the faster cooling of the head and tail of the hot-rolled strip it may be preferable to aim for a higher coiling temperature of the head and tail by employing a so called U-type cooling pattern.

[0078] The body coiling temperature after hot-rolling is preferably below 610° C. to avoid surface grain boundary oxidation and preferably below 600° C., or even below 590 or 580° C. The lower the coiling temperature the lower the risk of surface grain boundary oxidation. In addition, pearlite/cementite starts to form in band-like morphology and this is also preferably avoided. In addition, the mid-coil strength shows a considerable drop and the alloy composition may have the risk of not achieving the right strength at mid-coil position. On the other hand, too low a coiling temperature leads to hard edges and a reduction of elongation at the coil head and tail edges. Hence the body coiling temperature is 500° C. or higher and preferably 520° C. or higher. Any adverse tail and head effects may be compensated by a suitably chosen higher head- and or tail coiling temperature.

Cold-Rolling

[0079] The strip is pickled and subsequently cold-rolled. Due to the alloying elements in solid solution the rolling forces are considerable. The cold-rolling reduction is therefore at most 80%, preferably at most 75% and more preferably at most 70% and even more preferably at most 65%. Higher reductions lead to high rolling forces, and thus to a higher risk of shape defects and a smaller dimensional

window. A smaller dimensional window means a lower width to thickness ratio while the mechanical properties of the steel over the width of the strip remain suitable. The cold rolling reduction also influences the final microstructure after annealing. If the cold rolling reduction exceeds a certain (composition dependent) threshold the microstructure is deformed to such an extent that the microstructure loses its potential to form retained austenite during the annealing treatment. This small amount of retained austenite is crucial for obtaining the desired formability and hence the cold rolling reduction is critical. In addition, high cold rolling reductions result in smaller ferrite grains which may negatively affect $n4_6$ values.

Continuous Annealing

[0080] Although all preceding process steps contribute to the final properties of the steel, the continuous annealing step is a last and crucial step in obtaining the desired final properties.

[0081] The cold rolled strip is heated at an average heating rate of between 3-30° C./s in a Non-Oxidising Furnace (NOF), a direct fired furnace (DFF), an induction furnace, a radiation tube furnace (RTF) or heating by hot gas. During heating it is important that overlap between recrystallisation of the cold-rolled microstructure and the austenite formation is low. At least a substantial part and preferably all of the deformed microstructure has recrystallised before the transformation to austenite starts. This is because the unrecrystallised ferrite grains recover prior to recrystallisation, and carbon in these recovered grains tends to diffuse to the low-angle grain boundaries in the recovered grains. The recrystallised grains do not “cling” to its carbon, and the carbon diffuses to the austenite that has formed and due to the lower carbon content of the ferrite, the recrystallised grains tend to transform before the unrecrystallised grains. As the annealing cycle is an intercritical annealing cycle, this means that the last remaining ferrite is the unrecrystallised ferrite with the higher carbon content. This carbon does not contribute to the formation of martensite during cooling after annealing, and the carbon does also not help in stabilising retained austenite, which is deemed essential for the formability. The unrecrystallised ferrite moreover adversely affects the formability.

[0082] The strip surface is pre-oxidised in the section of the furnace to allow the formation of an oxide layer on the surface with sufficient adhesion which is suitable for galvanizing. The composition of the steels according to the invention suffer from poor wettability of the molten metal during hot-dip galvanizing. By oxidising the surface with e.g. steam, HNX, water vapour or by a controlled dew point. The iron oxide film also contains some manganese oxide. The other oxides from the alloying elements such as Si and Al, which are not soluble in iron/manganese oxides, are rejected at the steel/oxide interface. Then, following the reduction of the oxide in the section of the furnace with a reducing atmosphere reduction, the oxidised alloying elements do not totally cover the surface but are located in nodules, which enables better wettability and coating adhesion.

[0083] The maximum annealing temperature (T_{top} (T2)) is chosen between ($Ac1+50^{\circ}$ C.) to ($Ac3-30^{\circ}$ C.). The phase transformation temperatures during reheating ($Ac1$ and $Ac3$) and during cooling (e.g. bainite and martensite transformation temperatures) can be simply determined by means of

simulating the thermal profile as schematically depicted in FIGS. 5 to 8 in a dilatometer. A preferred temperature range for T_{top} (T2) is between 790° C. and 880° C. The strip may be held at T_{top} (T2) for a soaking period of up to 300 seconds. Preferably the soaking period is at most 90 s. Alternatively T_{top} (T2) may be a peak temperature wherein the temperature is held above 790° C. for a soaking period of up to 300 seconds and wherein the temperature is slowly raised to T_{top} and slowly decreased from T_{top} (T2). After the soaking period the strip is slowly cooled to T_{slow} (T3) which is between 9° C. to 100° C. lower than the top temperature. During cooling epitaxial ferrite is formed while the remaining austenite enriches in carbon. It is important that T_{slow} (T3) (cooling from T2 to T3 is the slow cooling section) is higher than 650° C. so that enough austenite remains which can be transformed to bainite, martensite or further enriched in retained austenite. A minimum amount of austenite of at least 15-20% ensures that the required 590 MPa is achieved. The cooling rate after the slow cooling section (CR2) is between 5 and 70° C./s and preferably at least 30° C./sec so that austenite is prevented from forming carbides.

[0084] In the invention the properties should both be attainable with or without overageing. A direct cooling from T_{slow} (T3) to the molten metal bath may be used (see FIG. 8), followed by hot-dip galvanizing, a gas jet wiper and cooling to ambient temperature. The hot-dip galvanised strip may optionally be annealed prior to cooling to ambient temperature (galvannealing). The galvanizing may be followed by an annealing treatment, known as galvannealing.

[0085] Overageing may have the advantage that additional bainite is formed and that the existing microstructure is tempered. Carbon partitioning may take place for retained austenite formation. Also hydrogen dissolved in the strip may be released, at least partly and strip tensions may also be partly relieved.

[0086] In an alternative the strip is cooled from T_{slow} (T3) to and held at an overageing temperature (T_{oa}) at a cooling rate (CR2) of at least 15° C./s between 370 and 470° C. for an overageing period of up to 50 seconds, subsequently optionally heated or cooled to the molten metal bath temperature, hot-dip galvanised by dipping the strip in the molten metal bath (such as depicted in FIGS. 5, 6 and 7). After dipping the galvanised strip is cooled to ambient temperature or optionally annealed and cooled to ambient temperature.

[0087] The molten metal bath may contain molten zinc, a molten zinc-alloy, molten aluminium or a molten aluminium alloy. The zinc alloy may comprise 0.3-4.0 wt. % Mg and 0.3-6.0 wt. % Al; optionally at most 0.2 wt. % of one or more additional elements, unavoidable impurities; the remainder being zinc. Preferably the alloying element contents in the zinc-alloy coating layer shall be 1.0-2.0% Mg and 1.0-3.0% Al, optionally at most 0.2% of one or more additional elements, unavoidable impurities and the remainder being zinc. In an even more preferred embodiment the zinc alloy coating comprises at most 1.6% Mg and between 1.6 and 2.5% Al, optionally at most 0.2% of one or more additional elements, unavoidable impurities and the remainder being zinc.

[0088] In another embodiment the steel strip, sheet or blank is provided with a (commercially pure) aluminium layer or an aluminium alloy layer. A typical metal bath for a hot dip coating such an aluminium layer comprises of

aluminium alloyed with silicon e.g. aluminium alloyed with 8 to 11 wt. % of silicon and at most 4 wt. % of iron, optionally at most 0.2 wt. % of one or more additional elements such as calcium, unavoidable impurities, the remainder being aluminium. Silicon is present in order to prevent the formation of a thick iron-intermetallic layer which reduces adherence and formability. Iron is preferably present in amounts between 1 and 4 wt. %, more preferably at least 2 wt. %.

[0089] The annealed strip can also be electrogalvanized with a zinc or zinc alloy coating instead of hot-dip galvanising. Optionally, the annealed strip may be electroplated with nickel, cobalt or chrome and combinations thereof.

[0090] The coated steel strip may optionally be subjected to temper rolling or tension levelling with a reduction of the coated steel strips thickness of at most 0.70%. Preferably the reduction is at most 0.45%. Preferably the reduction is at least 0.05%. The maximum of the reduction is governed by the need to secure a sufficiently high value of n_{4_6}. A minimum reduction may be used to improve the surface texture of the coated strip and/or to suppress the yield point elongation (YPE) and/or to increase the yield strength (Rp) of the material.

Examples

[0091] In table 2 the composition of the comparative (C) and inventive (I) steels is shown. Table 3 to 5 provide the process settings for the experiments. It is noted that table 4 relates to industrial scale trials and table 5 to laboratory scale trials.

TABLE 2

Comparative (C) and Inventive (I) examples (in m %, B and N in ppm).															
Alloy		C	Mn	Mn/C	P	Si	S	Al	zo	Si + Al	B	N	Ca	Cr	*
C	a	92	1555	16.9	14	69	1	606		675	21	21			LT
I	b	95	1558	16.4	12	397	0	599		996	21	27			LT
C	c	80	1763	22.0	12	300	4	508		808	22	20		204	LT
C	d	50	2250	45.0	10	300	0	250		550	22	28			LT
I	e	105	1736	16.5	12	309	0	511		820	22	28			LT
I	f	102	1726	16.9	12	413	0	577		990	23	31			LT
I	g	99	1661	16.8	11	308	0	518		826	21	19			LT
I	h	104	1633	15.7	13	406	13	541		947	22	26			LT
I	i	104	1642	15.8	13	306	5	83		389	20	28			LT
I	j	99	1639	16.6	13	306	0	297		603	19	22			LT
I	k	117	1501	12.8	0	595	0	570		1165	15	13			LT
I	l	121	1514	12.5	0	606	0	302		908	16	12			LT
I	1	105	1704	16.2	10	317	1	487		804	21	25	8		IT
I	2	106	1708	16.1	8	299	0	491		790	21	29	5		IT
I	3	110	1691	15.4	10	404	0	486		890	20	26	9		IT
I	4	121	1491	12.3	11	605	1	284		889	20	55	5		IT
I	5	124	1529	12.3	8	609	0	329		938	20	29	12	101	IT

* LT: lab-trials, IT: Industrial trials

TABLE 3a-continued

Hot rolling and cold rolling process setting of the lab cast alloys.					
Alloy	Process	T finish	Cooling rate (° C./s)	Tcoil	Cold rolling reduction (%)
d	7	>900	~30	570	75
e	8	>900	~30	570	75
f	9	>900	~30	570	75
f	10	>900	~30	570	75
g	11	>900	~30	550	73
h	12	>900	~30	550	73
i	13	>900	~30	570	73
j	14	>900	~30	570	73
k	15	>900	~30	610	70
l	16	>900	~30	600	70

TABLE 3b

Hot rolling and cold rolling process setting of the line-trialed alloys					
Alloy	Process	T finish	Cooling rate (° C./s)	Tcoil	Cold rolling reduction (%)
2	p-a	>890	>90	530	55
1	p-b	>890	>90	530	55
2	p-c	>890	>90	570	55
1	p-d	>890	>90	570	55
1	p-e	>890	>90	630	55
2	p-f	>890	>90	630	54
2	p-g	>890	>90	530	54
2	p-h	>890	>90	560	54
3	p-i	>890	>90	540	54

TABLE 3a

Hot rolling and cold rolling process setting of the lab cast alloys.					
Alloy	Process	T finish	Cooling rate (° C./s)	Tcoil	Cold rolling reduction (%)
a	1	>900	~30	550	60
a	2	>900	~30	550	75
b	3	>900	~30	570	75
b	4	>900	~30	570	60
c	5	>900	~30	570	70
d	6	>900	~30	570	70

TABLE 3b-continued

Hot rolling and cold rolling process setting of the line-trialed alloys					
Alloy	Process	T finish	Cooling rate (° C./s)	Tcoil	Cold rolling reduction (%)
3	p-j	>890	>90	570	54
4	p-k	>890	>90	540	54
4	p-l	>890	>90	570	54
5	p-m	>900	>90	5	50

TABLE 4

Process conditions of the line trialed alloys (% CR = 55%, % TR = 0.45%)												
Alloy	Process	Line	T1 (° C./s)	HR1 (° C./s)	T2 (° C.)	CR1 (° C./s)	T3 (° C.)	Austenite (%)	CR2 (° C./s)	T _{oa} (° C.)	t _{oa} (s)	Zn
2	p-a	2	718	23	796	1	786	—*	38	459	0	Y
1	p-b	S	780	4	860	4	718	25-30%	40	431	30	Y
2	p-c	2	720	23	797	1	787	—	41	459	0	Y
1	p-d	S	790	4	865	4	719	20-30%	21	400	30	Y
1	p-e	2	700	22	800	1	790	—	40	465	0	Y
2	p-f	2	724	23	802	1	792	—	41	465	0	Y
2	p-g	2	711	22	790	1	780	—	40	464	0	Y
2	p-h	2	709	22	796	1	786	—	40	461	0	Y
3	p-i	S	780	4	850	4	720	20-25%	20	420	30	Y
3	p-j	S	810	4	860	4	718	20-25%	21	400	30	Y
4	p-k	S	800	4	850	4	718	20-25%	20	400	30	Y
4	p-l	S	800	4	860	4	718	20-25%	22	400	30	Y
5	p-m	S	800	4	870	4	730	20-25%	20	460	30	Y

*—: not determined

TABLE 5

Process conditions of the lab cast alloys											
Alloy	Process	Line	T1 (° C.)	HR1 (° C./s)	T2 (° C.)	CR1 (° C./s)	T3 (° C.)	CR2 (° C./s)	T _{oa} (° C.)	t _{oa} (s)	Zn
a	1	2	710	23	790	2	780	39	N.A.	0	Y
a	2	3	680	21	825	4	650	38	400	0	Y
b	3	3	720	23	830	3	700	38	450	<60	Y
b	4	2	710	23	790	2	780	39	N.A.	0	Y
c	5	2	710	23	790	2	780	39	N.A.	0	Y
d	6	3	680	21	800	3	700	43	420	<60	Y
d	7	3	720	23	840	3	700	46	400	<60	Y
e	8	2	700	23	800	2	790	40	N.A.	0	Y
f	9	3	720	23	830	3	700	38	450	<60	Y
f	10	2	680	22	790	2	780	39	N.A.	0	Y
g	11	2	720	23	780	2	770	38	N.A.	0	Y
h	12	2	700	23	800	2	790	39	N.A.	0	Y
i	13	2	740	24	800	2	790	39	N.A.	0	Y
j	14	2	740	24	800	2	790	39	N.A.	0	Y
k	15	2	690	23	800	2	790	39	N.A.	0	Y
l	16	2	690	23	800	2	790	39	N.A.	0	Y

* Y: the samples were subjected to galvanization

[0092] Table 6 and 7 give the results of the industrial and laboratory scale trials. It should be noted that the results of the inventive examples all meet the requirements of the VDA-239 as presented in table 1. The values for Rp and Rm meet the requirements of VDA-239 of Rp between 330 and 430 MPa and Rm between 590 and 700 MPa

[0093] To determine the hole-expanding coefficient, which is a criterion for stretch-flangeability, three square samples (90×90 mm²) were cut out from each sheet, followed by punching a hole of 10 mm in diameter in the sample. Hole-expansion testing of the samples was done with upper burring. A conical punch of 60° was pushed up from below and the hole diameter df was measured when a through-thickness crack formed. The HEC was calculated using the formula below with d0=10 mm:

$$HEC = \frac{d_f - d_0}{d_0} \times 100\%$$

[0094] Stretch flangeability is evaluated on the basis of the maximum HEC-value and is deemed satisfactory when HEC>35%

BRIEF DESCRIPTION OF THE DRAWINGS

[0095] The invention will now be explained by means of the following, non-limiting figures.

[0096] FIG. 1 shows a schematic diagram of a continuous annealing and hot-dip coating line equipped with a pot to hold a molten metal (in this case zinc). The location indicated with the asterisk is the location where the optional galvannealing section is to be found. In most of these lines the hot-dip coating is optional, so the zinc pot can be bypassed if so desired.

[0097] FIG. 2 shows a schematic diagram of a galvannealing section located between the zinc pot and the turning roll at the top. The galvannealing section consists of a heating section, a holding section and a cooling section.

[0098] FIG. 3 shows a comparison of the results of a tensile test where the sample is taken under 0° (i.e. longitudinal), under 90° (transversal) and under 45° to the rolling direction of the strip. The Rm and Rp are relatively independent of the direction, but the total strain, uniform strain are not.

[0099] FIG. 4 shows a microstructure of alloy 4 (L) at ¼t. The microstructure consists of over 90% of ferrite (F)>90% (light grey matrix), 4-6% retained austenite (off-white), some martensite (dark grey/black), some bainite (grey) and the remainder (if any) is cementite.

[0100] FIGS. 5 to 8 show various (non-limitative) options for annealing the steels according to the invention:

[0101] FIG. 5: Continuous annealing with a maximum T at T2 and no plateau and hot-dip galvanizing with overaging at the hot-dip galvanizing temperature;

[0102] FIG. 6: CA & HDG with overaging below the temperature of the molten metal including a heating to the HDG-temperature;

[0103] FIG. 7: CA with a plateau at T2 & HDG with overaging below the temperature of the molten metal including a heating to the HDG-temperature;

[0104] FIG. 8: CA with a plateau at T2 & HDG without overaging;

[0105] FIG. 9 shows a schematic representation of the U-type coiling temperature profile of a hot-rolled strip prior to coiling.

TABLE 6

A80 tensile properties and HEC values - Results from line trials.											
Alloy	ID	Rp (MPa)	Rm (MPa)	Ag (%)	A80 (%)	A80 – Ag (%)	n (—)	R (—)	n4_6 (—)	d (mm)	HEC
2	I a	367	610	19.9	30.0	10.1	0.192	—	0.236	1.22	45
1	I b	381	595	19.7	30.3	10.6	0.191	—	0.207	1.21	—
2	I c	358	605	19.8	30.6	10.8	0.192	—	0.233	1.22	—
1	I d	366	584	20.8	32.5	11.7	0.199	—	0.22	1.21	—
1	I e	310	589	20.5	29.0	8.5	0.210	0.90	0.22	1.22	—
2	I f	374	608	19.1	32.7	13.6	0.187	0.90	0.22	1.22	—
2	I g	367	610	20.0	29.9	9.9	0.200	0.85	0.24	1.21	—
2	I h	351	592	19.0	32.2	13.2	0.210	0.85	0.24	1.21	—
3	I i	392	605	19.9	29.9	10.0	0.190	0.80	0.21	1.28	61*
3	I j	370	608	22.4	31.8	9.4	0.220	0.79	0.23	1.24	60*
4	I k	386	607	22.0	32.8	10.8	0.210	0.92	0.22	1.27	55*
4	I l	378	615	20.5	29.8	9.4	0.200	0.93	0.22	1.26	51*
5	I m	374	600	22.4	35.3	12.9	0.21	0.95	0.22	1.6	60**

*average from 11 HEC tests,
**average from 5 HEC tests

TABLE 7

JIS tensile properties and HEC values - Results from Lab trials												
Alloy	Process		Rp (MPa)	Rm (MPa)	Ag (%)	JIS (%)	JIS – Ag (%)	n (—)	Ae (%)	n4_6	Gauge (mm)	HEC
a	C	1	303	555	21.3	33.1	11.8	0.236	1.5	—	1.6	—
a	C	2	392	524	18.7	29.9	11.2	0.197	3.2	—	1.0	—
b	C	3	390	578	22.6	33.5	10.9	0.234	2.3	—	1.0	72
b	C	4	259	612	20.9	30.8	9.9	0.214	0.0	0.299	1.6	—
c	C	5	262	624	17.3	26.1	8.8	0.193	0.0	0.267	1.0	—
d	C	6	307	622	15.0	24.7	9.7	0.165	0.0	—	1.2	—
d	C	7	358	664	13.8	23.2	9.4	0.14	0.0	0.186	1.2	56
e	I	8	317	627	21.5	31.8	10.3	0.232	0.8	0.303	1.0	—
f	I	9	382	618	22.2	31.6	9.4	0.241	1.9	—	1.0	44
f	I	10	328	615	23.0	32.8	9.8	0.245	1.2	—	1.0	—
g	I	11	321	599	22.0	31.2	9.2	0.236	1.0	—	1.1	—
h	I	12	325	608	22.3	32.4	10.1	0.238	1.2	—	1.1	—
i	I	13	316	599	20.2	29.7	9.5	0.212	0.0	—	1.1	—
j	I	14	340	622	20.0	30.1	10.1	0.224	0.7	—	1.1	—
k	I	15	341	606	23.3	33.0	9.7	0.259	1.4	0.305	1.2	60
l	I	16	322	613	24.5	33.9	9.4	0.26	0.9	—	1.2	—

1. A dual phase steel strip consisting of, in wt. %:

C:	0.090-0.140	B:	0.0010-0.0050
Mn:	1.200-1.900	S:	at most 0.050
Si:	0.200-0.800	P:	at most 0.050
Al:	0.200-0.800	N:	at most 0.015

and optionally one or more of the elements selected from:

Cr:	at most 0.500	Nb:	at most 0.060
Mo:	at most 0.400	Ti:	at most 0.060
V:	at most 0.200	Ca:	at most 0.0050

the remainder being iron and unavoidable impurities, wherein the steel has a tensile strength Rm of 580-720 MPa and a yield strength Rp of 310-430 MPa, and wherein the steel strip is optionally i) a hot-dip galvanized or ii) electrogalvanized or iii) electroplated.

2. The dual phase steel strip according to claim 1, consisting of, in wt. %:

C:	0.090-0.130	B:	0.0010-0.0040
Mn:	1.400-1.800	S:	at most 0.004
Si:	0.250-0.750	P:	at most 0.040
Al:	0.200-0.750	N:	at most 0.012

and optionally one or more of the elements selected from:

Cr:	at most 0.400	Nb:	at most 0.050
Mo:	at most 0.300	Ti:	at most 0.050
V:	at most 0.150	Ca:	at most 0.0050

the remainder being iron and unavoidable impurities.

3. The dual phase steel strip according to claim 1, consisting of, in wt. %:

C:	0.090-0.130	B:	0.0010-0.0030
Mn:	1.400-1.750	S:	at most 0.003
Si:	0.250-0.700	P:	at most 0.015
Al:	0.200-0.650	N:	at most 0.006

and optionally one or more of the elements selected from:

Cr:	at most 0.120	Nb:	at most 0.010
Mo:	at most 0.015	Ti:	at most 0.010
V:	at most 0.010	Ca:	at most 0.0050

the remainder being iron and unavoidable impurities.

4. The dual phase steel strip according to claim 1, consisting of, in wt. %:

C:	0.110-0.130	B:	0.0012-0.0028
Mn:	1.400-1.750	S:	at most 0.003
Si:	0.500-0.700	P:	at most 0.015
Al:	0.200-0.400	N:	at most 0.006

and optionally one or more of the elements selected from:

Cr:	at most 0.040	Nb:	at most 0.010
Mo:	at most 0.015	Ti:	at most 0.010
V:	at most 0.010	Ca:	at most 0.0050

the remainder being iron and unavoidable impurities.

5. The dual phase steel strip according to claim 1, consisting of, in wt. %:

C:	0.110-0.130	B:	0.0012-0.0028
Mn:	1.400-1.750	S:	at most 0.003
Si:	0.500-0.700	P:	at most 0.015
Al:	0.200-0.400	N:	at most 0.006
Cr:	0.020-0.150		

and optionally one or more of the elements selected from:

Mo:	at most 0.015	Ti:	at most 0.010
V:	at most 0.010	Ca:	at most 0.0050
Nb:	at most 0.010		

the remainder being iron and unavoidable impurities.

6. The dual phase steel strip according to claim 1, consisting of, in wt. %:

C:	0.090-0.130	B:	0.0010-0.0035
Mn:	1.300-1.800	S:	at most 0.003
Si:	0.250-0.750	P:	at most 0.030
Al:	0.200-0.650	N:	at most 0.010

and optionally one or more of the elements selected from:

Cr:	at most 0.150	Nb:	at most 0.040
Mo:	at most 0.200	Ti:	at most 0.040
V:	at most 0.100	Ca:	at most 0.0050

the remainder being iron and unavoidable impurities.

7. The dual phase steel strip according to claim 1, wherein one or more or all of Cr, Mo, V, Nb and Ti is present only as an impurity which means that, if applicable, Cr is below 0.020 wt. %, Ti is below 0.010 wt. %, V is below 0.010 wt. %, Mo is below 0.015 wt. %, Nb is below 0.010 wt. %.

8. The dual phase steel strip according to claim 1, wherein the steel has a tensile strength Rm of 590-720 MPa and/or a yield strength Rp of 320-430 MPa.

9. The dual phase steel strip according to claim 1, wherein the steel has a uniform elongation Ag of at least 18%.

10. The dual phase steel strip according to claim 1, wherein the steel has a hole expansion capacity HEC of at least 35%.

11. The dual phase steel strip according to claim 1, wherein the steel has an average strain hardening exponent or n-value of at least 0.16.

12. The dual phase steel strip according to claim 1, which is provided with the metallic coating by hot-dip coating immediately subsequent to continuous annealing or by electrodeposition of the strip after cooling the continuously annealed strip to ambient temperature.

13. A method for producing a coated and optionally temper-rolled dual phase steel strip with a tensile strength Rm of 580-720 MPa and a yield strength Rp of 310-430 MPa according to claim 1, comprising the following steps:
providing a hot-rolled steel strip by hot rolling a continuously cast slab with a composition consisting of, in wt. %:

C:	0.090-0.140	B:	0.0010-0.0050
Mn:	1.200-1.900	S:	at most 0.050
Si:	0.200-0.800	P:	at most 0.050
Al:	0.200-0.800	N:	at most 0.015

and optionally one or more of the elements selected from:

Cr:	at most 0.500	Nb:	at most 0.060
Mo:	at most 0.400	Ti:	at most 0.060
V:	at most 0.200	Ca:	at most 0.0050

the remainder being iron and unavoidable impurities, to a hot-rolled strip having a thickness of 2.0-4.5 mm, wherein finish-rolling is performed while the strip has an austenitic microstructure;

- Cooling the hot-rolled strip after finish-rolling with a cooling rate of at least 30° C./s;
- Coiling the cooled strip at a coiling temperature CT in the range of 500 to 650° C. wherein the amount of austenite is more than 60% and the coiled strip is subsequently allowed to cool to ambient temperatures;
- Uncoiling the coiled hot-rolled strip, followed by pickling and cold-rolling with a reduction of 40-80%;
- Continuous annealing of the cold-rolled strip by:
- i. heating the strip to a temperature T1 in the range of 550-710° C. with a heating rate HR1 of at least 3° C./s to allow the recrystallisation of the cold-rolled strip to proceed for at least 60% prior to the start of the austenite formation;
 - ii. further heating of the strip with an average heating rate HR2 to a temperature T2 in the range of (Ac1+50° C.) to (Ac3-30° C.) to form a partially austenitic microstructure;
 - iii. subsequently either:
 - a. keeping the strip at T2 for a time period t2 of up to 300 s, followed by slow cooling of the strip with a cooling rate CR1 in the range of 0.5-12° C./s to a temperature T3 in the range of 600-790° C., preferably 620-790° C. so that 16-30% austenite is present in the slow cooled strip, or
 - b. immediately slowly cooling the strip from T2 with a cooling rate CR1 in the range of 0.5-12° C./s to a temperature T3 in the range of 600-790° C. so that 16-30% austenite is present in the slow cooled strip;

- iv. rapidly cooling the slowly cooled strip comprising 16-30% austenite with an average cooling rate CR2 in the range of 5-70° C./s either:
 - a. to a temperature T4 in the range of 330-470° C.; or
 - b. to an overaging temperature T_oa for a time t_oa in the range of 5 to 100 s, wherein T_oa is between 390 and 465° C.;
 - v. cooling the steel strip at a cooling rate CR3 of at least 4° C./s to a temperature below 300° C.;
- wherein the strip is a). provided with a metallic coating by hot-dip coating between step iv. and v., optionally followed by galvannealing the hot-dip coated strip, or b). with a metallic coating by electrodeposition after step v.;
- Optionally temper rolling or tension levelling of the coated steel strip with a reduction of at most 0.70%;
- Coiling the coated steel strip or cutting the coated steel strip into sheets or blanks;
- Optionally shaping the coated steel strip, sheet or blank via cold-forming operations like stamping, bending, deep drawing or via warm press forming or hot press forming.
14. The method according to claim 13, wherein one or more of the following process parameters apply:
Coiling temperature is below 630° C.;
Coiling temperature is above 520° C.;
Cold rolling reduction is at least 44%;
T3 is between 400 and 470° C.
15. The method according to claim 13 to produce a coated steel strip, sheet or blank having one or more of
a uniform elongation Ag of at least 18%;
a total elongation A80 or JIS of at least 26%;
a hole expansion capacity HEC of at least 35%;
a total elongation minus uniform elongation of at least 7.0%.
16. A car or truck component, such as an automotive chassis or safety component, a B-pillar, a reinforcement (crash) part, a front crash beam, a seat member, a bumper part, a door part, a component of the body in white, a component of the frame or the sub-frame, an electric battery holder or container part, said component having been produced from the steel according to claim 1.
17. The dual phase steel strip according to claim 1, wherein the steel has a uniform elongation Ag of at least 30%.
18. The dual phase steel strip according to claim 1, wherein the steel has an average strain hardening exponent or n-value of at least 0.24.
19. The method of claim 13, wherein the temperature T3 is in the range of 620-790° C.
20. The method according to claim 13, wherein one or more of the following process parameters apply:
Coiling temperature is below 600° C.;
Coiling temperature is above 530° C.;
Cold rolling reduction is at least 67%;
T3 is between 400 and 470° C.

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