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(54) **HIGH-STRENGTH STEEL SHEET WITH EXCELLENT PROCESSABILITY AND PROCESS FOR PRODUCING SAME**

HOCHFESTES STAHLBLECH MIT AUSGEZEICHNETER VERARBEITBARKEIT UND VERFAHREN ZUR HERSTELLUNG DAVON

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Description

Technical Field

5 **[0001]** The present invention relates to a high strength steel sheet which is required to have excellent formability (stretch flangeability) and is suitable for use as a strength member or the like of automobile parts, and a method of manufacturing the high strength steel sheet.

Background Art

10 **[0002]** In recent years, an improvement in fuel economy of vehicles has become an important issue from the point of view of preservation of the global environment. Thus, trends toward increasing the strength of the materials used, decreasing the thickness of the structural members used, and reducing the weight of the automobile bodies have accelerated. As materials to be used, high strength steel sheets having a tensile strength of 540 MPa or higher are
15 particularly required. However, since increasing the strength of steel sheets leads to a decrease in formability, high strength steel sheets having excellent formability are in demand. This demand is particularly high for steel sheets having a small thickness (thin steel sheets).

[0003] To meet such demand, various multi phase steel sheets have been proposed such as a dual phase steel sheet (DP steel sheet) having a dual phase microstructure composed of a ferrite phase and a martensite phase and a steel
20 sheet having a multi phase microstructure including a ferrite phase, a martensite phase, and a bainite phase.

[0004] For example, PTL 1 describes a method of manufacturing a high strength cold rolled steel sheet having excellent local ductility. According to this method, a cold rolled steel sheet having a composition including 0.08 to 0.30% of C, 0.1 to 2.5% of Si, 0.5 to 2.5% of Mn, and 0.01 to 0.15% of P is subjected to recrystallization annealing at a temperature equal to or higher than an A_{c1} point, forcibly air-cooled to a temperature region ranging from an A_{r1} point to 600°C,
25 rapidly cooled at a cooling rate equal to or higher than 100°C/s to form a multi phase microstructure composed of a ferrite phase and a low-temperature transformed phase, and overaged at a temperature in the range of 350°C to 600°C so that a ratio $H_v(L)/H_v(\alpha)$ of the hardness $H_v(L)$ of the low-temperature transformed phase to ferrite hardness $H_v(\alpha)$, which is obtained by a predetermined relational expression, is in the range of 1.5 to 3.5. According to the technique described in PTL 1, the volume fraction of the low-temperature transformed phase is increased by increasing a quenching
30 start temperature and then the overaging is performed at a temperature of 350°C to 600°C to precipitate C in the ferrite, soften the low-temperature transformed phase, and thereby reduce the ratio $H_v(L)/H_v(\alpha)$ and improve local elongation.

[0005] PTL 2 describes a method of manufacturing a high tensile hot rolled steel sheet with a low yield ratio and excellent corrosion resistance, the method including hot-rolling a steel slab containing 0.02 to 0.25% of C, 2.0% or less of Si, 1.6 to 3.5% of Mn, 0.03 to 0.20% of P, 0.02% or less of S, 0.05 to 2.0% of Cu, 0.005 to 0.100% of sol.Al, and
35 0.008% or less of N to form a hot rolled coil, pickling the hot rolled coil, and annealing the hot rolled coil at a temperature of 720°C to 950°C by a continuous annealing line. According to the technique described in PTL 2, it is possible to manufacture a high tensile hot rolled steel sheet which maintains a low yield ratio, high ductility, and excellent hole expandability, exhibits excellent corrosion resistance, and has a multi phase microstructure.

[0006] PTL 3 describes a high strength cold rolled steel sheet with an excellent balance between the strength and the stretch flangeability. This high strength cold rolled steel sheet has a composition containing 0.03 to 0.17% of C, 1.0% or less of Si, 0.3 to 2.0% of Mn, 0.010% or less of P, 0.010% or less of S, and 0.005 to 0.06% of Al and satisfying $C(\%) > (3/40) \times Mn$, has a microstructure composed of a ferrite phase and a second phase including mainly bainite or pearlite, and satisfies $(\text{Vickers hardness of second phase})/(\text{Vickers hardness of ferrite phase}) < 1.6$. The high strength cold rolled steel sheet described in PTL 3 is obtained by an annealing treatment followed by an overaging treatment at a temperature
45 of 500 to 250°C. In this annealing treatment, steel (slab) having the above-described composition is hot-rolled, coiled at a temperature equal to or lower than 650°C, pickled, cold-rolled, soaked at a temperature equal to or higher than an A_1 point and equal to or lower than (A_3 point +50°C), gradually cooled to a temperature T_1 in the range of 750°C to 650°C at a rate of 20°C/s or lower, and cooled T_1 to 500°C at a rate of 20°C/s or higher.

Citation List

Patent Literatures

55 **[0007]**

[PTL 1] Japanese Unexamined Patent Application Publication No. 63-293121

[PTL 2] Japanese Unexamined Patent Application Publication No. 05-112832

[PTL 3] Japanese Unexamined Patent Application Publication No. 10-60593 JPH09118952 discloses a high strength

hot rolled steel sheet member made by hot rolling a steel comprising (by wt.) 0.05-0.20 % C, 0.01-1.0 % Si, 0.50-2.0 % Mn, respectively up to 0.05 % P, 0.03 % S, 0.08 % sol. Al, and 0.01 % N, and balance Fe, at the finishing temperature of the Ar3 point, winding it at 500-650°C, cold working the sheet, reheating it to at least the Ac3 point, hot working, and cooling it to give a structure of over 30% ferrite and over 30% pearlite.

Summary of Invention

Technical Problem

[0008] However, the technique described in PTL 1 has problems in that a continuous annealing facility which can perform rapid cooling (quenching) after recrystallization annealing is required, and addition of large amounts of alloy elements is required to suppress a rapid decrease in strength due to the overaging at a high temperature.

[0009] In the technique described in PTL 2, it is essential to add large amounts of P and Cu in combination. However, when a large amount of Cu is contained, hot formability decreases, and when a large amount of P is contained, steel is embrittled. In addition, P shows a marked tendency to segregate in the steel, and this segregated P causes problems such as a decrease in the stretch flangeability of the steel sheet and embrittlement of a welded portion.

[0010] The high strength cold rolled steel sheet described in PTL 3 has excellent stretch flangeability. However, at a strength as high as 540 MPa or higher, the elongation is less than 26% and a problem occurs in that the elongation sufficient for maintaining desired excellent formability cannot be ensured.

[0011] An object of the invention is to address such problems in the related art and to provide a high strength steel sheet having a small sheet thickness of about 1.0 to 1.8 mm and excellent formability and a method for manufacturing the high strength steel sheet. Here, the "high strength" means that the steel sheet has a tensile strength TS equal to or higher than 540 MPa and preferably equal to or higher than 590 MPa. The "excellent formability" means that the elongation El is equal to or greater than 30% (with a JIS No. 5 test piece) and a hole expanding ratio λ in a hole expanding test based on the Japan Iron and Steel Federation Standard JFST 1001-1996 is equal to or higher than 80%. Solution to Problem

[0012] In order to achieve the above-described object, the inventors of the invention have conducted intensive studies on the effect of the composition and the microstructure on the strength and formability. As a result, they have found that a microstructure composed of a ferrite phase as a main phase and a second phase including mainly fine pearlite can be formed by subjecting a hot rolled sheet in which the amounts of alloy elements are adjusted within an appropriate range to an annealing treatment which includes heating to an appropriate dual phase temperature region and an appropriate cooling treatment without cold rolling, and thus a desired high strength can be ensured, formability is significantly improved, and a high strength steel sheet having excellent formability with desired elongation and a desired hole expanding ratio can be obtained.

[0013] The detailed mechanism with regard to a significant improvement in formability by directly performing an appropriate annealing treatment on a hot rolled sheet without cold rolling has not been clear until now, but following has been presumed by the inventors of the invention.

[0014] When a hot rolled sheet is subjected to an annealing treatment of heating the sheet to a dual phase temperature region without cold rolling, the only transformation that occurs during the annealing heating is from α to γ and new recrystallization does not occur. In this case, transformation from α to γ only occurs preferentially in portions with a high C concentration, and a more uniform microstructure can be obtained. In addition, C that diffuses rapidly is redistributed into α and γ up to an equilibrium composition during the annealing treatment. Therefore, presumably, precipitation of film cementite at grain boundaries has been suppressed, which particularly contributes to improvements in stretch flangeability. In contrast, when the hot rolled sheet is cold-rolled and then subjected to an annealing treatment, recrystallization and α to γ transformation competitively occur during the annealing heating, and thus the microstructure tends to become nonuniform and a significant improvement in formability is not readily expected.

[0015] The invention has been completed on the basis of such findings with further examination. That is, the gist of the invention is as follows.

(1) A high strength steel sheet having excellent formability which has a composition consisting of, including, by mass%, 0.08 to 0.15% of C, 0.5 to 1.5% of Si, 0.5 to 1.5% of Mn, 0.1% or less of P, 0.01% or less of S, 0.01 to 0.1% of Al, 0.005% or less of N, optionally one or more of 0.05 to 0.5% of Cr, 0.005 to 0.2% of V, 0.005 to 0.2% of Mo, 0.01 to 0.1% of Ti, 0.01 to 0.1% of Nb, 0.0003 to 0.0050% of B, 0.05 to 0.5% of Ni, 0.05 to 0.5% of Cu, 0.001 to 0.005% of Ca, 0.001 to 0.005% of REM, and the balance Fe with inevitable impurities, the steel sheet having a microstructure composed of a ferrite phase which is a main phase and a second phase including at least pearlite, wherein an area fraction of the ferrite phase is in the range of 75 to 90% and an area fraction of the pearlite is in the range of 10 to 25% with respect to the entire microstructure, an average grain size of the pearlite is 5 μ m or smaller, and an area fraction of the pearlite is 70% or greater with respect to the total area of the second phase.

(7) A method for manufacturing a high strength steel sheet having excellent formability, including: a hot rolling step of hot-rolling a steel having a composition which consists of, by mass%, 0.08 to 0.15% of C, 0.5 to 1.5% of Si, 0.5 to 1.5% of Mn, 0.1% or less of P, 0.01% or less of S, 0.01 to 0.1% of Al, 0.005% or less of N, optionally one or more of 0.05 to 0.5% of Cr, 0.005 to 0.2% of V, 0.005 to 0.2% of Mo, 0.01 to 0.1% of Ti, 0.01 to 0.1% of Nb, 0.0003 to 0.0050% of B, 0.05 to 0.5% of Ni, 0.05 to 0.5% of Cu, 0.001 to 0.005% of Ca, 0.001 to 0.005% of REM, and the balance Fe with inevitable impurities to form a hot rolled sheet; and a continuous annealing step including an annealing treatment of pickling the hot rolled sheet, and holding the pickled hot rolled sheet in a first temperature region of an A_{c1} transformation point to an A_{c3} transformation point for 5 to 400 s, and a cooling treatment of cooling the sheet at an average cooling rate of 5 °C/s or higher from the first temperature region to 700°C after the annealing treatment and adjusting a residence time in a second temperature region of 700°C to 400°C in the range of 30 to 400 s by using a continuous annealing line.

(8) The method for manufacturing a high strength steel sheet according to (7), in which the hot rolling step includes heating the steel at a temperature in the range of 1100°C to 1280°C, hot-rolling the heated steel with a finish hot rolling temperature of 870°C to 950°C to form a hot rolled sheet, and coiling the hot rolled sheet at a coiling temperature of 350°C to 720°C upon completion of the hot rolling.

(9) The method for manufacturing a high strength steel sheet according to (7) or (8), in which a cooling time in a temperature region of 700°C to 550°C in the second temperature region is 10 s or longer.

Advantageous Effects of Invention

[0016] According to the invention, a high strength steel sheet with excellent formability having a high strength, i.e., a tensile strength TS of 540 MPa or higher, an elongation El of 30% or greater, and a stretch flangeability λ of 80% or higher can be easily manufactured at a low cost, and thus the invention has particularly significant industrial advantages. The invention also has an effect of significantly contributing to reduction of manufacturing cost, improvements of productivity, etc., since cold rolling can be omitted. When the steel sheet of the invention is applied to parts of automobile bodies, it can significantly contribute to the weight-reduction of automobile bodies. Description of Embodiments

[0017] First, reasons for limitations on the composition of a steel sheet of the invention will be described. Hereinafter, mass% will be simply expressed by % unless otherwise noted.

C: 0.08 to 0.15%

[0018] C is an element that contributes to an increase in the strength of a steel sheet and effectively acts on formation of a multi phase microstructure composed of a ferrite phase and a second phase other than the ferrite phase. In the invention, 0.08% or more of C is required to be contained in order to ensure a desired high strength, i.e., a tensile strength of 540 MPa or higher. When more than 0.15% of C is contained, spot weldability is lowered and formability such as ductility is lowered. Therefore, the C content is limited in the range of 0.08 to 0.15% and preferably 0.10 to 0.15%.

Si: 0.5 to 1.5%

[0019] Si is an element that dissolves in steel and effectively acts to strengthen the ferrite, and also contributes to an improvement in ductility. 0.5% or more of Si is required to be contained in order to ensure a desired high strength, i.e., a tensile strength of 540 MPa or higher. When an excessive amount more than 1.5% of Si is contained, generation of red scale and the like is accelerated, the surface quality of a steel sheet is lowered, and chemical treatability is lowered. When an excessive amount of Si is contained, resistance weldability is deteriorated with an increase in electric resistance in resistance welding. Therefore, the Si content is limited in the range of 0.5 to 1.5% and preferably 0.7 to 1.2%.

Mn: 0.5 to 1.5%

[0020] Mn is an element that contributes to an increase in the strength of a steel sheet and effectively acts on formation of a multi phase microstructure. 0.5% or more of Mn is required to be contained in order to obtain such an effect. When more than 1.5% of Mn is contained, a martensite phase is easily formed in the course of cooling in the annealing, and thus formability, particularly, stretch flangeability, is lowered. Therefore, the Mn content is limited in the range of 0.5 to 1.5% and preferably 0.7 to 1.5%.

P: 0.1% or less

[0021] P is an element that dissolves in steel and acts to increase the strength of a steel sheet. However, P shows a marked tendency to segregate to grain boundaries and lowers the bonding power of grain boundaries. This results in a

decrease in formability and concentration of P in the surface of the steel sheet, thereby decreasing chemical treatability and corrosion resistance. Such an adverse effect of P is notably shown when more than 0.1% of P is contained. Therefore, the P content is limited to 0.1% or less. In order to avoid such an adverse effect of P, the P content is preferably decreased to 0.1% or less as much as possible. However, the excessive decrease leads to a rise in manufacturing cost, and thus the P content is preferably about 0.001% or more.

S: 0.01% or less

[0022] S mainly forms sulfides (inclusions) such as MnS in steel and lowers formability of a steel sheet, particularly, local elongation. In addition, presence of sulfides (inclusions) also lowers weldability. Such an adverse effect of S is notably shown when more than 0.01% of S is contained. Therefore, the S content is limited to 0.01% or less. In order to avoid such an adverse effect of S, the S content is preferably decreased to 0.01% or less as much as possible. However, the excessive decrease leads to a rise in manufacturing cost, and thus the S content is preferably about 0.0001% or more.

Al: 0.01 to 0.1%

[0023] Al is an element that acts as a deoxidizing agent and is necessary for improving the cleanliness of a steel sheet. Furthermore, Al effectively acts to improve a yield of carbide-forming elements. 0.01% or more of Al is required to be contained in order to obtain such an effect. When less than 0.01% of Al is contained, Si-based inclusions which serve as starting points for delayed fracture are not sufficiently removed, and thus the risk of occurrence of delayed fracture is increased. However, when more than 0.1% of Al is contained, the above-described effect is saturated and thus the effect matching the content cannot be expected, resulting in economic disadvantages. In addition, formability is lowered and the tendency for generation of surface defects is increased. Therefore, the Al content is limited in the range of 0.01 to 0.1% and preferably 0.01 to 0.05%.

N: 0.005% or less

[0024] In the invention, the N content is preferably decreased as much as possible since N is an intrinsically harmful element, but up to 0.005% of N can be permitted. Therefore, the N content is limited to 0.005% or less. Since excessively decreasing the N content leads to a rise in manufacturing cost, the N content is preferably about 0.0001% or more.

[0025] The above-described components are basic components. However, in addition to the basic components, one or more selected from among 0.05 to 0.5% of Cr, 0.005 to 0.2% of V, and 0.005 to 0.2% of Mo, and/or one or two selected from between 0.01 to 0.1% of Ti and 0.01 to 0.1% of Nb, and/or 0.0003 to 0.0050% of B, and/or one or two selected from between 0.05 to 0.5% of Ni and 0.05 to 0.5% of Cu, and/or one or two selected from between 0.001 to 0.005% of Ca and 0.001 to 0.005% of REM can be contained in accordance with need.

One or More Selected from among 0.05 to 0.5% of Cr, 0.005 to 0.2% of V, and 0.005 to 0.2% of Mo

[0026] All of Cr, V, and Mo are elements that increase the strength of a steel sheet and contribute to formation of a multi phase microstructure, and one or more selected in accordance with need can be contained. In order to obtain such an effect, it is desired that 0.05% or more of Cr, 0.005% or more of V, and 0.005% or more of Mo be contained. When more than 0.5%, more than 0.2%, and more than 0.2% of Cr, V, and Mo, respectively, are contained, it is difficult to form a desired amount of pearlite in the cooling treatment after the annealing treatment, and thus a desired multi phase microstructure cannot be ensured, thereby lowering stretch flangeability and formability. Therefore, when Cr, V, and/or Mo are contained, the Cr content is preferably limited in the range of 0.05 to 0.5%, the V content is preferably limited in the range of 0.005 to 0.2%, and the Mo content is preferably limited in the range of 0.005 to 0.2%.

One or Two Selected from between 0.01 to 0.1% of Ti and 0.01 to 0.1% of Nb

[0027] Both of Ti and Nb are elements that increase the strength of a steel sheet by precipitation strengthening, and one or two selected in accordance with need can be contained. In order to obtain such an effect, it is desired that 0.01% or more of Ti and 0.01% or more of Nb be contained, respectively. When more than 0.1% of Ti and more than 0.1% of Nb are contained, formability and shape fixability are lowered. Therefore, when Ti and/or Nb is contained, the Ti content is preferably limited in the range of 0.01 to 0.1% and the Nb content is preferably limited in the range of 0.01 to 0.1%.

B: 0.0003 to 0.0050%

[0028] B is an element that segregates to austenite grain boundaries and acts to suppress formation and growth of ferrite from the grain boundaries. B can be contained in accordance with need. In order to obtain such an effect, it is desired that 0.0003% or more of B be contained. However, when more than 0.0050% of B is contained, formability is lowered. Therefore, when B is contained, the B content is preferably limited in the range of 0.0003 to 0.0050%. In addition, in order to obtain the above-described effect of B, it is necessary to suppress formation of BN, and Ti is preferably contained together with B.

One or Two Selected from between 0.05 to 0.5% of Ni and 0.05 to 0.5% of Cu

[0029] Both of Ni and Cu are elements that act to increase the strength of a steel sheet and also act to promote internal oxidation to thereby improve adhesion of the coating. Ni and Cu can be selected and contained in accordance with need. In order to obtain such an effect, it is desired that 0.05% or more of Ni and 0.05% or more of Cu be contained, respectively. However, when more than 0.5% of Ni and more than 0.5% of Cu are contained, it is difficult to form a desired amount of pearlite in the cooling treatment after the annealing treatment, and thus a desired multi phase microstructure cannot be ensured and stretch flangeability and formability are lowered. Therefore, when Ni and/or Cu is contained, the Ni content is preferably limited in the range of 0.05 to 0.5% and the Cu content is preferably limited in the range of 0.05 to 0.5%.

One or Two Selected from between 0.001 to 0.005% of Ca and 0.001 to 0.005% of REM

[0030] Both of Ca and REM are elements that contribute to controlling the form of sulfides. They act to spheroidize the form of sulfides and suppress the adverse effects of sulfides on the formability, particularly, stretch flangeability. In order to obtain such an effect, it is desired that 0.001% or more of Ca and 0.001% or more of REM be contained. However, when more than 0.005% of Ca and more than 0.005% of REM are contained, the amount of inclusions increases and surface defects and internal defects occur frequently. Therefore, when Ca and/or REM is contained, the Ca content is preferably limited in the range of 0.001 to 0.005%, and the REM content is preferably limited in the range of 0.001 to 0.005%.

[0031] The balance other than the above-described components includes Fe and inevitable impurities.

[0032] The steel sheet of the invention has the above-described composition and has a microstructure composed of a ferrite phase as a main phase and a second phase including at least pearlite.

[0033] In the steel sheet of the invention, the area fraction of the ferrite phase, i.e., the main phase, with respect to the entire microstructure is 75 to 90%. When the area fraction of the ferrite phase is lower than 75%, desired elongation and a desired hole expanding ratio cannot be obtained and formability is lowered. On the other hand, when the area fraction of the ferrite phase exceeds 90%, the area fraction of the second phase is lowered and a desired high strength cannot be obtained. Therefore, the area fraction of the ferrite phase which is the main phase is limited to the range of 75 to 90% and preferably 80 to 90%.

[0034] In the steel sheet of the invention, at least pearlite is included in the second phase. The area fraction of the pearlite is 10 to 25% with respect to the entire microstructure. When the area fraction of the pearlite is lower than 10%, a desired hole expanding ratio cannot be obtained and stretch flangeability and formability are lowered. On the other hand, when the area fraction of the pearlite exceeds 25%, the number of interfaces between the ferrite phase and the pearlite increases and voids are easily formed during the forming. Accordingly, stretch flangeability is lowered and formability is lowered.

[0035] The pearlite is fine grains having an average grain size of 5 μm or less. When the average grain size of the pearlite is large, that is, exceeding 5 μm , stress concentration occurs at the pearlite grains (interfaces) in forming the steel sheet and microvoids are formed. Accordingly, stretch flangeability is lowered and formability is lowered. Therefore, the average grain size of the pearlite is limited to 5 μm or less and preferably 4.0 μm or less.

[0036] The second phase of the microstructure of the steel sheet of the invention is a phase that includes at least pearlite and that is mainly composed of pearlite, area fraction of which is 70% or more of the total area of the second phase. When the area fraction of pearlite is less than 70% with respect to the total area of the second phase, the amount of a hard martensite or bainite phase, or retained γ becomes too large, and thus formability is easily lowered. Therefore, the area fraction of pearlite is limited to 70% or greater and preferably 75 to 100% with respect to the total area of the second phase.

[0037] The second phase may include bainite, martensite, retained austenite (retained γ) and the like, in addition to pearlite. However, particularly, since bainite and martensite are hard phases and retained γ is transformed into martensite during the forming, bainite, martensite, and retained austenite lower formability. Therefore, it is desired that the amounts of the bainite, martensite and retained austenite are as small as possible, and the area fraction of these with respect to the entire microstructure is preferably 5% or less in total and more preferably 3% or less in total.

[0038] Next, a preferred method for manufacturing the steel sheet of the invention will be described.

[0039] A steel having the above-described composition is used as a starting material. It is not necessary to particularly limit the method for manufacturing the steel. However, from the point of view of productivity, molten steel having the above-described composition is preferably refined through a general refining method using a steel converter, an electric furnace or the like, and formed into a steel such as a slab through a common casting method such as a continuous casting method. An ingot making-slabbing method, a thin-slab casting method, and the like can also be applied.

[0040] A steel having the above-described composition is hot-rolled into a hot rolled sheet. The hot rolling step preferably includes heating the steel at a temperature in the range of 1100°C to 1280°C, hot rolling the heated steel with a finish hot rolling temperature of 870°C to 950°C to form a hot rolled sheet, and, upon completion of the hot rolling, coiling the hot rolled sheet at a coiling temperature of 350°C to 720°C.

[0041] When the heating temperature of the steel is lower than 1100°C, deformation resistance becomes too high, and thus a rolling load becomes excessive and it becomes difficult to perform the hot rolling in some cases. On the other hand, when the heating temperature is higher than 1280°C, the crystal grains become too coarse, and thus a desired fine steel sheet microstructure cannot be easily obtained even when hot rolling is performed. Therefore, the heating temperature for hot rolling is preferably in the range of 1100°C to 1280°C and more preferably lower than 1280°C.

[0042] When the finish hot rolling temperature is lower than 870°C, ferrite (α) and austenite (γ) are formed during the rolling, and a banded microstructure is easily formed in the steel sheet. This banded microstructure remains even after annealing, and sometimes causes generation of anisotropy in the obtained steel sheet characteristics and lowers the formability. On the other hand, when the finish hot rolling temperature is higher than 950°C, the microstructure of the hot rolled sheet becomes coarse, and thus a desired microstructure cannot be obtained even after annealing in some cases. Therefore, the finish hot rolling temperature is preferably in the range of 870°C to 950°C.

[0043] When the coiling temperature after the hot rolling is lower than 350°C, bainitic ferrite, bainite, martensite and the like are formed and the hot rolled microstructure tends to become hard and nonuniform in grain size. In the subsequent annealing step, the microstructure tends to be nonuniform in grain size due to this hot rolled microstructure, and desired formability cannot be obtained in some cases. On the other hand, when the coiling temperature is high, that is, higher than 720°C, it becomes difficult to ensure uniform mechanical characteristics over the entire steel sheet in the longitudinal direction and in the width direction of the steel sheet. Therefore, the coiling temperature is preferably in the range of 350°C to 720°C and more preferably 500°C to 680°C.

[0044] The hot rolled sheet obtained through the hot rolling step is pickled according to a common method to remove scales on surfaces of the steel sheet, and then directly subjected to a continuous annealing step that includes an annealing treatment and a subsequent cooling treatment in a continuous annealing line without cold-rolling the hot rolled sheet.

[0045] The annealing treatment is a process in which the sheet is held in a first temperature region of an A_{c1} transformation point to an A_{c3} transformation point for 5 to 400 s.

[0046] When the temperature (heating temperature) in the first temperature region of the annealing treatment is lower than the A_{c1} transformation point or when the holding time (annealing time) in the first temperature region is shorter than 5 s, carbides in the hot rolled sheet are not sufficiently dissolved and/or a sufficient α -to- γ transformation may not occur or the α -to- γ transformation does not occur at all. Accordingly, a desired multi phase microstructure cannot be obtained by the subsequent cooling treatment, and thus a steel sheet having ductility and stretch flangeability that satisfy desired elongation and a desired hole expanding ratio cannot be obtained. On the other hand, when the heating temperature of the annealing treatment is high, that is, higher than the A_{c3} transformation point, coarsening of austenite grains is notably shown, the microstructure formed by the subsequent cooling treatment is coarsened, and the formability is thereby decreased in some cases. In addition, when the holding time (annealing time) in the first temperature region is longer than 400 s, the amount of time for the treatment is increased, the amount of consumed energy is increased, and the manufacturing cost is increased. Therefore, the annealing treatment is limited to a process in which holding is performed for 5 to 400 s in the first temperature region of the A_{c1} transformation point to the A_{c3} transformation point.

[0047] A value calculated using Expression (1) below is used as the A_{c1} transformation point of each steel sheet and a value calculated using Expression (2) below is used as the A_{c3} transformation point. When a steel sheet does not contain all elements set forth in the expressions, the contents of elements that are not contained are assumed to be zero in performing the calculations.

$$A_{c1} \text{ Transformation Point } (^{\circ}\text{C}) = 723 + 29.1\text{Si} - 10.7\text{Mn} -$$

$$16.9\text{Ni} + 16.9\text{Cr} + 6.38\text{W} + 290\text{As} \dots (1)$$

$$A_{c3} \text{ Transformation Point } (^{\circ}\text{C}) = 910 - 203\sqrt{C} + 44.7\text{Si} - 30\text{Mn} + 700\text{P} + 400\text{Al} - 15.2\text{Ni} - 11\text{Cr} - 20\text{Cu} + 31.5\text{Mo} + 104\text{V} + 400\text{Ti} + 13.1\text{W} + 120\text{As} \dots (2)$$

where C, Si, Mn, Ni, Cr, W, As, C, P, Al, Cu, Mo, V, and Ti represents contents (mass%) of the respective elements.

[0048] The cooling treatment after the annealing treatment is a process of cooling the sheet at an average cooling rate of 5°C/s or higher from the above-described first temperature region to 700°C and adjusting the residence time in a second temperature region of 700°C to 400°C in the range of 30 to 400 s.

[0049] When the average cooling rate from the first temperature region to 700°C is lower than 5°C/s, the amount of formed ferrite becomes too large. As a result, a desired multi phase microstructure is not obtained, the formability is lowered, and a desired tensile strength (540 MPa or higher) cannot be ensured in some cases. Therefore, the average cooling rate from the first temperature region to 700°C is limited to 5°C/s or higher, preferably 20°C/s or lower, and more preferably 5 to 15°C/s.

[0050] The residence time in the second temperature region of 700°C to 400°C is an important factor for the formation of pearlite included in the second phase. Here, the "residence time" means the length of time the sheet remains in the above-described second temperature region. This covers the case where the sheet is held at a specific temperature in the second temperature region, a case where the sheet is cooled in the second temperature region at a specific cooling rate, and a case where the sheet is cooled by the combination of the two cases. When the residence time in the second temperature region is shorter than 30 s, pearlite transformation does not occur or the amount of formed pearlite is insufficient, and thus a desired multi phase microstructure cannot be obtained. On the other hand, when the residence time in the second temperature region is long, that is, longer than 400 s, productivity is lowered. Therefore, the residence time in the second temperature region is limited in the range of 30 to 400 s and preferably 150 s or shorter. For securing of a desired amount of pearlite, the cooling time in a temperature region of 700°C to 550°C in the second temperature region is preferably 10 s or longer, that is, the average cooling rate in the temperature region of 700°C to 550°C is preferably 15°C/s or lower. When the cooling time in the temperature region of 700°C to 550°C is shorter than 10 s, pearlite is not sufficiently formed, a desired multi phase microstructure is not obtained, and desired formability cannot be obtained in some cases.

[0051] Hereinafter, the invention will be more specifically described on the basis of the examples. The invention is not limited to these examples.

EXAMPLES

[0052] Molten steels each having a composition shown in Table 1 were refined and formed into steels by a common method. The steels were hot-rolled at the heating temperatures and the finish hot rolling temperatures shown in Table 2 to form 1.6 mm-thick hot rolled sheets. Upon completion of the hot rolling, the hot rolled sheets were coiled at the coiling temperatures shown in Table 2. Thereafter, pickling was performed. Some of the hot rolled sheets (sheet thickness: 3.2 mm) were subjected to pickling and then to cold rolling with a rolling reduction of 50% to form 1.6 mm-thick cold rolled sheets, which were used as the comparative examples.

[0053] Under the conditions shown in Table 2, the obtained hot rolled sheets or cold rolled sheets were further subjected to a continuous annealing step that includes an annealing treatment of heating the sheets to a temperature in the first temperature region and holding the sheets thereat, and a cooling treatment of cooling the sheets at average cooling rates shown in Table 2 from the temperature in the first temperature region to 700°C, cooling the sheets at average cooling rates (cooling time) shown in Table 2 from 700°C to 550°C in a second temperature region, and adjusting the residence time in the second temperature region of 700°C to 400°C to the time shown in Table 2 so as to form annealed sheets. The transformation points of the respective steel sheets shown in Table 2 are values calculated using the above-described Expressions (1) and (2).

[0054] Test pieces were taken from the obtained annealed sheets, and observation of microstructure, a tensile test, and a hole expanding test were performed thereon. The test methods were as follows.

(1) Observation of Microstructure

[0055] A test piece for observation of microstructure was taken from an obtained annealed sheet. A cross-section (L cross-section) parallel to the rolling direction was polished and corroded with a nital solution, and the microstructure was observed in three or more fields by using a scanning electron microscope (magnification: 3000) and photographed to determine the type of the microstructure and an area fraction of each phase with respect to the entire microstructure.

Moreover, an area fraction of the total area of the second phase with respect to the entire microstructure was calculated. The average crystal grain size of the pearlite included in the second phase was also calculated. Regarding the average crystal grain size of the pearlite, an area of each pearlite grain was measured, an equivalent circle diameter was calculated from the area, the obtained equivalent circle diameters of the grains were arithmetically averaged, and the arithmetically averaged value was used as the average crystal grain size of the pearlite grains. The number of the measured pearlite grains was equal to or more than 20. An area fraction of the pearlite with respect to the total area of the second phase was also calculated.

(2) Tensile Test

[0056] A JIS No. 5 test piece was taken from an obtained annealed sheet so that the tensile direction is coincident with a direction perpendicular to the rolling direction. The tensile test was performed on the basis of the provisions of JIS Z 2241 and tensile characteristics (yield point YP, tensile strength TS, and elongation EI) were determined.

(3) Hole Expanding Test

[0057] A 100 mm-square test piece for a hole expanding test was taken from an obtained annealed sheet. The hole expanding test was performed on the basis of the Japan Iron and Steel Federation Standard JFST 1001-1996, and a hole expanding ratio λ (%) was determined.

[0058] The obtained results are shown in Table 3.

[Table 1]

Steel No.	Chemical Composition (mass%)												Remarks
	C	Si	Mn	P	S	Al	N	Cr, V, Mo	Ti, Nb	B	Ni, Cu	Ca, REM	
A	0.10	0.72	1.22	0.014	0.002	0.043	0.0038	-	-	-	-	-	Suitable Example
B	0.09	0.72	1.21	0.023	0.002	0.035	0.0028	Cr:0.35	-	-	-	-	Suitable Example
C	0.09	1.30	0.80	0.013	0.002	0.036	0.0029	-	-	-	-	-	Suitable Example
D	0.14	1.02	1.32	0.015	0.001	0.035	0.0025	-	-	-	-	-	Suitable Example
E	0.09	1.55	1.24	0.013	0.001	0.043	0.0034	-	-	-	-	-	Comparative Example
F	0.15	0.23	1.40	0.015	0.002	0.042	0.0041	-	-	-	-	-	Comparative Example
G	0.14	1.02	1.62	0.012	0.001	0.039	0.0034	-	-	-	-	-	Comparative Example
H	0.14	0.71	2.00	0.012	0.001	0.041	0.0036	-	-	-	-	-	Comparative Example
I	0.15	0.80	0.75	0.010	0.001	0.038	0.0029	Mo:0.1	-	-	-	-	Suitable Example
J	0.10	1.02	1.23	0.015	0.001	0.037	0.0039	V:0.1	-	-	-	-	Suitable Example
K	0.13	1.00	0.82	0.015	0.002	0.037	0.0040	-	Nb:0.031	-	-	-	Suitable Example
L	0.14	1.41	0.88	0.013	0.001	0.040	0.0028	-	Ti:0.022	B:0.0012	-	-	Suitable Example
M	0.14	1.02	0.84	0.015	0.001	0.037	0.0039	-	-	-	Ni:0.2	-	Suitable Example
N	0.14	0.81	0.82	0.015	0.002	0.042	0.0034	-	-	-	-	-	Suitable Example
O	0.12	1.43	0.91	0.011	0.002	0.042	0.0041	-	-	-	-	REM:0.002	Suitable Example
P	0.07	1.00	1.20	0.014	0.003	0.038	0.0035	-	-	-	-	-	Comparative Example
Q	0.16	1.01	1.24	0.013	0.002	0.043	0.0035	-	-	-	-	-	Comparative Example
R	0.13	0.40	1.32	0.012	0.001	0.036	0.0038	-	-	-	-	-	Comparative Example

[Table 2]

Steel sheet no.	Hot rolling step			Cold rolling	Sheet thickness (mm)	Continuous annealing step							Transformation point		Remarks	
	Heating temperature (°C)	Finish hot rolling temperature (°C)	Coiling temperature (°C)			Annealing treatment		Cooling treatment			Second temperature region			A _{C1} (°C)		A _{C3} (°C)
						Temperature (°C)	Holding time (s)	Average cooling rate up to 700°c* (°C/s)	Cooling time** (s)	Cooling rate** (°C/s)	Residence time*** (s)					
1	A	1200	900	600	No	1.6	800	100	10	15	10	35	731	868	Suitable Example	
2	B	1200	900	600	No	1.6	820	100	15	15	10	150	737	873	Suitable Example	
3	B	1200	900	600	No	1.6	850	100	15	15	10	300	737	873	Suitable Example	
4	B	1200	900	600	<u>Yes</u>	1.6	820	100	15	15	10	150	737	873	Comparative Example	
5	C	1200	920	560	No	1.6	820	100	10	15	10	35	752	907	Suitable Example	
6	C	1200	920	560	No	1.6	800	100	10	15	10	20	752	907	Comparative Example	
7	D	1200	900	560	No	1.6	820	100	10	15	10	150	739	865	Suitable Example	
8	D	1200	900	560	<u>Yes</u>	1.6	820	100	10	15	10	150	739	865	Comparative Example	
9	<u>E</u>	1200	920	560	No	1.6	850	100	<u>3</u>	15	10	150	755	908	Comparative Example	
10	<u>F</u>	1200	900	600	No	1.6	800	100	10	15	10	35	715	827	Comparative Example	
11	<u>F</u>	1200	900	600	Yes	1.6	800	100	10	15	10	150	715	827	Comparative Example	

(continued)

Steel sheet no.	Steel no.	Hot rolling step			Cold rolling	Sheet thickness (mm)	Continuous annealing step							Transformation point		Remarks
							Annealing treatment				Cooling treatment			A _{C-1} (°C) A _{C-3} (°C)		
		Heating temperature (°C)	Finish hot rolling temperature (°C)	Coiling temperature (°C)	Yes/No		Temperature (°C)	Holding time (s)	Average cooling rate up to 700°C* (°C/s)	Second temperature region						
										Cooling time** (s)	Cooling rate** (°C/s)	Residence time*** (s)				
12	G	1200	900	600	No	1.6	800	100	10	20	8	70	735	855	Comparative Example	
13	H	1200	900	600	No	1.6	800	100	10	20	8	70	722	831	Comparative Example	
14	I	1200	900	640	No	1.6	820	100	10	15	10	150	738	871	Suitable Example	
15	I	1200	900	640	No	1.6	820	100	10	0.2	750	0.4	738	871	Comparative Example	
16	J	1250	900	600	No	1.6	820	100	15	15	10	150	740	890	Suitable Example	
17	K	1250	900	560	No	1.6	820	150	10	30	5	35	743	882	Suitable Example	
18	K	1250	900	560	No	1.6	820	100	10	5	30	20	743	882	Comparative Example	
19	K	1250	900	560	No	1.6	715	100	10	30	5	150	743	882	Comparative Example	
20	L	1250	920	600	No	1.6	820	150	10	15	10	35	755	905	Suitable Example	
21	L	1250	920	600	No	1.6	820	150	10	15	10	150	755	905	Suitable Example	
22	L	1250	920	600	No	1.6	820	150	10	15	10	300	755	905	Suitable Example	

(continued)

Steel sheet no.	Steel no.	Hot rolling step			Cold rolling	Sheet thickness (mm)	Continuous annealing step							Transformation point		Remarks	
		Heating temperature (°C)	Finish hot rolling temperature (°C)	Coiling temperature (°C)			Yes/No	Annealing treatment	Cooling treatment			A _{c1} (°C)	A _{c3} (°C)				
									Temperature (°C)	Holding time (s)	Average cooling rate up to 700°C* (°C/s)			Second temperature region			
														Cooling time** (s)	Cooling rate** (°C/s)		Residence time*** (s)
23	L	1250	920	600	No	1.6	920	150	10	15	10	150	755	905	Comparative Example		
24	M	1200	920	600	No	1.6	840	100	15	20	8	70	740	877	Suitable Example		
25	M	1200	900	600	No	1.6	840	100	15	15	10	300	740	877	Suitable Example		
26	N	1200	900	640	No	1.6	820	100	15	15	10	150	738	873	Suitable Example		
27	O	1200	920	600	No	1.6	820	100	15	15	10	150	755	901	Suitable Example		
28	P _—	1200	920	600	No	1.7	800	100	10	15	10	35	739	890	Comparative Example		
29	P _—	1200	920	600	No	1.6	800	100	10	5 _—	30 _—	10 _—	739	890	Comparative Example		
30	Q _—	1200	900	640	No	1.6	820	100	10	15	10	70	739	863	Comparative Example		
31	R _—	1200	900	640	No	1.6	820	100	10	15	10	70	721	838	Comparative Example		

*) Cooling rate (average) from temperature in first temperature region to 700°C
**) Cooling time and average cooling rate (°C/s) between 700°C and 550°C
***) Residence time between 700°C and 400°C

*) Cooling rate (average) from temperature in first temperature region to 700°C

**) Cooling time and average cooling rate (°C/s) between 700°C and 550°C

***) Residence time between 700°C and 400°C

[Table 3]

Steel sheet no.	Steel no.	Microstructure										Tensile characteristics			Stretch-flange-ability	Remarks
		Type	Fraction of Microstructure (area fraction)						P average grain size (μm)	Yield strength YS (MPa)	Tensile strength TS (MPa)	Elongation El (%)				
			Main phase	Second Phase			Ret. γ*	Total					P fraction **			
				F*	P*	M*										
1	A	F+P+M	87.7	12.1	0.2	-	12.3	98	3.1	436	545	33.4	127	Invention Example		
2	B	F+P+M	84.5	15.3	0.2	-	15.5	99	3.0	442	552	32.1	116	Invention Example		
3	B	F+P+M	84.1	15.8	0.1	-	15.9	99	3.2	450	558	31.6	121	Invention Example		
4	B	F+P+M	87.9	<u>3.2</u>	8.9	-	12.1	<u>26</u>	2.8	458	599	31.8	67	Comparative Example		
5	C	F+P+M+Ret. γ	83.6	15.2	0.1	1.1	16.4	93	3.2	453	553	34.5	132	Invention Example		
6	C	F+M	84.8	<u>0.0</u>	15.2	-	15.2	0	-	488	724	28.2	65	Comparative Example		
7	D	F+P+M	82.6	13.2	4.2	-	17.4	76	3.4	452	624	31.7	108	Invention Example		
8	D	F+P+M	82.2	10.6	7.2	-	17.8	<u>60</u>	3.2	448	625	32.4	69	Comparative Example		
9	<u>E</u>	F+P+M	86.7	<u>1.2</u>	12.1	-	13.3	9	2.1	387	624	33.6	58	Comparative Example		
10	<u>F</u>	F+P+M	88.8	10.6	0.6	-	11.2	95	2.9	425	518	34.2	85	Comparative Example		
11	<u>F</u>	F+P+M	88.2	10.9	0.9	-	11.8	92	3.2	429	526	33.7	83	Comparative Example		
12	<u>G</u>	F+M	85.9	<u>0.0</u>	14.1	-	14.1	0	-	425	638	30.2	56	Comparative Example		

(continued)

Steel sheet no.	Steel no.	Microstructure										Tensile characteristics			Stretch-flange-ability		Remarks
		Type	Fraction of Microstructure (area fraction)						P average grain size (μm)	Yield strength YS (MPa)	Tensile strength TS (MPa)	Elongation El (%)	λ (%)				
			Main phase	Second Phase													
				P*	M*	Ret. γ*	Total	P fraction **									
			F*														
13		F+M	83.6	0.0	16.4	-	16.4	0	-	463	672	28.9	51		Comparative Example		
14		F+P+M	85.0	14.8	0.2	-	15.0	99	2.9	449	564	33.1	114		Invention Example		
15		F+M	54.8	0.0	45.2	-	45.2	0	-	653	1082	13.8	32		Comparative Example		
16		F+P+M	83.6	16.2	0.2	-	16.4	99	3.7	462	592	32.4	118		Invention Example		
17		F+P+M	83.4	15.1	1.5	-	16.6	91	3.4	472	611	31.8	111		Invention Example		
18		F+M	84.8	0.0	15.2	-	15.2	0	-	502	752	24.7	54		Comparative Example		
19		F+C	90.4	0.0	-	-	0.0	0	-	402	535	31.1	79		Comparative Example		
20		F+P+M+Retained γ	85.9	11.5	1.4	1.2	14.1	82	3.8	453	603	32.5	105		Invention Example		
21		F+P+M+Ret. γ	85.3	12.4	1.2	1.1	12.3	84	3.9	468	601	32.6	111		Invention Example		
22		F+P+M	86.5	12.2	1.3	-	13.5	90	3.9	462	597	32.4	104		Invention Example		
23		F+P+M	83.0	15.8	1.2	-	17.0	93	6.5	488	625	31.1	75		Comparative Example		
24		F+P+M	87.3	12.5	0.2	-	12.7	98	3.2	482	614	31.5	117		Invention Example		

(continued)

Steel sheet no.	Steel sheet no.	Microstructure										Tensile characteristics			Stretch-flange-ability	Remarks
		Type	Fraction of Microstructure (area fraction)						P average grain size (μm)	Yield strength YS (MPa)	Tensile strength TS (MPa)	Elongation El (%)	λ (%)			
			Main phase	Second Phase				P fraction **								
				F*	P*	M*	Ret. γ*							Total		
25	M	F+P	86.1	13.9	-	-	-	13.9	100	3.4	469	586	32.0	120	Invention Example	
26	N	F+P+M	86.8	13.1	0.1	-	-	13.2	99	3.0	442	564	32.9	118	Invention Example	
27	O	F+P+M+Ret. γ	86.0	12.4	0.2	1.4	14.0	89	89	3.2	467	608	31.8	109	Invention Example	
28	P _—	F+P+M	91.7	8.2	0.1	-	-	8.3	99	3.1	385	482	35.1	105	Comparative Example	
29	P _—	F+P+M	87.8	5.8	6.4	-	-	12.2	48	2.2	425	572	32.4	62	Comparative Example	
30	Q _—	F+P+M	90.9	4.9	4.2	-	-	9.1	54	2.2	524	694	26.8	69	Comparative Example	
31	R _—	F+P	89.7	10.3	-	-	-	10.3	100	3.1	409	528	32.9	110	Comparative Example	
*) F: ferrite, P: pearlite, M: martensite, Ret. γ: retained austenite, C: cementite																
**) P fraction: P area/second phase total area ratio																

[0059] In all of the invention examples, high strength steel sheets having excellent formability with a high strength, i.e., a tensile strength TS of 540 MPa or higher, high ductility, i.e., elongation El of 30% or greater, and excellent stretch flangeability, i.e., hole expanding ratio λ of 80% or higher are obtained. In contrast, in the comparative examples outside the range of the invention, a desired high strength is not obtained, desired elongation is not obtained, or a desired hole expanding ratio λ is not obtained, and thus formability is lowered.

Claims

1. A high strength steel sheet having excellent formability which has a composition consisting of, by mass%:

0.08 to 0.15% of C, 0.5 to 1.5% of Si, 0.5 to 1.5% of Mn, 0.1% or less of P, 0.01% or less of S, 0.01 to 0.1% of Al, 0.005% or less of N,
and optionally one or more of 0.05 to 0.5% of Cr, 0.005 to 0.2% of V, 0.005 to 0.2% of Mo, 0.01 to 0.1% of Ti, 0.01 to 0.1% of Nb, 0.0003 to 0.0050% of B, 0.05 to 0.5% of Ni, 0.05 to 0.5% of Cu, 0.001 to 0.005% of Ca, 0.001 to 0.005% of REM,
and the balance Fe with inevitable impurities,
the steel sheet having a microstructure composed of a ferrite phase which is a main phase and a second phase including at least pearlite, wherein an area fraction of the ferrite phase is in the range of 75 to 90% and an area fraction of the pearlite is in the range of 10 to 25% with respect to the entire microstructure, an average grain size of the pearlite is 5 μ m or smaller, and an area fraction of the pearlite is 70% or greater with respect to the total area of the second phase.

2. A method for manufacturing a high strength steel sheet having excellent formability, comprising:

a hot rolling step of hot-rolling a steel having a composition consisting of, by mass%,
0.08 to 0.15% of C, 0.5 to 1.5% of Si, 0.5 to 1.5% of Mn, 0.1% or less of P, 0.01% or less of S, 0.01 to 0.1% of Al, 0.005% or less of N,
and optionally one or more of 0.05 to 0.5% of Cr, 0.005 to 0.2% of V, 0.005 to 0.2% of Mo, 0.01 to 0.1% of Ti, 0.01 to 0.1% of Nb, 0.0003 to 0.0050% of B, 0.05 to 0.5% of Ni, 0.05 to 0.5% of Cu, 0.001 to 0.005% of Ca, 0.001 to 0.005% of REM,
and the balance Fe with inevitable impurities to form a hot rolled sheet;
and a continuous annealing step including an annealing treatment of pickling the hot rolled sheet, and holding the pickled hot rolled sheet in a first temperature region of an A_{c1} transformation point to an A_{c3} transformation point for 5 to 400 s,
and a cooling treatment of cooling the sheet at an average cooling rate of 5 °C/s or higher from the first temperature region to 700°C after the annealing treatment and adjusting a residence time in a second temperature region of 700°C to 400°C in the range of 30 to 400 s by using a continuous annealing line, wherein a cooling time in a temperature region of 700°C to 550°C in the second temperature region is 10 s or longer.

3. The method for manufacturing a high strength steel sheet according to Claim 2, wherein the hot rolling step includes heating the steel at a temperature in the range of 1100°C to 1280°C, hot-rolling the heated steel with a finish hot rolling temperature of 870°C to 950°C to form a hot rolled sheet, and coiling the hot rolled sheet at a coiling temperature of 350°C to 720°C upon completion of the hot rolling.

Patentansprüche

1. Hochfestes Stahlblech mit ausgezeichneter Verformbarkeit, das eine Zusammensetzung aufweist, die in Ma% aus Folgendem besteht:

0,08 bis 0,15 % C, 0,5 bis 1,5 % Si, 0,5 bis 1,5 % Mn, 0,1 % oder weniger P, 0,01 % oder weniger S, 0,01 bis 0,1 % Al, 0,005 % oder weniger N,
und optional einem oder mehreren aus 0,05 bis 0,5 % Cr, 0,005 bis 0,2 % V, 0,005 bis 0,2 % Mo, 0,01 bis 0,1 % Ti, 0,01 bis 0,1 % Nb, 0,0003 bis 0,0050 % B, 0,05 bis 0,5 % Ni, 0,05 bis 0,5 % Cu, 0,001 bis 0,005 % Ca, 0,001 bis 0,005% REM,
wobei der Rest Fe mit unvermeidbaren Verunreinigungen ist,
wobei das Stahlblech eine Mikrostruktur aufweist, die aus einer Ferritphase besteht, bei der es sich um eine

Hauptphase handelt, und einer zweiten Phase, die mindestens Perlit enthält, wobei ein Flächenanteil der Ferritphase im Bereich von 75 bis 90 % liegt und ein Flächenanteil des Perlits im Bereich von 10 bis 25 % bezüglich der gesamten Mikrostruktur liegt, eine durchschnittliche Korngröße des Perlits 5 μm oder kleiner ist, und einen Flächenanteil des Perlits 70 % oder größer bezüglich der Gesamtfläche der zweiten Phase ist.

2. Verfahren zur Herstellung eines hochfesten Stahlblechs mit ausgezeichneter Verformbarkeit, umfassend folgende Schritte:

einen Warmwalzschritt zum Warmwalzen eines Stahls mit einer Zusammensetzung, die in Ma% aus Folgendem besteht:

0,08 bis 0,15 % C, 0,5 bis 1,5 % Si, 0,5 bis 1,5 % Mn, 0,1 % oder weniger P, 0,01 % oder weniger S, 0,01 bis 0,1 % Al, 0,005 % oder weniger N,

und optional einem oder mehreren aus 0,05 bis 0,5 % Cr, 0,005 bis 0,2 % V, 0,005 bis 0,2 % Mo, 0,01 bis 0,1 % Ti, 0,01 bis 0,1 % Nb, 0,0003 bis 0,0050 % B, 0,05 bis 0,5 % Ni, 0,05 bis 0,5 % Cu, 0,001 bis 0,005 % Ca, 0,001 bis 0,005 % REM,

und wobei der Rest Fe mit unvermeidbaren Verunreinigungen ist, um ein warmgewalztes Blech zu erzeugen; und einen Schritt zum Durchlaufglühen einschließlich einer Glühbehandlung zum Beizen des warmgewalzten Blechs und Halten des gebeizten warmgewalzten Blechs in einem ersten Temperaturbereich von einem A_{c1} -Transformationspunkt bis zu einem A_{c3} -Transformationspunkt für 5 bis 400 s,

und eine Kühlbehandlung zum Abkühlen des Blechs bei einer durchschnittlichen Abkühlungsgeschwindigkeit von 5 °C/s oder höher von dem ersten Temperaturbereich bis auf 700 °C nach der Glühbehandlung und Einstellen einer Verweilzeit in einem zweiten Temperaturbereich von 700 °C bis 400 °C im Bereich von 30 bis 400 s unter Verwendung einer Durchlaufglühanlage, wobei die Abkühlzeit in einem Temperaturbereich von 700 °C bis 550 °C im zweiten Temperaturbereich 10 s oder länger beträgt.

3. Verfahren zur Herstellung eines hochfesten Stahlblechs nach Anspruch 2, wobei der Warmwalzschritt das Erwärmen des Stahls auf eine Temperatur im Bereich von 1100 °C bis 1280 °C, das Warmwalzen des erwärmten Stahls mit einer Schluss-Warmwalztemperatur von 870 °C bis 950 °C zum Bilden eines warmgewalzten Blechs, und das Aufwickeln des warmgewalzten Blechs bei einer Aufwicklungstemperatur von 350 °C bis 720 °C nach Abschluss des Warmwalzens beinhaltet.

Revendications

1. Tôle d'acier à résistance élevée ayant une excellente aptitude au formage qui présente une composition constituée, en % en masse, de :

0,08 à 0,15 % de C, 0,5 à 1,5 % de Si, 0,5 à 1,5 % de Mn, 0,1 % ou moins de P, 0,01 % ou moins de S, 0,01 à 0,1 % d'Al, 0,005 % ou moins de N,

et éventuellement d'un ou plusieurs parmi 0,05 à 0,5 % de Cr, 0,005 à 0,2 % de V, 0,005 à 0,2 % de Mo, 0,01 à 0,1 % de Ti, 0,01 à 0,1 % de Nb, 0,0003 à 0,0050 % de B, 0,05 à 0,5 % de Ni, 0,05 à 0,5 % de Cu, 0,001 à 0,005 % de Ca, 0,001 à 0,005 % de terres rares,

le reste étant du fer avec des impuretés inévitables,

la tôle d'acier ayant une microstructure composée d'une phase ferrite qui est une phase principale et d'une seconde phase comprenant au moins de la perlite, où une fraction surfacique de la phase ferrite est dans la plage de 75 à 90 % et une fraction surfacique de la perlite est dans la plage de 10 à 25 % par rapport à la microstructure entière, une taille de grain moyenne de la perlite est inférieure ou égale à 5 μm et une fraction surfacique de la perlite est supérieure ou égale à 70 % par rapport à la surface totale de la seconde phase.

2. Procédé de fabrication d'une tôle d'acier à résistance élevée ayant une excellente aptitude au formage, comprenant :

une étape de laminage à chaud consistant à laminier à chaud de l'acier ayant une composition constituée, en % en masse, de

0,08 à 0,15 % de C, 0,5 à 1,5 % de Si, 0,5 à 1,5 % de Mn, 0,1 % ou moins de P, 0,01 % ou moins de S, 0,01 à 0,1 % d'Al, 0,005 % ou moins de N,

et éventuellement d'un ou plusieurs parmi 0,05 à 0,5 % de Cr, 0,005 à 0,2 % de V, 0,005 à 0,2 % de Mo, 0,01 à 0,1 % de Ti, 0,01 à 0,1 % de Nb, 0,0003 à 0,0050 % de B, 0,05 à 0,5 % de Ni, 0,05 à 0,5 % de Cu, 0,001 à 0,005 % de Ca, 0,001 à 0,005 % de terres rares,

le reste étant du fer avec des impuretés inévitables pour former une tôle laminée à chaud ;
et une étape de recuit continu incluant un traitement de recuit consistant à décapier la tôle laminée à chaud, et
maintenir la tôle laminée à chaud décapée dans une première région de température allant d'un point de
transformation A_{c1} à un point de transformation A_{c3} pendant 5 à 400 s,
et un traitement de refroidissement consistant à refroidir la tôle à une vitesse de refroidissement moyenne de
5 °C/s ou plus depuis la première région de température jusqu'à 700 °C après le traitement de recuit et ajuster
un temps de séjour dans une seconde région de température de 700 °C à 400 °C dans la plage de 30 à 400 s
en utilisant une ligne de recuit continu, une durée de refroidissement dans une région de température de 700
°C à 550 °C dans la seconde région de température étant de 10 s ou plus.

- 3.** Procédé de fabrication d'une tôle d'acier à résistance élevée selon la revendication 2, dans lequel l'étape de laminage
à chaud comprend le chauffage de l'acier à une température dans la plage de 1 100 °C à 1 280 °C, le laminage à
chaud de l'acier chauffé avec une température de finissage de laminage à chaud de 870 °C à 950 °C pour former
une tôle d'acier laminée à chaud et l'enroulement de la tôle d'acier laminée à chaud à une température d'enroulement
de 350 °C à 720 °C à la fin du laminage à chaud.

REFERENCES CITED IN THE DESCRIPTION

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