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(54) **HOT ROLLED STEEL SHEET WITH ULTRA-HIGH STRENGTH AND IMPROVED FORMABILITY AND METHOD FOR PRODUCING THE SAME**

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(57) **ABSTRACT**

A hot rolled high strength steel sheet with improved formability, including of the following elements in weight %:0.13-0.35 C; 0.80-3.50 Mn; 0.01-2.50 Si; 0.02-1.00 Al; wherein the sum of Si and Al is 0.52-2.50; and optionally one or more selected from: 0.0002-0.0030 Ca; 0.0004-0.0100 REM; 0.10-1.00 Cu; 0.10-1.50 Cr; 0.10-1.00 Ni; 0.05-0.50 Mo; 0.0005-0.0050 B; 0.010-0.100 Nb; 0.010-0.100 Ti; 0.020-0.200 V; wherein the sum of Nb+Ti+V is 0-0.250 and wherein the sum of Cr+Ni+Cu+Mo is 0-1.50, and at most 0.100 P; at most 0.050 S; at most 0.0100 N; the balance inevitable impurities and Fe, wherein the hot rolled high strength steel sheet has a complex phase microstructure including in vol. %:65% or more of a matrix of a mixture of carbide free bainitic ferrite (BF) and/or tempered martensite (TM), and 5 to 25% of retained austenite (RA).

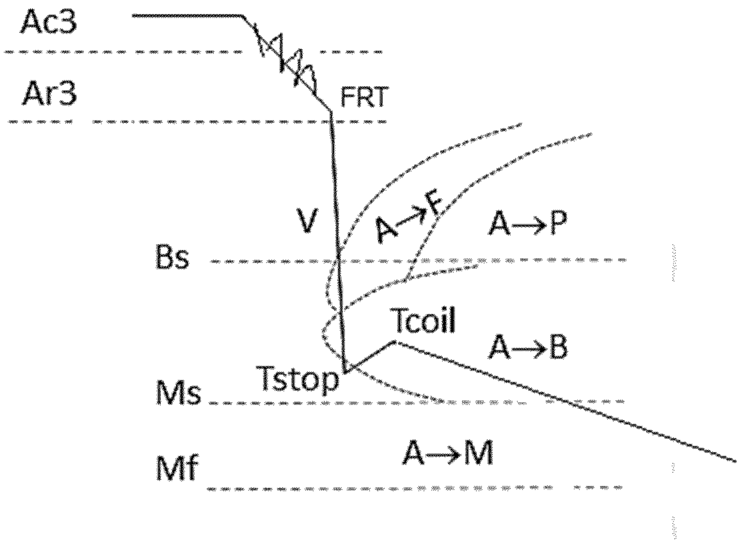

Figure 1

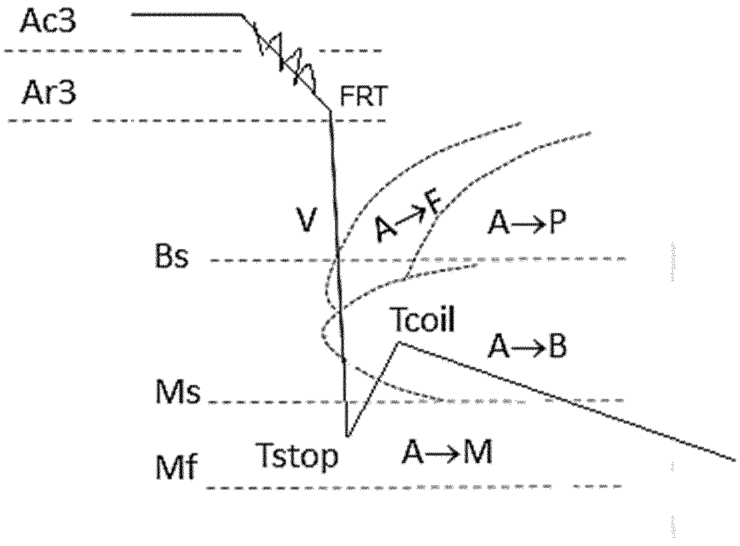

Figure 2

HOT ROLLED STEEL SHEET WITH ULTRA-HIGH STRENGTH AND IMPROVED FORMABILITY AND METHOD FOR PRODUCING THE SAME

[0001] The invention relates to a hot rolled high strength steel sheet having improved formability and a method for producing such a steel sheet.

[0002] In the automotive industry there exists a constant wish for improved steel types having a high strength and a good formability, e.g. for use as chassis. For such steel types, special compositions with increased levels of one or more micro-alloying elements may be needed. In order to make automobiles or other equipment lighter, a high strength of the steel sheets is required. However, an increase in strength of a steel sheet is normally accompanied by a decrease in its ductility or formability. Therefore, it is difficult to produce a steel sheet having both good formability and high strength.

[0003] Advanced high strength steels (AHSS) such as dual phase (DP) steels and transformation induced plasticity (TRIP) steels are currently typical of high-ductility, high-strength steels used for automobiles. In DP steels, the presence of martensite within a ferrite matrix enables to obtain a tensile strength greater than 450 MPa, combined with good cold formability.

[0004] TRIP (Transformation Induced Plasticity) steels have been developed to provide even higher tensile strength and elongation. These properties are associated with the structure of such steels, which consists of a ferritic matrix containing bainite/martensite and retained austenite. The retained austenite is generally stabilized by an addition of silicon or aluminium to the steel composition, retarding the precipitation of carbides. The presence of retained austenite brings high ductility in an undeformed steel sheet. The retained austenite is progressively transformed to martensite, resulting in substantial hardening and delaying the occurrence of necking in the uniaxial tensile deformation, thus providing an excellent elongation. However, due to the difference of the deformation capabilities between the ferrite, bainite or martensite and the retained austenite, the steels are generally inferior in stretch flangeability, and therefore, TRIP steels are limitedly used in automobile parts which do not require a high formability.

[0005] U.S. Pat. No. 4,472,208 for example discloses a hot-rolled high tensile titanium steel plate and production thereof. The steel plate has improved toughness and cold formability and is made of a killed steel which consists essentially of (in wt %): C: 0.05-0.20, Si: not more than 1.2, Mn: 0.5-2.0, Ti: 0.04-0.20, P: not more than 0.025, S: not more than 0.015, sol. Al: 0.005-0.15, O: not more than 0.0080, N: not more than 0.0080, B: 0-0.0030, Cr: 0-1.0, Ca: 0-0.010, the balance being Fe and incidental impurities, the Ti content comprising not less than 0.02 of incoherently precipitated Ti and not more than 0.015 of coherently precipitated Ti, and said killed steel containing 20 to 90% by volume of a bainitic structure and not less than 10% by volume of a ferritic structure.

[0006] US20130061989 discloses a high strength hot-rolled steel sheet with a tensile strength of not less than 780 MPa which exhibits excellent stretch flangeability and excellent fatigue resistance. The steel has a composition containing (in wt %) C at 0.05 to 0.15, Si at 0.2 to 1.2, Mn at 1.0 to 2.0, P at not more than 0.04, S at not more than 0.005, Ti at 0.05 to 0.15, Al at 0.005 to 0.10 and N at not more than 0.007.

[0007] EP0997548 discloses a hot rolled high strength steel sheet having a satisfactory strength-elongation balance and an excellent formability, which contains (in wt %): C: 0.05-0.15, Si: 0.5-2.0, Mn: 0.5-2.0, P: 0.05 or less, S: 0.010 or less, Al: 0.005-0.10, and the balance of Fe and inevitable impurities, and in which the metal structure mainly comprises three phases of ferrite, retained austenite and bainite.

[0008] EP1790737 discloses a high strength steel sheet having a strength not lower than 780 MPa and exhibiting a high yield ratio, an excellent balance between the strength and the uniform elongation, and a good plating property in addition to a good balance between the strength and the stretch flangeability. The invention provides a high strength steel, comprising 0.05 to 0.25% of C, less than 0.5% of Si, 0.5 to 3.0% of Mn, not more than 0.06% of P, not more than 0.01% of S, 0.50 to 3.0% of Sol. Al, not more than 0.02% of N, 0.1 to 0.8% of Mo, 0.02 to 0.40% of Ti, and the balance of iron and unavoidable impurities, wherein the steel has a structure formed of at least three phases including a bainite phase, and a retained austenite phase in addition to a ferrite phase having a composite carbide containing Ti and Mo dispersed and precipitated therein, wherein the total volume of the ferrite phase and the bainite phase is not smaller than 80%, the volume of the bainite phase is 5% to 60%, and the volume of the retained austenite phase is 3 to 20%.

[0009] U.S. Pat. No. 5,470,529 discloses a high tensile strength, hot or cold rolled steel sheet having improved ductility and hole expandability consisting essentially, on a weight basis, of: C: 0.05-0.3%, Si: 2.5% or less, Mn: 0.05-4%, Al: greater than 0.10% and not greater than 2.0% wherein 0.5% Si+Al 3.0%, optionally one or more of Cu, Ni, Cr, Ca, Zr, rare earth metals (REM), Nb, Ti, and V, and a balance of Fe and inevitable impurities with N being limited to 0.01% or less. The steel sheet has a structure comprising at least 5% by volume of retained austenite in ferrite or in ferrite and bainite. A hot rolled steel sheet is produced by hot rolling with a finish rolling end temperature in the range of 780-840° C., cooling to a coiling temperature in the range of 300-450° C. either by rapid cooling to the coiling temperature at a rate of 10-50° C./s or by initial rapid cooling to a temperature range of 600-700° C., then air-cooling for 2-10 seconds, and final rapid cooling to the coiling temperature. A cold rolled steel sheet is produced by hot rolling, cooling to a coiling temperature in the range of 300-720° C., descaling, cold rolling with a reduction of 30-80%, and annealing. Annealing is performed by heating between the Ac1 point and the Ac3 point and cooling such that the temperature is either kept for at least 30 seconds in the range of 550 to 350° C. or slowly decreased at a rate of 400° C./min or less in that temperature range.

[0010] US20090107588 relates to a hot-rolled steel sheet having a tensile strength of greater than 1200 MPa, an Re/Rm ratio of less than 0.75 and an elongation at break of greater than 10%, the composition of which contains, the contents being expressed by weight: 0.10% C 0.25%; 1% Mn 3%; Al 0.015%; Si 1.985%; Mo 0.30%; Cr 1.5%; S 0.015%; P 0.1%; Co 1.5%; B 0.005%; it being understood that 1% Si+Al 2%; Cr+(3×Mo) 0.3%, the balance of the composition consisting of iron and inevitable impurities resulting from the smelting, the microstructure of the steel consisting of at least 75% bainite, retained austenite in an amount equal to or greater than 5% and martensite in an amount equal to or greater than 2%.

[0011] US20080202645 relates to a steel sheet excellent in both a balance between strength and elongation and a balance between strength and hole expandability, in other words, a multi-phase steel sheet having an excellent balance between strength and hole expandability. The invention is a multi-phase steel sheet excellent in hole expandability characterized in that: the steel sheet contains, as chemical components in mass, C: 0.03 to 0.15%, P: not more than 0.010%, S: not more than 0.003%, and either one or both of Si and Al in a total amount of 0.5 to 4%, and one or more of Mn, Ni, Cr, Mo and Cu in a total amount of 0.5 to 4%, with the balance consisting of Fe and unavoidable impurities; the microstructure at a section of the steel sheet is composed of either one or both of retained austenite and martensite which account(s) for 3 to 30% in total in area percentage and the balance consisting of either one or both of ferrite and bainite; the maximum length of the crystal grains in the microstructure is not more than 10 microns; and the number of inclusions 20 microns or larger in size at a section of the steel sheet is not more than 0.3 piece per square millimeter.

[0012] EP2714947 relates to a bainite steel consisting of the following elements in weight %: C: 0.25-0.55, Si: 0.5-1.8, Mn: 0.8-3.8, Cr: 0.2-2.0, Ti: 0.0-0.1, Cu: 0.0-1.2, V: 0.0-0.5, Nb: 0.0-0.06, Al: 0.0-2.75, N: <0.004, P: <0.025, S: <0.025 and a method for manufacturing a bainite steel strip that comprises the step of cooling the coiled strip of such composition to ambient temperature, during which the bainite transformation takes place.

[0013] One of the major problems of hot rolled high strength steel sheets, either having DP microstructure or utilizing the TRIP phenomenon of retained austenite, is the low formability, in particular flangeability and/or bendability characterised by low hole expansion ratio and low bending angle respectively despite their high elongation in tensile test. As a result, they cannot be successfully used in fabrication by press forming.

[0014] It is believed that deterioration in press formability of the advanced high strength steels in general is attributable to the fact that the local ductility in the press-formed area is greatly deteriorated at a late stage of deformation in press forming, because of the large hardness difference between ferrite and martensite. In case of hole expansion, the high hardness of the martensite phase causes the formation of minute cracks around the initial hole, which are extended or propagated in the subsequent hole expansion stage, thereby deteriorating the hole expandability. As a result, the hole expandability or bendability of these steel sheets is insufficient for certain applications, in particular if forming processes such as press forming are part of the entire production process.

[0015] An object of the invention is to provide a hot rolled high strength steel sheet having improved formability, suitable for structural or high strength components to be shaped by press forming or flange forming in automobiles, industrial machinery and equipment. A further object of the invention is to provide a hot rolled high strength steel sheet having improved formability including increased ductility, improved hole expandability and bendability.

[0016] It is another object of the invention to provide such a hot rolled high strength steel sheet without the need of various expensive alloying elements, thereby lowering the cost and simplifying the thermomechanical manufacturing schemes.

[0017] A further object of the invention is to provide a method of manufacturing a hot rolled high strength steel sheet according to the objects described hereinabove.

[0018] According to a first aspect of the invention, one or more of these objects is reached with a hot rolled high strength steel sheet having a tensile strength of at least 950 MPa consisting of the following elements in weight %:

[0019] 0.13-0.35 C;

[0020] 0.80-3.50 Mn;

[0021] 0.01-2.50 Si;

[0022] 0.02-1.00 Al;

[0023] wherein the sum of Si and Al is 0.52-2.50; and

[0024] optionally one or more selected from:

[0025] 0.0002-0.0030 Ca;

[0026] 0.0004-0.0100 REM;

[0027] 0.10-1.00 Cu;

[0028] 0.10-1.50 Cr;

[0029] 0.10-1.00 Ni;

[0030] 0.05-0.50 Mo;

[0031] 0.0005-0.0050 B;

[0032] 0.010-0.100 Nb;

[0033] 0.010-0.100 Ti;

[0034] 0.020-0.200 V;

[0035] wherein the sum of Nb+Ti+V is 0-0.250 and

[0036] wherein the sum of Cr+Ni+Cu+Mo is 0-1.50, and

[0037] at most 0.100 P;

[0038] at most 0.050 S;

[0039] at most 0.0100 N;

[0040] the balance consisting of inevitable impurities and Fe,

wherein the hot rolled high strength steel sheet has a complex phase microstructure comprising in vol. %:

[0041] 65% or more of a matrix of a mixture of carbide free bainitic ferrite (BF) and/or tempered martensite (TM), and

[0042] 5 to 25% of retained austenite (RA),

and wherein the microstructure comprises less than 3% carbides

[0043] The invention provides a hot rolled high strength steel sheet having an excellent improved formability including increased ductility, improved hole expandability and bendability. This excellent combination is derived from a good balance in the strength and ductility of the carbide-free bainitic ferrite/tempered martensite matrix and the TRIP effect of the retained austenite. Bainitic ferrite refers to plate-shaped ferrite having higher density of dislocations (which may have a lath-shaped substructure or a granular substructure), and is different from pro-eutectoid ferrite that has a low density of dislocations. Tempered martensite has a similar substructure as bainitic ferrite albeit with a finer size substructure. The matrix of bainitic ferrite and/or tempered martensite is carbide-free and is different from the conventional bainite that contains carbides. Without wishing to be bound by theory it is believed that the matrix of bainitic ferrite and/or tempered martensite provides high strength due to a high dislocation density and a supersaturated carbon content. The matrix also provides high global elongation and stretch flangeability since it is carbide-free and the fine retained austenite grains can be present on the boundary of lath-shaped bainitic ferrite and/or tempered martensite. It is controlled so that the bainitic ferrite and/or tempered martensite matrix occupies at least 65% in terms of a volume fraction to the entire structure. Retained austenite enhances ductility partly through the TRIP effect, which manifests itself in an increase in uniform elongation.

[0044] In addition, the hardness difference between the bainitic ferrite and/or tempered martensite matrix and retained austenite is minimal, thereby improving the local ductility at a late stage of deformation in press forming and preventing related issues such as premature edge cracking during stretch flanging, resulting in improved hole expandability and bendability. The local ductility is improved even further by ensuring the absence of carbides in the bainitic ferrite and/or tempered martensite matrix.

[0045] The steel according to the invention can be obtained without the need of various expensive alloying elements, only requiring the presence of carbon, manganese, silicon and aluminium. This is beneficial for cost reasons as well as for simplifying the process, as thermomechanical manufacturing systems are not necessary.

[0046] The role of the alloying elements for the present invention is as follows. All amounts of elements in compositions are given in weight %, unless specified otherwise.

[0047] Compulsory Elements

[0048] Carbon (C) is controlled to a value between 0.13 and 0.35. Carbon serves to strengthen the steel and to stabilize the austenite. During coil cooling after hot rolling, a bainite transformation takes place and bainitic ferrite laths are initially formed within a still predominantly austenitic matrix. C becomes concentrated in an untransformed austenite phase as bainitic transformation proceeds in the course of coil cooling (austempering) after hot rolling, thereby stabilizing the austenite phase. The retained austenite provides TRIP effect. In order to obtain sufficient tensile strength and satisfactory stability of the retained austenite, C should be 0.13 or more, preferably 0.15 or more, more preferably 0.17 or more, most preferably 0.20 or more. Below 0.13, a sufficient tensile strength cannot be obtained and the stability of the retained austenite is unsatisfactory. Above 0.35, the weldability is significantly reduced. Therefore, the C content is limited to 0.35 less, preferably to 0.32 or less, more preferably 0.30 or less, most preferably 0.26 or less.

[0049] Manganese (Mn) is controlled to a value between 0.80 and 3.50. Mn promotes the formation of the y-phase and stabilizes the austenite by lowering the transformation temperature Ar3. The addition of manganese also contributes to effective solid-solution hardening and to achieving a higher tensile strength. In order to obtain sufficient hardenability, Mn should be 0.80 or more, preferably 1.20 or more, more preferably 1.40 or more, most preferably 1.60 or more. A Mn content in excess of 3.50 will result in a decrease of the ductility and hampers the formation of acicular ferrite in a sufficient amount during cooling after hot rolling or annealing, thereby affecting the stability of the austenite phase. The risk of forming deleterious banded microstructures is also increased. Therefore, the Mn content is limited to 3.50 or less, preferably to 3.00 or less, more preferably to 2.60 or less, most preferably to 2.40 or less.

[0050] Silicon (Si) is controlled to a value between 0.01 and 2.50 and has the main function to prevent carbon from precipitating in the form of iron carbides, most commonly cementite and to suppress decomposition of retained austenite. Si is a ferrite stabilizer and is known to be quite effective for the retention of austenite during cooling after hot rolling. A minimal amount of 0.50 Si is needed to effectively suppress the carbide formation, when no substantial amount of Al is present. When a substantial amount of Al is added, the lower limit of the Si content is not critical as Al also suppresses the formation of carbides, and a minimal Si content of 0.01 is present. Addition of Si in excess of 2.50 is undesirable, as this results in the formation of coarse bainite grains or hard martensite, thereby deteriorating the hole expandability and also causes Si-rich scales inherent in Si-containing steels to form in an appreciable amount, thereby deteriorating the surface appearance and surface processability by alloyed galvanizing. Therefore, the Si content is limited to 2.50 or less and preferably to 2.00 or less, more preferably to 1.50 or less.

[0051] Aluminium (Al) is controlled to a value between 0.02 to 1.00. A primary function of Al is deoxidising the liquid steel before casting, therefore at least 0.02 is needed for this function. In addition, Al is a ferrite stabilizer like Si and assists in retention of austenite during cooling after hot rolling by accelerating the formation of acicular ferrite, facilitating the enrichment of C in the untransformed austenite phase, and retarding the precipitation of carbides.

[0052] When a substantial amount of Si is added, the lower limit of the Al content is not critical as Al also suppresses the formation of carbides. Addition of Al in excess of 1.00 increases the Ac3 temperature of the steels significantly, which may introduce too much ferrite in the microstructure, thereby adversely affecting the strength and ductility. Therefore, the Al content is limited to 1.00 or less, preferably 0.80 or less, more preferably 0.60 or less.

[0053] As the effects of aluminium and of silicon on the retardation of carbides and on the stabilization of austenite are very similar, they can be replaced by each other. The total amount of the Si and the Al content (Si+Al) is controlled at a value between 0.52 and 2.50, thereby achieving the desired microstructures to be formed while maintaining satisfactory usage properties. Preferably the total amount of Si+Al is controlled to a value between 0.80 and 2.00, more preferably between 1.00 and 1.60. The upper value should not exceed 2.50 because the beneficial effects of Si and Al saturate and the steel processing becomes difficult. The lower value should be at least 0.52 to effectively suppress carbide formation and deoxidize the liquid steel before casting.

[0054] Although Al has a similar function as Si to prevent the formation of carbides and to stabilize the retained austenite, it is not as effective as Si and has no significant effect on strengthening. Therefore, the ratio of Si:Al is preferably higher than 2:1, more preferably 3:1 when they are used together.

[0055] Optional Elements

[0056] Furthermore, the composition according to the invention may comprise one or more of the following elements. These elements can be present as alloying elements or residual elements.

[0057] Residual elements (aka inevitable impurities, such as Cu, Ni, As, Pb, Sn, Mo, Cr, etc.) are defined as elements which are not added on purpose to steel and which cannot be removed by simple metallurgical processes. Some of the elements can be present as residual element, like chromium, but chromium can also be added to the steel during steelmaking, in which case it would no longer be a residual element or inevitable impurity, but an alloying element. In many cases it cannot be determined in the finished steel product whether (e.g.) chromium was added as an alloying element, or whether it was present as a residual element.

Residual elements enter steel from impurities in ore, coke, flux and scrap; from these, scrap is considered to be the main source of residuals. Consequently, the level of residual elements in the Electric Arc Furnace process route (100% scrap based) is significantly higher than in the Basic Oxygen Steelmaking (BOS) process route. Therefore, the steel according to the invention is preferably produced in a BOS-process route.

[0058] The composition according to the invention may contain one or two elements selected from Ca and a rare earth metal (REM), in an amount consistent with a treatment for MnS inclusion control. Examples of the rare earth element include scandium, yttrium, and lanthanide. Ca and/or a rare earth metal (REM) is added to the steel composition to modify the shape of the sulfides. Ca is controlled between 0.0002 and 0.0030 and REM is controlled to a value between 0.0004 and 0.0100. Ca and/or REM combines with sulfur and oxygen, thus creating oxysulfides that do not exert a detrimental effect on ductility, as in the case of elongated manganese sulfides which would form if no Ca or REM is present. This effect is saturated when Ca content is higher than 0.0030 or the REM content is higher than 0.0100. Preferably the amount of Ca, if present, is controlled to a value below 0.0030, more preferably below 0.0020. Preferably the amount of REM, if present, is controlled to a value below 0.0080, more preferably below 0.0050. The lower limit of Ca, if present, is controlled at 0.0002, preferably 0.0005. The lower limit of REM, if present, is controlled at 0.0004, preferably 0.0005.

[0059] Chromium (Cr) is an optional element for the present invention and can contribute to the prevention of pro-eutectoid ferrite formation and to the hardening and refinement of the bainitic microstructure. However, in combination with the other elements of the composition that also increase hardenability, a Cr addition is not necessary but can be present as a residual element.

[0060] For the hot rolled high strength steel sheet according to the invention, the allowable level of Cr as a residual element is 0.04. If it is present as an alloying element, the Cr content is controlled between 0.10 and 1.50. In order to obtain a hardening effect of the bainitic microstructure by Cr, a lower limit of 0.10 is needed, preferably 0.15 and more preferably 0.20. Cr should not exceed 1.50 because it will significantly slow down the bainite transformation kinetics and increases the cost of the steel. Preferably the level of Cr is 1.00 or less, more preferably 0.50 or less.

[0061] Molybdenum (Mo) is another optional element which, if present, is controlled between 0.05 and 0.50. The allowable level of Mo as a residual element is 0.02. Mo can retard the bainite transformation, may contribute to solid-solution hardening and can also refine the size of the bainite laths formed. In order to obtain these effects of Mo, a lower limit of 0.05 is needed, preferably 0.10, more preferably 0.15. However, in order to avoid excessive formation of hardening structures and to control costs, Mo should be 0.50 or less, preferably 0.30 or less, more preferably 0.20 or less.

[0062] Boron (B) is another optional element which, if present, is controlled between 0.0005 and 0.0050. An addition of boron increases the quench hardenability and contributes to the elimination of pro-eutectoid ferrite and also helps to increase the tensile strength levels, to obtain these effects of B, a lower limit of 0.0005 is needed, preferably 0.0007, more preferably 0.0010. However, when too much B is added, the effect is saturated. B should be controlled to be 0.0050 or less, preferably 0.0040 or less, more preferably 0.0030 or less, because when B is present in excess, the advantages of this element may be saturated.

[0063] Copper (Cu) is another optional element which if present, is controlled between 0.10 and 1.00. The allowable level of Cu as a residual element is 0.04. Cu serves to facilitate the removal of high-Si scales formed in the hot rolling stage which is mainly beneficial for subsequent cold rolled product. Cu can improve the corrosion resistance when the cold rolled steel sheet is used as such without surface treatment or improve the wettability by molten zinc or the alloy forming ability when it is subjected to galvanizing or alloyed galvanizing in a continuous galvanizing line. To obtain these effects, Cu should preferably be present in 0.10 or more, preferably 0.15 or more. However, an excess of Cu may result in surface defects such as hairline cracks during heating prior to hot rolling, which is due to the formation of a Cu-rich, low melting alloy phase along the austenite grain boundaries resulting in hot shortness. Therefore, the Cu content should be controlled to be 1.00 or less, preferably 0.80 or less, more preferably 0.50 or less, when Cu is added.

[0064] Nickel (Ni) is another optional element which if present, is controlled between 0.10 and 1.00. The allowable level of Ni as a residual element is 0.04. Ni serves to minimize the formation of the surface defects caused by Cu by increasing the melting point of the alloy phase. This effect of Ni is appreciable when the Ni content is >[Cu(%)/3].

[0065] In addition, the sum of Cr+Ni+Cu+Mo should not exceed 1.50, preferably 1.00 for cost reasons.

[0066] Niobium (Nb), titanium (Ti) and vanadium (V) all precipitate as a carbo-nitride in a ferrite matrix, thereby contributing to a further increase in tensile strength of the steel sheet and are also optional elements for the steel according to the invention. In order to obtain this effect the lower limit for any of these elements should be controlled at 0.010 or more, preferably 0.030 or more. The allowable levels of Nb, Ti and V as a residual element is 0.005 for each. The effect becomes saturated when the content exceeds 0.100 for Nb and Ti or 0.200 for V. Therefore, when these elements are added, the Nb content and Ti content are controlled between 0.010 and 0.100 and the V content between 0.020 and 0.200. Preferably, the upper limit is 0.050 or less for Nb and Ti and 0.100 or less for V, because when these are added excessively, too much carbo-nitride is precipitated which result in deteriorating the workability.

[0067] In addition the sum of Ti+Nb+V should not exceed 0.250, preferably 0.100 for workability and cost.

[0068] In another preferred embodiment of the invention, no additional Ti and/or Nb and/or V and/or Ni and/or Cu and/or B and/or Mo and/or Cr are added as alloying elements in order to reduce the cost of the final product while still obtaining a hot rolled high strength steel sheet having an excellent improved formability including increased ductility, improved hole expandability and bendability.

[0069] Specified Inevitable Impurities

[0070] Nitrogen (N) is present as an incidental impurity, and hence the steel according to the invention preferably has a minimized N content. The maximum acceptable N content is 0.0100 since a greater N content significantly increases the amount of Al consumed as AlN, thereby not only diminishing the above-described favorable effects of Al but also causing a prominent deterioration in ductility. Preferably, the N content is 0.0050 or less, more preferably 0.0040 or less.

[0071] Phosphorus (P) is another incidental impurity and adversely affects weldability and ductility. Preferably, P is limited to 0.100 or less. In order to assure uniform distribution of polygonal ferrite grains, preferably the P content is 0.050 or less, more preferably 0.020 or less.

[0072] Sulphur (S) is also an incidental impurity, which adversely affects ductility and press formability by the formation of sulfide inclusions. Preferably, S is limited to 0.050 or less. In order to further improve the press formability, it is more preferable that the S content be 0.010 or less, even more preferably 0.005 or less.

[0073] Phase Transformation Temperatures

[0074] The various phase transformation temperatures of steels are dependent on the alloy compositions and are defined as follows, as well known to a person skilled in the art.

[0075] Ac3: the temperature at which, during heating, transformation of the ferrite into austenite ends.

[0076] Ar3: the temperature at which austenite begins to transform to ferrite during cooling.

[0077] Ar1: the temperature at which austenite transforms to ferrite during cooling ends.

[0078] Bs: the temperature at which, during cooling, transformation of the austenite into bainite starts.

[0079] Ms: the temperature at which, during cooling, transformation of the austenite into martensite starts.

[0080] Mf: the temperature at which, during cooling, transformation of the austenite into martensite ends. A practical problem with Mf is that the martensite fraction during cooling approaches the maximum achievable amount only asymptotically, meaning that it takes very long for the last martensite to form. For practical reasons and in the context of this invention, Mf is therefore defined as the temperature at which 90% of martensite has formed.

[0081] These critical phase transformation temperatures can be determined by dilatometer experiments. Alternatively, the Bs and Ms points can be calculated beforehand, according to the standard practice.

[0082] Steel Microstructure

[0083] The steel according to the invention has a complex phase microstructure, comprising in vol. % 65% or more of a matrix of carbide free bainitic ferrite (BF) and/or tempered martensite (TM) and 5 to 25% of retained austenite (RA).

[0084] The matrix of a mixture of carbide free bainitic ferrite (BF) and/or tempered martensite (TM) ensures a good balance in the strength and ductility of the steel, while the retained austenite contributes to the TRIP effect. A matrix of BF may be obtained during austempering when the quenching stop temperature and the coiling temperature are in the temperature range between Ms and Bs−50° C. The BF in these steels is generally present in the form of plates with an ultrafine grain size (typically ~5 μm long and ~200 nm thick). The TM can be obtained during quenching and partitioning when the quenching stop temperature is between Ms and Mf and the coiling is conducted in the temperature range between Ms and Bs. The TM has a similar substructure to BF but a finer size of substructure. The precipitation of carbides between the ferritic laths, which is known to be detrimental to ductility, is suppressed by alloying with Si and/or Al. Such carbide free BF and/or TM microstructures are expected to provide high strength and good elongation due to the intermediate hard bainitic ferrite structure and to achieve good stretch-flangeability due to their uniform fine lath structure.

[0085] To ensure these properties, the BF and/or TM should be the matrix of the steel in a volume fraction of 65% or more, more preferably 70% or more. While the upper limit of the proportion depends on the balance between the bainitic ferrite structure and other structures, it is recommended to set the upper limit at about 95%, or preferably at 90%. Below the lower limit, the high strength cannot be guaranteed. Above the upper limit, insufficient retained austenite will be present, and thus elongation will be too low.

[0086] Retained austenite enhances ductility partly through the TRIP effect, which manifests itself in an increase in uniform elongation. The volume fraction of retained austenite is controlled to be 5% or higher, preferably 8% or higher. Below 5% the desired level of ductility and uniform elongation will not be achieved. The upper limit is set by the composition of the steel and processing parameters in a production line. For a given composition, the C content in the retained austenite becomes too low if the amount of the retained austenite is too high. The retained austenite is not stable enough and the stretch flangeability will deteriorate. Therefore, it is recommended that the volume fraction of retained austenite is within an upper limit of 25%, preferably less than 20%, more preferably less than 15%.

[0087] In a further embodiment of the invention, the matrix of the steel consists of carbide free BF only, and no TM is present. Such a hot rolled high strength steel is expected to provide a good balance of strength and ductility".

[0088] The concentration of C in the retained austenite has great influence on the TRIP (strain-induced transformation processing) characteristics. In a further embodiment of the invention, the retained austenite is effective in improving the elongation property when the carbon concentration is 0.90% or higher, preferably 0.95% or higher, and more preferably 1.00% or higher. While the upper limit of carbon in the retained austenite is 1.60% or lower, preferably 1.50% or lower, more preferably 1.40% or lower to ensure the TRIP effect.

[0089] The complex phase steel may further comprise amounts of other types of microstructures such as pro-eutectoid ferrite and fresh martensite, which volumes are dependent on both the alloying composition as well as the process settings.

[0090] The presence of some pro-eutectoid ferrite other than BF has a remarkable effect on the tensile properties and formability. The pro-eutectoid ferrite includes the polygonal ferrite (PF), typically formed in the hot rolling temperature range (above Ar1) and the acicular ferrite (AF), typically formed during cooling stage to the coiling temperature. Pro-eutectoid ferrite in general decreases yield strength and increases total elongation. Acicular ferrite, formed at lower temperatures during the fast cooling stage, has an acicular shape, is similar to bainitic ferrite in morphology and has some dislocations.

[0091] In a further embodiment of the invention, the hot rolled high strength steel sheet has a microstructure wherein the matrix further comprises acicular ferrite, and wherein the total volume of bainitic ferrite+tempered martensite+acicular ferrite is between 70 to 95%.

[0092] In a further embodiment of the invention, the microstructure comprises less than 20% acicular ferrite, preferably less than 15%. The presence of a small amount of acicular ferrite may increase global ductility significantly without sacrificing local ductility too much. Therefore, if present, the microstructure of the hot rolled high strength steel sheet comprises 3% or more acicular ferrite, preferably 5% or more. The formation of acicular ferrite allows more carbon to partition to the remaining austenite, which is beneficial for the stability of the retained austenite. However, the amount of acicular ferrite should not be too high for the local ductility of the steel.

[0093] In a further embodiment of the invention, the microstructure comprises less than 5% polygonal ferrite, preferably less than 3%, more preferably less than 1%. Most preferably, the microstructure is free from polygonal ferrite (PF). The PF is very soft and is detrimental to local ductility and leads to insufficient HEC. Therefore, the formation of PF is preferably controlled at lower levels.

[0094] In a further embodiment of the invention, the overall microstructure comprises less than 3% carbides, either present as separate phase or in the form of pearlite, preferably less than 2%, more preferably less than 1%. Most preferably, the microstructure is free from carbides as the formation of carbides is detrimental to strength and ductility.

[0095] In a further embodiment of the invention the microstructure also comprises fresh martensite. The amount of fresh martensite is preferably less than 15%, more preferably less than 10%, most preferably less than 5%. The presence of some fresh martensite can have a beneficial effect on work hardening rate. However over 15% it will negatively influence the hole expansion capacity.

[0096] The microstructure of the steel can be determined by optical microscopy, scanning electronic microscopy, x-ray measurement and dilatation test analysis, as well known in the art. Specific reference is made to chapter 3 of Micro-structural Quantification of Multi-Phase Steels (Micro-Quant the article Aarnts, M. P.; Rijkenberg, R. A. et al.). Contract No RFSR-CT-2006-00017; 2005-2009; Research project report; European Union: Brussels, (2011). The steel according to the invention will have good balanced properties of improved formability including increased ductility, improved hole expandability and bendability in combination with high strength as illustrated by the examples below.

[0097] In an embodiment of the invention, the hot rolled steel has a yield strength (YS) of at least 600 MPa, more preferably at least 700 MPa, most preferably at least 800 MPa. The yield strength is determined according to NEN-EN10002-1:2001. Steels with these higher yield strength are particularly suitable for automotive applications, as frame, wheel and other parts in cars. They can also be applied to structural materials for architecture, bridge, ship and rolling stock body.

[0098] In an embodiment of the invention, the hot rolled steel has a tensile strength (TS) of at least 950 MPa, more preferably at least 1050 MPa, most preferably 1150 MPa. The tensile strength is determined via NEN-EN10002-1:2001. Steels with these higher yield strength are particularly suitable for automotive applications, as frame, wheel and other parts in heavy trucks. They can also be applied to structural materials for architecture, bridge, ship and rolling stock body to reduce weight.

[0099] In a further embodiment of the invention the hot rolled steel has a total elongation (TE) of at least 10%, preferably at least 13%, more preferably at least 15%. The higher the elongation is, the better the ductility is. The total elongation is determined according to NEN-EN10002-1:2001.

[0100] In a further embodiment of the invention the hot rolled steel has a hole expansion capacity (HEC) of at least 20%, preferably at least 30%, more preferably at least 40%. With increasing HEC values, the risk for cracks during formation decreases, and the local formability becomes better. The HEC is measured according to the stretch flangeability evaluation test, in accordance with The Japan Iron and Steel Federation Standards JFS T 1001.

[0101] In a further embodiment of the invention, the hot rolled steel has a bending angle (BA) of 50° or more, preferably 60° or more, more preferably 70° or more. With increasing bending angle, the formability becomes better. The bending angle is determined via a three-point bending test according to the VDA 238-100 standard.

[0102] According to a second aspect of the invention a method is provided for producing the hot rolled high strength steel sheet according to the first aspect of the invention, comprising the steps of:

- **[0103]** Casting a slab having a chemical composition as described above,
- **[0104]** Reheating the slab above the Ac3 temperature, preferably above 1150° C.,
- **[0105]** Hot rolling the slab in a temperature range in which the structure of the steel is entirely austenitic,
- **[0106]** Finishing the hot rolling at a finish hot rolling temperature higher than the Ar3 transformation point,
- **[0107]** cooling the hot-rolled steel strip at a cooling rate (Vcool) of 20 to 100° C./s to a stop temperature (Tstop) below Bs−25° C. and above Mf+25° C.,
- **[0108]** coiling the hot-rolled steel strip at a temperature (Tcoil) below Bs−80° C. and above Ms.

[0109] The method according to the invention provides a manufacturing process of the hot rolled steel sheet, in which small variations in the parameters do not cause substantial modifications to the microstructure and to the mechanical properties. The hot-rolling is carried out as well known in the art. The finish rolling temperature, cooling rate, stop temperature and coiling temperature are important to obtain a steel sheet with the desired properties.

[0110] In general, the manufacturing process is carried out as follows: a steel of the composition as described above is made and the slab is casted using conventional methods such as ingot casting or continuous casting.

[0111] The cast slab is reheated to a temperature above the Ac3 point and preferably above 1150° C. to completely dissolve the added alloying elements in the austenite phase to form a solid solution prior to hot rolling. The optimum reheating temperature depends on the alloying elements. However, the inventors found that a suitable range is above the Ac3, preferably between 1150° C. and 1250° C. to obtain homogeneous microstructures.

[0112] The slab is hot-rolled in a temperature range in which the structure of the steel is entirely austenitic. A suitable temperature for hot rolling is between 1150° C. and 1250° C.

[0113] Standard hot rolling processes can be applied. The final sheet thickness is normally based on the final application of the steel sheet.

[0114] Hot rolling is completed by finishing hot rolling at a temperature being above the Ar3 point of the steel, preferably in the range of 800-950° C., whereby the austenite phase is refined into fine grains.

[0115] If the finish rolling temperature is too low, premature formation of ferrite may occur during hot rolling to form polygonal ferrite, thereby deteriorating the press formability of the resulting hot rolled steel sheet. On the other hand, if the finish rolling temperature is above 950° C., the austenite grain size may become too large, which affect the formation of the pro-eutectoid ferrite in the following cooling step and the formation of the bainitic ferrite during coiling, thereby deteriorating the strength and ductility.

[0116] The strip is then cooled at a cooling rate (Vcool) of 20 to 100° C./s to a temperature Tstop below Bs−25 and above Mf+25° C., preferably below Bs−50° C. and above Mf+50° C. During cooling from hot rolling to the Tstop temperature, some acicular ferrite is formed, which has an acicular shape and a fine grain size.

[0117] If the cooling rate (Vcool) after hot rolling is lower than 20° C./s, pro-eutectoid ferrite may form in an excess amount and/or pearlite may form during cooling. The presence of too much amount of the ferrite can increase the elongation but significantly reduce the strength and HEC value. If Vcool is higher than 100° C./s acicular ferrite will not be formed in an amount sufficient to enable the retention of a substantial amount of austenite. The Vcool is preferably in the range of 25 to 80° C./s, more preferably 30 to 60° C./s.

[0118] The Tstop temperature should be below Bs−25° C., more preferably Bs−50° C. If the temperature Tstop is too high, proeutectoid polygonal ferrite or pearlite may form in an excess amount during cooling, which will significantly reduce the yield strength and HEC value.

[0119] The Tstop should be above Mf+25° C., more preferably above Mf+50° C. to ensure the presence of a sufficient amount of the retained austenite, which is necessary to obtain the TRIP effect.

[0120] Coiling is then carried out at a temperature Tcoil below Bs−80° C. and above Ms, preferably below Bs−100° C. and above Ms, to obtain a microstructure of at least 65% of a matrix of carbide-free bainitic ferrite and tempered martensite. If Tcoil is too high, the bainitic ferrite becomes too coarse and carbides might be formed along the original austenite grain boundaries or in the form of the pearlite microstructure. As a consequence, the required amount of the BF and the retained austenite will not be obtained. If Tcoil is too low, bainite transformation may be not complete and the carbon enrichment of the austenite may be insufficient. The cooling rate during coil cooling (Vcoil) is preferably less than 0.05° C./s, more preferably 0.03° C./s to allow bainitic transformation to complete before the strip temperature reaches 300° C.

[0121] In one embodiment of the invention, the Tstop temperature is below Bs−25° C. and above Ms, preferably below Bs−50° C. and above Ms, and the Tcoil is below Bs−80° C. and above Ms. This process results in a steel sheet with a carbide-free bainitic ferrite (BF) base microstructure. In this case, if the Tstop is higher than the Tcoil temperature, the hot rolled strip is cooled down to the Tcoil in the run-out table, preferably with a cooling rate (Vrot) not lower than 1° C./s to avoid the formation of carbides. The natural cooling in the air is sufficient for this purpose. If the Tstop is lower than the Tcoil temperature, the latent heat generated by bainitic phase transformation will heat the strip up to the Tcoil in the run-out table. Thus the matrix of BF can be obtained during austempering when the quenching stop temperature (Tstop) and the coiling temperature (Tcoil) are in the temperature range of above Ms and below Bs−25° C., as shown in FIG. **1** which depicts a schematic temperature (vertical) vs. time (horizontal) diagram for austempering steel to obtain a BF base microstructure.

[0122] In an alternative embodiment according to the invention, the Tstop is above Mf+25° C. and below Ms, and the Tcoil is below Bs−80° C. and above Ms. This process results in a steel sheet with a matrix of a mixture of carbide-free bainitic ferrite and tempered martensite. Without wishing to be bound by theory, it is believed that some amount of martensite is formed at Tstop temperature and then is tempered during coiling. In general, the lower the Tstop temperature is, the more the amount of tempered martensite is. The Tstop should be above Mf+25° C., preferably above Mf+50° C. to ensure the presence of a sufficient amount of the retained austenite. In this case, the hot rolled strip is heated in the run-out table to the Tcoil temperature for coiling. The heating can be realized by adding an induction heating unit in the run-out table. The heating time between Tstop and Tcoil is preferably between 1 second and 1 hour. Thus, the TM can be obtained during quenching and partitioning when the quenching stop temperature (Tstop) is between Ms and Mf+25° C. and the coiling is conducted in the temperature range between Ms and Bs−80° C., as shown in FIG. **2**, which depicts a schematic temperature (vertical) vs. time (horizontal) diagram for quenching and partitioning the steel to obtain a TM+BF base microstructure.

[0123] The steel according to the invention can be further processed according to common procedures known in the art, such as pickling and surface treatment.

EXAMPLES

[0124] Examples are performed using laboratory cast ingots.

[0125] Steels A to K having compositions according to the invention, as shown in Table 1, were cast into 25 kg ingots of 200 mm×110 mm×110 mm in dimensions using vacuum induction. Then, the following process schedule was used to manufacture hot rolled strips of 3 mm thickness to mimic the manufacturing process.

- **[0126]** Reheating of the ingots at 1225° C. for 2 hours;
- **[0127]** Rough rolling of the ingots from 140 mm to 35 mm;
- **[0128]** Reheating of the rough-rolled ingots at 1200° C. for 30 min;
- **[0129]** Hot rolling from 35 mm to 3 mm (35-27-19-11-7-5-3 mm) with the finish rolling temperature (FRT) at about 880±40° C.;
- **[0130]** Run out table cooling according to the process parameters as given in Table 2;
- **[0131]** Cooling the steel sheet at a cooling rate (Vcool);
- **[0132]** Transfer the sheet to a preheated furnace at coiling temperature (Tcoil);
- **[0133]** Cooling the sheet with furnace cooling to room temperature to simulate the coiling process;
- **[0134]** Pickling the hot rolled sheets in HCl at 85° C. to remove the oxide layers;
- **[0135]** Samples for microstructure observations, tensile tests and hole expansion tests were machined from hot rolled strips.

[0136] Dilatometry was done on the rolled samples of 10 mm (length)×5 mm (width)×3 mm (thickness) dimensions (length along the rolling direction). The samples were heated at a rate of 10° C./s, held at the austenizing temperature of 900° C. for 2 minutes and then quenched at 100° C./s to room temperature. The critical phase transformation points were determined from the dilatometry curves and are given in Table 1.

[0137] Tensile tests—JIS5 test pieces (gauge length=50 mm; width=25 mm) were machined from the obtained hot-rolled sheets such that the tensile direction was parallel to the rolling direction. Room temperature tensile tests were performed in a Schenk TREBEL testing machine following NEN-EN10002-1:2001 standard to determine tensile properties (yield strength YS (MPa), ultimate tensile strength UTS (MPa), total elongation TE (%)). For each condition, three tensile tests were performed and the average values of mechanical properties are reported.

[0138] Hole Expansion Test (Stretch Flangeability Evaluation Test)—Test pieces for testing hole expandability (size: 90×90 mm) were sampled from the obtained rolled sheet. In accordance with The Japan Iron and Steel Federation Standards JFS T 1001, a 10 mm diameter punch hole was punched in the centre of the test piece and a 60° conical punch was pushed up and inserted into the hole. When a crack penetrated the sheet thickness, the hole diameter d (mm) was measured. The hole expansion ratio HEC (%) was calculated by the following equation: $HEC(\%)=\{(d-d_0)/d_0\}\times100$, with do being 10 mm.

[0139] Bending test—Bending specimens (40 mm×30 mm) from parallel and transverse to rolling directions were prepared from each of the conditions and tested by three-point bending test according to the VDA 238-100 standard. The experiments were stopped at different bending angles and the bent surface of the specimen was inspected for identification of failure in order to determine the bending angle (BA). The bending angles of the samples with bending axis parallel to the rolling direction are lower than those of the samples with bending axis perpendicular to the rolling direction. For each type of tests, three samples were tested and the average values from three tests are presented for each condition.

[0140] The tensile properties, the HEC and minimal BA value (the bending axis parallel to the rolling direction) are given in Table 2.

[0141] The amount of retained austenite (RA) and the carbon content in the retained austenite were determined using X-ray measurement (XRD) and subsequent Rietveld analysis. Reference is made to Onink M et. al, "The lattice parameters of austenite and ferrite in Fe—C alloys as functions of carbon concentration and temperature". Scr. Metal. Mater. 29: 1011-1016 (1993). The microstructures were evaluated by optical microscopy and scanning electronic microscopy. The volume fractions were obtained from dilatation tests in combination with microstructural analysis. The volume fractions of microstructure are given in Table 3.

[0142] From the examples in the tables, it is clear that a steel according to the invention has excellent balanced properties both in terms of strength and ductility. From the steels that aren't according to the invention, the properties and/or the microstructures that do not meet the requirements are underlined. The microstructure of examples A3, A4, E3 and F3 for example comprise carbides along the original grain boundaries and/or pearlite microstructure and a low amount of the retained austenite, resulting in low tensile strength. In addition A3 & A4 show a mediocre performance in the HEC test, whereas Al & A2, which are according to the invention, show excellent properties on all tests. It is also clear from table 2 and 3 that the coiling temperature (Tcoil) and the cooling rate (Vcool) have a large effect on the microstructure and tensile properties. In general the tensile strength and the HEC value decrease as the coiling temperature is increased. As clear from K1 and K2, a low Vcool can result in a higher amount of acicular ferrite, thereby influencing the final properties of the steel.

TABLE 1

Steel compositions (wt %) and the phase transformation points (° C.)

Alloy	C	Si	Al	Mn	P	S	N	Ti	Cr	Mo	B	Ac3	Bs	Ms	Mf
A	0.30	1.02	0.360	1.82	<0.005	<0.003	0.0035	<0.005	0.31	<0.005	<0.0005	855	522	330	213
B	0.31	1.01	0.350	1.84	<0.005	<0.003	0.0042	<0.005	0.31	0.26	<0.0005	851	520	326	208
C	0.31	1.03	0.370	1.86	<0.005	<0.003	0.0038	0.029	0.31	<0.005	0.0015	856	518	308	199
D	0.31	0.89	0.090	2.02	0.011	<0.003	0.0030	0.027	1.03	<0.005	<0.0005	821	506	308	189
E	0.25	1.46	0.026	1.80	<0.005	<0.003	0.0043	0.020	<0.04	<0.005	<0.0005	865	522	347	232
F	0.25	0.97	0.370	1.78	<0.005	<0.003	0.0042	0.040	<0.04	<0.005	<0.0005	875	539	358	243
G	0.26	1.02	0.350	1.82	<0.005	0.012	0.0032	0.030	0.32	<0.005	<0.0005	870	534	351	236
H	0.25	1.50	0.025	1.96	<0.005	0.014	0.0031	0.030	0.95	<0.005	<0.0005	874	521	357	242
J	0.24	1.02	0.036	2.25	<0.005	<0.003	0.0045	<0.005	<0.04	<0.005	<0.0005	830	541	353	186
K	0.25	0.49	0.830	1.75	<0.005	0.010	0.0040	0.030	<0.04	<0.005	<0.0005	899	562	362	248

TABLE 2

Process parameters and mechanical properties

Alloy-process	FRT (° C.)	Vcool (° C./s)	Tstop (° C.)	Tcoil (° C.)	YS (MPa)	TS (MPa)	YS/ TS	UE (%)	TE (%)	BA (°)	HEC (%)	Inv.
A-1	868	24	415	400	754	1055	0.71	10.7	17.8	72	42	y
A-2	871	24	438	425	710	1010	0.70	13.5	22.2	93	30	y
A-3	869	26	455	450	719	992	0.72	9.9	16.3	59	19	n
A-4	884	24	504	480	692	921	0.75	8.2	15.2	74	17	n
A-5	853	49	459	400	892	1169	0.76	5.7	10.2	78	50	y
A-6	866	49	509	425	726	1039	0.70	9.2	14.4	55	32	y
B-1	885	25	436	425	640	1095	0.58	12.1	19.1	57	26	y
C-1	870	23	463	425	643	1089	0.59	12.9	20.4	61	24	y

TABLE 2-continued

Process parameters and mechanical properties

Alloy-process	FRT (° C.)	Vcool (° C./s)	Tstop (° C.)	Tcoil (° C.)	YS (MPa)	TS (MPa)	YS/TS	UE (%)	TE (%)	BA (°)	HEC (%)	Inv.
D-1	862	24	456	450	765	1326	0.58	10.2	14.3	65	29	y
E-1	882	31	446	400	700	1002	0.70	9.1	16.2	78	47	y
E-2	875	46	445	400	877	1133	0.77	5.8	12.5	62	50	y
E-3	854	46	505	450	780	941	0.83	7.5	12.8	77	42	n
F-1	886	33	418	400	717	952	0.75	7.5	15.9	102	41	y
F-2	862	48	463	400	955	1272	0.75	4.8	11.7	84	56	y
F-3	849	41	537	450	760	905	0.84	5.1	12.9	91	52	n
G-1	875	26	407	400	753	1014	0.74	8.8	16.3	79	67	y
G-2	852	49	458	400	759	1013	0.75	8.5	15.7	69	66	y
G-3	857	49	486	425	733	1026	0.71	8.6	14.2	64	44	y
H-1	881	23	436	375	730	1297	0.56	9.8	16.5	74	30	y
J-1	850	40	475	425	718	1226	0.59	12	16.8	99	28	y
J-2	853	40	350	425	793	1127	0.70	11.8	16.9	101	35	y
J-3	865	62	300	425	850	1160	0.73	8.6	12.4	87	50	y
J-4	862	62	250	425	910	1230	0.74	11.5	15.7	93	47	y
K-1	891	15	426	400	809	915	0.88	4.1	10.4	159	51	y
K-2	856	46	469	400	897	1079	0.83	6.1	12.5	151	57	y

* YS—yield strength; UE—uniform elongation; TE—total elongation; BA—bending angle; HEC—hole expansion capacity, Inv.—According to the invention.

TABLE 3

Corresponding microstructures of the alloys of table 2.

Alloy-process	PF* (%)	AF (%)	P (%)	BF (%)	TM (%)	M (%)	RA (%)	C in RA (wt %)	Invention
A-1	2.2	10.5	0	72	0	2.8	12.5	1.03	y
A-2	0	12.2	0	66	0	3.6	18.2	0.98	y
A-3	0	12.5	5.6	81.4	0	0	0.5	—	n
A-4	0	15.3	20.3	64.4	0	0	0	—	n
A-5	2.5	5.5	0	77.3	0	5.2	9.5	1.05	y
A-6	3.6	6.2	0	74.3	0	4.6	11.3	1.02	y
B-1	0	12.8	0	69.5	0	6.2	11.5	1.11	y
C-1	0	15.5	0	65.4	0	4.8	14.3	0.96	y
E-1	0	10.6	0	76.4	0	5.2	7.8	1.10	y
E-2	0	3.3	0	83.7	0	6.3	6.7	1.07	y
E-3	0	3.2	6.3	90.3	0	0	0.2	—	n
F-1	0	3.2	0	83.6	0	4.5	8.7	1.12	y
F-2	2.5	0	0	83.8	0	6.1	7.6	1.06	y
F-3	0	0	9.8	90.1	0	0	0.1	—	n
H-1	0	12.4	0	73.9	0	5.2	8.5	0.99	y
J-1	0	5.3	0	76	0	6.5	12.2	1.02	y
J-2	0	4.9	0	32.1	49.5	2.2	11.3	1.16	y
J-3	0	0	0	21.1	68.3	1.1	9.5	1.08	y
J-4	0	0	0	14.6	74.6	0	10.8	1.05	y
K-1	0	23.5	0	66.5	0	3.5	6.5	1.12	y
K-2	0	5	0	83.4	0	4.8	6.8	1.11	y

*PF—polygonal ferrite; AF—acicular ferrite; P—carbides and pearlite; BF—bainitic ferrite; TM—tempered martensite; M—fresh martensite; RA—retained austenite.

1. A hot rolled high strength steel sheet having a tensile strength of at least 950 MPa consisting of the following elements in weight %:

0.13-0.35 C;
0.80-3.50 Mn;
0.01-2.50 Si;
0.02-1.00 Al;
wherein the sum of Si and Al is 0.52-2.50; and
optionally one or more selected from:
0.0002-0.0030 Ca;
0.0004-0.0100 REM;
0.10-1.00 Cu;
0.10-1.50 Cr;
0.10-1.00 Ni;
0.05-0.50 Mo;
0.0005-0.0050 B;
0.010-0.100 Nb;
0.010-0.100 Ti;
0.020-0.200 V;
wherein the sum of Nb+Ti+V is 0-0.250 and
wherein the sum of Cr+Ni+Cu+Mo is 0-1.50, and
at most 0.100 P;
at most 0.050 S;
at most 0.0100 N;
the balance consisting of inevitable impurities and Fe,
wherein the hot rolled high strength steel sheet has a complex phase microstructure comprising in vol. %:
65% or more of a matrix of a mixture of carbide free bainitic ferrite (BF) and/or tempered martensite (TM), and

5 to 25% of retained austenite (RA),
and wherein the microstructure comprises less than 3% carbides.

2. The hot rolled high strength steel sheet according to claim **1**, wherein for one or more elements the following range(s) is/are valid:
0.15-0.32 C;
1.20-3.00 Mn;
0.50-2.00 Si;
0.02-0.80 Al,
wherein the sum of Si and Al is 0.80-2.00; and
optionally one or more selected from:
0.0005-0.0020 Ca;
0.0005-0.0080 REM;
0.15-0.80 Cu;
0.15-1.00 Cr;
0.10-1.00 Ni;
0.10-0.30 Mo;
0.0007-0.0040 B;
0.030-0.050 Nb;
0.030-0.050 Ti;
0.030-0.100 V;
wherein the sum of Nb+Ti+V is 0-0.100;
at most 0.050 P;
at most 0.010 S;
at most 0.0050 N.

3. The hot rolled high strength steel sheet according to claim **1**,
wherein for one or more elements the following range(s) is/are valid:
0.17-0.30 C;
1.40-2.60 Mn;
0.50-1.50 Si;
0.02-0.60 Al,
wherein the sum of Si and Al is 1.00-1.60; and
optionally one or more selected from:
0.0005-0.001 Ca;
0.0005-0.008 REM;
wherein Cu belongs to the inevitable impurities, and/or
wherein Cr belongs to the inevitable impurities, and/or
wherein Ni belongs to the inevitable impurities, and/or
wherein Mo belongs to the inevitable impurities, and/or
wherein B belongs to the inevitable impurities, and/or
wherein Nb belongs to the inevitable impurities, and/or
wherein Ti belongs to the inevitable impurities, and/or
wherein V belongs to the inevitable impurities, and or;
at most 0.020 P;
at most 0.005 S;
at most 0.0040 N.

4. The hot rolled high strength steel sheet according to claim **1**, wherein the retained austenite has a carbon concentration of 0.90-1.60%

5. The hot rolled high strength steel sheet according to claim **1**, wherein the microstructure further comprises acicular ferrite (AF), and wherein the total volume of BF+TM+AF is 70-95%.

6. The hot rolled high strength steel sheet according to claim **1**, wherein the microstructure contains less than 5% polygonal ferrite.

7. The hot rolled high strength steel sheet according to claim **1**, wherein the microstructure comprises less than 20% acicular ferrite.

8. The hot rolled high strength steel sheet according to claim **1**, wherein the microstructure comprises less than 15% fresh martensite.

9. The hot rolled high strength steel sheet according to claim **1**, having a total elongation of at least 10%.

10. The hot rolled high strength steel sheet according to claim **1**, having a hole expansion capacity (HEC) of at least 20%.

11. A method for producing a hot rolled high strength steel sheet according to claim **1**, comprising the steps of
casting a slab having a chemical composition according to claim **1**,
Reheating the slab above the Ac3 temperature,
Hot rolling the slab in a temperature range in which the structure of the steel is entirely austenitic,
Finishing the hot rolling at a finish hot rolling temperature higher than the Ar3 transformation point,
cooling the hot-rolled steel strip at a cooling rate (Vcool) of 20 to 100° C./s to a stop temperature (Tstop) below Bs−25° C. and above Mf+25° C.,
Coiling the hot-rolled steel strip at a temperature (Tcoil) below Bs−80° C. and above Ms.

12. The method of claim **11**, wherein the cooling rate (Vcool) is between 25 and 80° C./s.

13. The method of claim **11**, wherein Tstop is below Bs−25° C. and above Ms, preferably below Bs−50° C.

14. The method of claim **11**, wherein Tstop is below Ms and above Mf+25° C., preferably above Mf+50° C.

15. The method of claim **11**, wherein Tcoil is below Bs−100° C. and above Ms.

16. The hot rolled high strength steel sheet according to claim **1**, wherein the microstructure contains less than 5% polygonal ferrite, wherein the microstructure comprises less than 20% acicular ferrite, wherein the microstructure comprises less than 15% fresh martensite.

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