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(54) **HIGH-STRENGTH STEEL HAVING EXCELLENT BRITTLE CRACK ARRESTABILITY AND WELDING PART BRITTLE CRACK INITIATION RESISTANCE, AND PRODUCTION METHOD THEREFOR**

(57) The purpose of another aspect of the present invention is to provide a high-strength steel having excellent brittle crack arrestability and welding part brittle crack initiation resistance, and a production method therefor. According to one aspect of the present invention, provided are a high-strength steel having excellent brittle crack arrestability and welding part brittle crack initiation resistance, and a production method therefor, the high-strength steel: comprising, in wt%, C: 0.05-0.09%, Mn: 1.5-2.2%, Ni: 0.3-1.2%, Nb: 0.005-0.04%, Ti: 0.005-0.004%, Cu: 0.1-0.8%, Si: 0.05-0.03%, Al: 0.005-0.05%, P: 100ppm or less, S: 40ppm or less, and a remainder made up by Fe and other inevitable impurities; having a center part microstructure comprising an acicular ferrite and granular bainite mixed-phase, upper bainite, and a remainder made up by one type or more selected from the group consisting of ferrite, pearlite, and a martensite-austenite (MA) constituent; having, in a 2mm or less subsurface region, a surface part microstructure comprising ferrite and a remainder made up by one type or more among bainite and martensite, and having a welding heat affected zone, which is formed during welding, that comprises, in area%, 5% or less of a martensite-austenite constituent. According to the present invention, high-strength steel having high yield strength, excellent brittle crack arrestability, and excellent welding part brittle crack initiation resistance may be obtained.

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Description

[Technical Field]

[0001] The present disclosure relates to a high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance, and to a method of manufacturing the same.

[Background Art]

[0002] Recently, there has been demand for the development of ultra-thick steel sheets having high strength properties in consideration of the design requirements of structures to be used in the shipping, maritime, architectural, and civil engineering fields, domestically and internationally.

[0003] In a case in which high-strength steel is included in the design of a structure, economic benefits may be obtained due to reductions in the weight of structures while processing and welding operations may be easily undertaken using a steel sheet having a relatively reduced thickness.

[0004] In general, in the case of high-strength steel, due to a reduction in a reduction ratio when thick steel plates are manufactured, sufficient deformation is not performed, as compared with a thin steel sheet. Thus, microstructures of thick steel plates may be coarse, so that low temperature properties on which grain sizes have the most significant effect may be degraded.

[0005] In detail, in a case in which brittle crack arrestability representing stability of a structure is applied to a main structure, such as a ship's hull, the number of cases of demanding assurances has been increased. However, in a case in which microstructures become coarse, a phenomenon in which brittle crack arrestability is significantly degraded may occur. Thus, it is difficult to improve brittle crack arrestability of an ultra-thick high-strength steel material.

[0006] In the meantime, in the case of high-strength steel having yield strength of 460 MPa or greater, various technologies, such as adjustment of grain size, by applying surface cooling during finishing milling and applying bending stress during rolling to refine the grain size of a surface portion, in order to improve brittle crack arrestability, have been introduced.

[0007] However, such technologies may contribute to refining a structure of a surface portion, but may not solve a problem in which impact toughness is degraded due to coarsening of structures other than the surface portion. Thus, such technologies may not be fundamental countermeasures to brittle crack arrestability.

[0008] In addition, recently, the design concept to improve safety of a ship by controlling brittle crack initiation of a steel material applied to large container ships has been introduced. Thus, in general, the number of cases of guaranteeing brittle crack initiation of a heat affected zone (HAZ), the most vulnerable portion in terms of brittle crack initiation, has been increased.

[0009] In general, since, in the case of high-strength steel, the microstructure in a HAZ includes low temperature transformation ferrite having high strength, such as bainite, there is a limitation in which HAZ properties, in detail, toughness, is significantly reduced.

[0010] In detail, in the case of brittle crack initiation resistance generally evaluated through a crack tip opening displacement (CTOD) test to evaluate the stability of the structure, martensite-austenite generated from untransformed austenite, when low temperature transformation ferrite is generated, becomes an active nucleation site of brittle crack occurrence. Thus, it is difficult to improve brittle crack initiation resistance of a high-strength steel material.

[0011] In the case of high-strength steel of the related art having a yield strength of 460 MPa or greater, in order to improve welding zone brittle crack initiation resistance, an effort to refine a microstructure in a HAZ using TiN or to form ferrite in a HAZ using oxide metallurgy has been made. However, the effort partially contributes to forming impact toughness through refining a structure, but does not have a great effect on reducing a fraction of martensite-austenite having a significant influence on reducing brittle crack initiation resistance.

[0012] In addition, in the case of brittle crack initiation resistance of a base material, martensite-austenite may be transformed to have a different phase through tempering, or the like, to secure physical properties. However, in the case of the HAZ in which an effect of tempering disappears due to thermal history, it is impossible to apply brittle crack initiation resistance.

[0013] In the meantime, in order to minimize the formation of martensite-austenite, the amount of elements, such as carbon (C) and niobium (Nb), should be reduced. However, in this case, it may be difficult to secure a specific level of strength. To this end, a relatively large amount of high-priced elements, such as molybdenum (Mo) and nickel (Ni), should be added. Thus, there is a limitation in which economic efficiency is deteriorated.

[Disclosure]

[Technical Problem]

[0014] An aspect of the present disclosure may provide a high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance.

[0015] Another aspect of the present disclosure may provide a method of manufacturing a high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance.

[Technical Solution]

[0016] According to an aspect of the present disclosure, a high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance comprises, by wt%, carbon (C): 0.05% to 0.09%, manganese (Mn): 1.5% to 2.2%, nickel (Ni): 0.3% to 1.2%, niobium (Nb): 0.005% to 0.04%, titanium (Ti): 0.005% to 0.04%, copper (Cu): 0.1% to 0.8%, silicon (Si): 0.05% to 0.3%, aluminum (Al): 0.005% to 0.05%, phosphorus (P): 100 ppm or less, sulfur (S): 40 ppm or less, iron (Fe) as a residual component thereof, and inevitable impurities, wherein a microstructure of a central portion includes, by area%, a mixed phase of acicular ferrite and granular bainite in an amount of 70% or greater, upper bainite in an amount of 20% or less, and one or more selected from a group consisting of ferrite, pearlite, and martensite-austenite (MA), as residual components; a circle-equivalent diameter of an effective grain of the upper bainite having a high angle grain boundary of 15° or greater measured using an electron backscatter diffraction (EBSD) method being 15 μm or less; a surface portion microstructure in a region at a depth of 2 mm or less, directly below a surface, includes, by area%, ferrite in an amount of 20% or greater and one or more of bainite and martensite as residual components; and a heat affected zone (HAZ) formed during welding includes, by area%, martensite-austenite (MA) in an amount of 5% or less.

[0017] Contents of Cu and Ni may be set such that a weight ratio of Cu to Ni may be 0.8 or less, and in more detail, 0.6 or less.

[0018] The high-strength steel material may have yield strength of 460 MPa or greater.

[0019] The high-strength steel material may have a Charpy fracture transition temperature of -40°C or lower in a 1/2t position in a steel material thickness direction, where t is a steel sheet thickness.

[0020] According to another aspect of the present disclosure, a method of manufacturing a high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance comprises rough rolling a slab at a temperature of 900°C to 1100°C after reheating the slab at 1000°C to 1100°C, including, by wt%, C: 0.05% to 0.09%, Mn: 1.5% to 2.2%, Ni: 0.3% to 1.2%, Nb: 0.005% to 0.04%, titanium (Ti) : 0.005% to 0.04%, copper (Cu): 0.1% to 0.8%, silicon (Si): 0.05% to 0.3%, aluminum (Al): 0.005% to 0.05%, phosphorus (P): 100 ppm or less, sulfur (S): 40 ppm or less, iron (Fe) as a residual component thereof, and inevitable impurities; obtaining a steel sheet by finish rolling a bar obtained from the rough rolling a slab, at a temperature in a range of Ar_3 + 60°C to Ar_3 °C, based on a temperature of a central portion; and cooling the steel sheet to 500°C or lower.

[0021] A reduction ratio per pass of three final passes during the rough rolling a slab may be 5% or greater, and a total cumulative reduction ratio may be 40% or greater.

[0022] A strain rate of three final passes during the rough rolling a slab may be 2/sec or lower.

[0023] A grain size of a central portion in a bar thickness direction before finish rolling after the rough rolling a slab may be 150 μm or less, in detail, 100 μm or less, and more specifically, 80 μm or less.

[0024] A reduction ratio during the finish rolling may be set such that a ratio of a slab thickness (mm) to a steel sheet thickness (mm) after the finish rolling may be 3.5 or greater, and in more detail, 4 or greater.

[0025] A cumulative reduction ratio during the finish rolling may be maintained to be 40% or greater, while the reduction ratio per pass, not including skin pass rolling, may be maintained to be 4% or greater. Skin pass rolling refers to a process of rolling a sheet at a relatively low reduction ratio in order to secure flatness of the sheet.

[0026] The cooling the steel sheet may be performed at a cooling rate of the central portion of 2°C/s or higher.

[0027] The cooling the steel sheet may be performed at an average cooling rate of 3°C/s to 300°C/s.

[0028] In addition, the present inventive concept may be exemplified in many different forms and should not be construed as being limited to the specific embodiments set forth herein. Rather, these embodiments are provided so that this disclosure will be thorough and complete, and will fully convey the scope of the disclosure to those skilled in the art.

[Advantageous Effects]

[0029] According to an aspect of the present disclosure, a high-strength steel material having a relatively high level of yield strength, as well as excellent brittle crack arrestability and welding zone brittle crack initiation resistance.

[Best Mode for Invention]

[0030] The inventors of the present disclosure conducted research and experiments to improve yield strength, brittle crack arrestability, and welding zone brittle crack initiation resistance of a thick steel material and proposed the present disclosure based on results thereof.
[0031] In an exemplary embodiment, a steel composition, a structure, and manufacturing conditions of a steel material may be controlled, thereby improving yield strength, brittle crack arrestability, and welding zone brittle crack initiation resistance of the thick steel material.
[0032] A main concept of an exemplary embodiment is as follows.

1) The steel composition is appropriately controlled to improve strength through solid solution strengthening. In detail, contents of manganese (Mn), nickel (Ni), copper (Cu), and silicon (Si) are optimized for solid solution strengthening.
2) The steel composition is appropriately controlled to improve strength by increasing hardenability. In detail, the contents of Mn, Ni, and Cu, as well as a carbon (C) content are optimized to increase hardenability.
A fine structure is secured in a central portion of the thick steel material even at a relatively slow cooling rate.
3) A composition is appropriately controlled to control a fraction of martensite-austenite. In detail, contents of C, Si, and niobium (Nb), affecting generation of martensite-austenite, are optimized.
As such, the steel composition may be optimized, thereby securing excellent brittle crack initiation resistance even in a heat affected zone (HAZ).
4) More specifically, a structure of the steel material may be controlled to improve strength and brittle crack arrestability. In detail, a structure of the central portion and a surface layer region is controlled in a direction of a steel material thickness.
As such, a microstructure may be controlled, thereby securing strength required in the steel material, while the microstructure facilitating generation of a crack may be excluded, thereby improving brittle crack arrestability.
5) In detail, rough rolling conditions may be controlled to refine the structure of the steel material.
In detail, the fine structure is secured in the central portion by controlling a rolling condition during rough rolling. Using a process described above, the generation of acicular ferrite and granular bainite is facilitated.
6) A finish rolling condition is controlled to further refine the structure of the steel material. In detail, a finish rolling temperature and rolling conditions may be controlled to generate a relatively large amount of strain bands in austenite during finish rolling and secure a large number of ferrite nucleation sites, thereby securing a fine structure in the central portion of the steel material. As such, generation of acicular ferrite and granular bainite is facilitated.

[0033] Hereinafter, the high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance according to an aspect of the present disclosure will be described in detail.
[0034] According to an aspect of the present disclosure, the high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance comprises, by wt%, carbon (C): 0.05% to 0.09%, manganese (Mn): 1.5% to 2.2%, nickel (Ni): 0.3% to 1.2%, niobium (Nb): 0.005% to 0.04%, titanium (Ti): 0.005% to 0.04%, copper (Cu): 0.1% to 0.8%, silicon (Si): 0.05% to 0.3%, aluminum (Al) : 0.005% to 0.05%, phosphorus (P): 100 ppm or less, sulfur (S) : 40 ppm or less, iron (Fe) as a residual component thereof, and inevitable impurities, wherein a microstructure of a central portion includes, by area%, a mixed phase of acicular ferrite and granular bainite in an amount of 70% or greater, upper bainite in an amount of 20% or less, and one or more selected from a group consisting of ferrite, pearlite, and martensite-austenite (MA), as residual components; a circle-equivalent diameter of an effective grain of the upper bainite having a high angle grain boundary of 15° or greater measured using an electron backscatter diffraction (EBSD) method being 15 μm or less; a surface portion microstructure in a region at a depth of 2 mm or less, directly below a surface, includes, by area%, ferrite in an amount of 20% or greater and one or more of bainite and martensite as residual components; and a heat affected zone (HAZ) formed during welding includes, by area%, martensite-austenite (MA) in an amount of 5% or less.
[0035] Hereinafter, a steel component and a component range of an exemplary embodiment will be described.

Carbon (C): 0.05 wt% to 0.09 wt% (hereinafter, referred to as "%")

[0036] Since C is the most significant element used in securing basic strength, C is required to be contained in steel within an appropriate range. In order to obtain an effect of addition, C may be added in an amount of 0.05% or greater.
[0037] However, in a case in which a C content exceeds 0.09%, a large amount of martensite-austenite is generated in the HAZ to degrade brittle crack initiation resistance. Low temperature toughness is degraded due to a relatively high level of strength of ferrite of a base material and generation of a relatively large amount of low temperature transformation ferrite. Thus, the C content may be limited to 0.05% to 0.09%.

[0038] In detail, the C content may be limited to 0.055% to 0.08%, and more specifically, to 0.06% to 0.075%.

Manganese (Mn): 1.5% to 2.2%

[0039] Mn is a useful element improving strength through solid solution strengthening and increasing hardenability to generate low temperature transformation ferrite. In addition, since Mn may generate low temperature transformation ferrite even at a relatively low cooling rate due to improved hardenability, Mn is a main element to secure strength of a central portion of a thick steel plate.
[0040] Therefore, in order to obtain an effect described above, Mn may be added in an amount of 1.5% or greater.
[0041] However, in a case in which a Mn content exceeds 2.2%, generation of upper bainite and martensite may be facilitated due to an increase in excessive hardenability, thereby degrading impact toughness and brittle crack arrestability and toughness of the HAZ.
[0042] Therefore, the Mn content may be limited to 1.5% to 2.2%.
[0043] In detail, the Mn content may be limited to 1.6% to 2.0%, and more specifically, to 1.65% to 1.95%.

Nickel (Ni): 0.3% to 1.2%

[0044] Ni is a significant element used in improving impact toughness by facilitating a dislocation cross slip at a relatively low temperature and increasing strength by improving hardenability. In order to obtain an effect described above, Ni may be added in an amount of 0.3% or greater. However, in a case in which Ni is added in an amount of 1.2% or greater, hardenability is excessively increased to generate low temperature transformation ferrite, thereby degrading toughness, and a manufacturing cost may be increased due to a relatively high cost of Ni, as compared with other hardenability elements. Thus, an upper limit value of the Ni content may be limited to 1.2%.
[0045] In detail, the Ni content may be limited to 0.4% to 1.0%, and more specifically, to 0.45% to 0.9%.

Niobium (Nb): 0.005% to 0.04%

[0046] Nb is educed to have a form of NbC or NbCN to improve strength of a base material.
[0047] In addition, Nb solidified when being reheated at a relatively high temperature is significantly finely educed to have the form of NbC during rolling to suppress recrystallization of austenite, thereby having an effect of refining a structure.
[0048] Therefore, Nb may be added in an amount of 0.005% or greater. However, in a case in which Nb is added excessively, generation of martensite-austenite in the HAZ may be facilitated to degrade brittle crack initiation resistance and cause a brittle crack in an edge of the steel material. Thus, an upper limit value of an Nb content may be limited to 0.04%.
[0049] In detail, the Nb content may be limited to 0.01% to 0.035%, and more specifically, to 0.015% to 0.03%.

Titanium (Ti): 0.005% to 0.04%

[0050] Ti is a component educed to be TiN when being reheated and inhibiting growth of the base material and a grain in the HAZ to greatly improve low temperature toughness. In order to obtain an effect of addition, Ti may be added in an amount of 0.005% or greater.
[0051] However, in a case in which Ti is added excessively, low temperature toughness may be degraded due to clogging of a continuous casting nozzle or crystallization of the central portion. Thus, a Ti content may be limited to 0.005% to 0.04%.
[0052] In detail, the Ti content may be limited to 0.008% to 0.03%, and more specifically, to 0.01% to 0.02%.

Silicon (Si): 0.05% to 0.3%

[0053] Si is a substitutional element improving strength of the steel material through solid solution strengthening and having a strong deoxidation effect, so that Si may be an element essential in manufacturing clean steel. Thus, Si may be added in an amount of 0.05% or greater. However, when a relatively large amount of Si is added, a coarse martensite-austenite phase may be formed to degrade brittle crack arrestability and welding zone brittle crack initiation resistance. Thus, an upper limit value of an Si content may be limited to 0.3%.
[0054] In detail, the Si content may be limited to 0.1% to 0.25%, and more specifically, to 0.1% to 0.2%.

Copper (Cu): 0.1% to 0.8%

[0055] Cu is a main element used in improving hardenability and causing solid solution strengthening to enhance strength of the steel material. In addition, Cu is a main element used in increasing yield strength through the generation of an epsilon Cu precipitate when tempering is applied. Thus, Cu may be added in an amount of 0.1% or greater. However, when a relatively large amount of Cu is added, a slab crack may be generated by hot shortness in a steelmaking process. Thus, an upper limit value of a Cu content may be limited to 0.8%.
[0056] In detail, the Cu content may be limited to 0.2% to 0.6%, and more specifically, to 0.25% to 0.5%.
[0057] Contents of Cu and Ni may be set such that the weight ratio of Cu to Ni may be 0.8 or less, and in more detail, 0.6 or less. More specifically, the weight ratio of Cu to Ni may be limited to 0.5 or less.
[0058] In a case in which the weight ratio of Cu to Ni is set as described above, surface quality may be improved.

Aluminum (Al): 0.005% to 0.05%

[0059] Al is a component functioning as a deoxidizer. In a case in which an excessive amount of Al is added, an inclusion may be formed to degrade toughness. Thus, an Al content may be limited to 0.005% to 0.05%.

Phosphorus (P): 100 ppm or less, Sulfur (S): 40 ppm or less

[0060] P and S are elements causing brittleness in a grain boundary or forming a coarse inclusion to cause brittleness. In order to improve brittle crack arrestability, a P content may be limited to 100 ppm or less, while an S content may be limited to 40 ppm or less.
[0061] A residual component of an exemplary embodiment is iron (Fe) .
[0062] However, since, in a manufacturing process of the related art, unintended impurities may be inevitably mixed from a raw material or an external source, which may not be excluded.
[0063] Since the impurities are apparent to those skilled in the art, all the contents thereof are not specifically described in the present disclosure.
[0064] In the case of a steel material of an exemplary embodiment, a microstructure of a central portion includes, by area%, a mixed phase of acicular ferrite and granular bainite in an amount of 70% or greater, upper bainite in an amount of 20% or less, and one or more selected from a group consisting of ferrite, pearlite, and martensite-austenite (MA), as residual components; a circle-equivalent diameter of an effective grain of the upper bainite having a high angle grain boundary of 15° or greater measured using an electron backscatter diffraction (EBSD) method being 15 μm or less; a microstructure in a region at a depth of 2 mm or less, directly below a surface, includes, by area%, ferrite in an amount of 20% or greater and one or more of bainite and martensite as residual components; and a heat affected zone (HAZ) formed during welding includes, by area%, martensite-austenite (MA) in an amount of 5% or less.
[0065] In a case in which a fraction of the mixed phase of acicular ferrite and granular bainite of the microstructure of the central portion is less than 70%, sufficient yield strength may be difficult to secure. For example, yield strength of 460 MPa or greater may be difficult to secure.
[0066] In detail, the fraction of the mixed phase of acicular ferrite and granular bainite may be 75% or greater, and more specifically, may be limited to 80% or greater.
[0067] A fraction of acicular ferrite may be 20% to 70%.
[0068] In a case in which the fraction of acicular ferrite exceeds 70%, sufficient yield strength may be difficult to secure due to a reduction in strength. For example, a yield strength of 460 MPa or greater may be difficult to secure. In a case in which yield strength is less than 20%, impact toughness may be degraded due to a relatively high level of strength.
[0069] In detail, the fraction of acicular ferrite may be limited to 30% to 50%, and more specifically, to 30% to 40%.
[0070] A fraction of granular bainite may be 10% to 60%.
[0071] In a case in which the fraction of granular bainite exceeds 60%, impact toughness may be degraded due to a relatively high level of strength. In a case in which the fraction of granular bainite is less than 10%, sufficient yield strength may be difficult to secure due to a reduction in strength. For example, yield strength of 460 MPa or greater may be difficult to secure.
[0072] In detail, the fraction of granular bainite may be limited to 20% to 50%, and more specifically, to 30% to 50%.
[0073] In a case in which a fraction of upper bainite in the central portion exceeds 20%, a microcrack may be generated in a front end of a crack during brittle crack propagation, thereby degrading brittle crack arrestability. Thus, the fraction of upper bainite in the central portion may be 20% or less.
[0074] In detail, the fraction of upper bainite may be limited to 15% or less, and more specifically, to 10% or less.
[0075] In a case in which the circle-equivalent diameter of the effective grain of upper bainite in the central portion having a high angle grain boundary of 15° or greater measured using an EBSD method exceeds 15 μm, there is a problem in which a crack may be easily generated despite a relatively low fraction of upper bainite. Thus, the circle-

equivalent diameter of the effective grain of upper bainite in the central portion may be 15 μm or less.

[0076] In a case in which the surface portion microstructure in the region at a depth of 2 mm or less, directly below the surface, includes ferrite in an amount of 20% or greater, crack propagation may be effectively prevented on the surface during brittle crack propagation, thereby improving brittle crack arrestability.

[0077] In detail, the fraction of ferrite may be limited to 30% or greater, and more specifically, to 40% or greater.

[0078] Ferrite in the microstructure in the central portion and the surface portion refers to polygonal ferrite or elongated polygonal ferrite.

[0079] In a case in which a fraction of martensite-austenite in the HAZ of the steel material exceeds 5%, martensite-austenite functions as a starting point of a crack, thereby degrading brittle crack initiation resistance. Thus, the fraction of martensite-austenite in the HAZ may be 5% or less.

[0080] Welding heat input during welding may be 0.5 kJ/mm to 10 kJ/mm.

[0081] A welding method during welding is not specifically limited and may include, for example, flux cored arc welding (FCAW), submerged arc welding (SAW), and the like.

[0082] The steel material may have yield strength of 460 MPa or greater.

[0083] The steel material may have a Charpy fracture transition temperature of -40°C or lower in a 1/2t position in a steel material thickness direction, where t is a steel sheet thickness.

[0084] The steel material have a thickness of 50 mm or greater, and in detail, a thickness of 50 mm to 100 mm.

[0085] Hereinafter, a method of manufacturing a high-strength steel material having excellent brittle crack arrestability according to another aspect of the present disclosure will be described in detail.

[0086] According to another aspect of the present disclosure, the method of manufacturing a high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance comprises rough rolling a slab at a temperature of 900°C to 1100°C after reheating the slab at 1000°C to 1100°C, including, by wt%, C: 0.05% to 0.09%, Mn: 1.5% to 2.2%, Ni: 0.3% to 1.2%, Nb: 0.005% to 0.04%, Ti: 0.005% to 0.04%, Cu: 0.1% to 0.8%, Si: 0.05% to 0.3%, Al: 0.005% to 0.05%, P: 100 ppm or less, S: 40 ppm or less, Fe as a residual component thereof, and inevitable impurities; obtaining a steel sheet by finish rolling a bar obtained from the rough rolling a slab, at a temperature in a range of Ar_3 + 60°C to Ar_3°C, based on a temperature of a central portion; and cooling the steel sheet to 500°C or lower.

Reheating a slab

[0087] A slab is reheated before rough rolling.

[0088] A reheating temperature of the slab may be 1000°C or higher so that a carbonitride of Ti and/or Nb, formed during casting, may be solidified.

[0089] However, in a case in which the slab is reheated at a significantly high temperature, austenite may become coarse. Thus, an upper limit value of the reheating temperature may be 1100°C.

Rough rolling

[0090] A reheated slab is rough rolled.

[0091] A rough rolling temperature may be a temperature Tnr at which recrystallization of austenite is halted, or higher. Due to rolling, a cast structure, such as a dendrite formed during casting, may be destroyed, and an effect of reducing a size of austenite may also be obtained. In order to obtain the effect, the rough rolling temperature may be limited to 900°C to 1100°C.

[0092] In more detail, the rough rolling temperature may be 950°C to 1050°C.

[0093] In an exemplary embodiment, in order to refine a structure of the central portion during rough rolling, a reduction ratio per pass of three final passes during rough rolling may be 5% or greater, and a total cumulative reduction ratio may be 40% or greater.

[0094] In detail, the reduction ratio per pass may be 7% to 20%.

[0095] In detail, the total cumulative reduction ratio may be 45% or greater.

[0096] In the case of a structure recrystallized by initial rolling during rough rolling, grain growth occurs due to a relatively high temperature. However, when three final passes are performed, a bar is air cooled while waiting for a rolling process, so that grain growth speed may be decreased. Thus, during rough rolling, a reduction ratio of the three final passes has the greatest impact on a grain size of a final microstructure.

[0097] In addition, in a case in which the reduction ratio per pass of rough rolling is reduced, sufficient deformation is not transmitted to the central portion, so that toughness may be degraded due to coarsening of the central portion. Therefore, the reduction ratio per pass of the three final passes may be limited to 5% or greater.

[0098] In the meantime, in order to refine a structure of the central portion, the total cumulative reduction ratio during rough rolling may be set to be 40% or greater.

[0099] A strain rate of the three final passes during rough rolling may be 2/sec or lower.

[0100] In general, rolling is difficult at a relatively high reduction ratio due to a relatively great thickness of the bar during rough rolling. Thus, there is a limitation in which it is difficult to transmit a rolling reduction to the central portion of a thick steel plate, thereby allowing an austenite grain size in the central portion to be coarsened. However, as the strain rate is reduced, deformation is transmitted to the central portion even at a relatively low rolling reduction. Thus, the grain size may be refined.

[0101] Therefore, in terms of the three final passes having the greatest impact on the final grain size during rough rolling, the strain rate may be limited to 2/sec or lower, thereby refining the grain size of the central portion. Thus, generation of acicular ferrite and granular bainite may be facilitated.

Finish rolling

[0102] A rough rolled bar may be finish rolled at a temperature of Ar_3 (a ferrite transformation initiation temperature) + 60°C to Ar_3°C to obtain a steel sheet so that a further refined microstructure may be obtained.

[0103] In a case in which rolling is performed at a temperature higher than Ar_3, a relatively large amount of strain bands may be generated in austenite to secure a relatively large number of ferrite nucleation sites, thereby obtaining an effect of securing a fine structure in the central portion of a steel material.

[0104] In addition, in order to effectively generate a relatively large amount of strain bands in austenite, a cumulative reduction ratio during finish rolling may be maintained to be 40% or greater. The reduction ratio per pass, not including skin pass rolling, may be maintained to be 4% or greater.

[0105] In detail, the cumulative reduction ratio may be 40% to 80%.

[0106] In detail, the reduction ratio per pass may be 4.5% or greater.

[0107] In a case in which a finish rolling temperature is reduced to Ar_3 or lower, coarse ferrite is generated before rolling and is elongated during rolling, thereby reducing impact toughness. In a case in which finish rolling is performed at a temperature of Ar_3 + 60°C or higher, the grain size is not effectively refined, so that the finish rolling temperature during finish rolling may be set to be a temperature of Ar_3 + 60 °C to Ar_3°C.

[0108] In an exemplary embodiment, a reduction ratio in an unrecrystallized region may be limited to 40% to 80% during finish rolling.

[0109] As described above, since the reduction ratio in the unrecrystallized region is controlled, thereby increasing a number of nucleation sites of acicular ferrite and granular bainite, generation of structures described above may be facilitated.

[0110] In a case in which the reduction ratio in the unrecrystallized region is significantly low, acicular ferrite and granular bainite may not be sufficiently secured. In a case in which the reduction ratio in the unrecrystallized region is significantly high, strength may be reduced due to generation of pro-eutectoid ferrite caused by a relatively high reduction ratio.

[0111] The grain size of the central portion of the bar in a thickness direction after rough rolling before finish rolling may be 150 μm or less, in detail, 100 μm or less, and more specifically, 80 μm or less.

[0112] The grain size of the central portion of the bar in a thickness direction after rough rolling before finish rolling may be controlled depending on a rough rolling condition, or the like.

[0113] As described above, in a case in which the grain size of the bar after rough rolling before finish rolling may be controlled, a final microstructure is refined due to refinement of an austenite grain. Thus, an advantage of improving low temperature impact toughness may be added.

[0114] The reduction ratio during finish rolling may be set such that a ratio of a slab thickness (mm) to a steel sheet thickness (mm) after finish rolling may be 3.5 or greater, and in detail, 4 or greater.

[0115] As described above, in a case in which the reduction ratio is controlled, as the rolling reduction is increased during rough rolling and finish rolling, an advantage of improving toughness of the central portion may be added by increasing yield strength/tensile strength, improving low temperature toughness, and decreasing the grain size of the central portion in the thickness direction through refinement of the final microstructure.

[0116] After finish rolling, the steel sheet may have a thickness of 50 mm or greater, and in detail, 50 mm to 100 mm.

Cooling

[0117] The steel sheet is cooled to a temperature of 500°C, or lower, after finish rolling.

[0118] In a case in which a cooling end temperature exceeds 500°C, a microstructure may not be properly formed, so that sufficient yield strength may be difficult to secure. For example, yield strength of 460 MPa or greater may be difficult to secure.

[0119] In a case in which the cooling end temperature exceeds 400°C, a generation amount of acicular ferrite and granular bainite may be reduced, and strength thereof may be reduced due to an auto-tempering effect.

[0120] The cooling end temperature may be 400°C or lower.

[0121] The steel sheet may be cooled at a cooling rate of the central portion of 2°C/s or higher. In a case in which the cooling rate of the central portion of the steel sheet is lower than 2°C/s, the microstructure may not be properly formed, so that it may be difficult to secure sufficient yield strength. For example, yield strength of 460 MPa or greater may be difficult.

[0122] In addition, the steel sheet may be cooled at an average cooling rate of 3°C/s to 300°C/s.

[Industrial Applicability]

[0123] Hereinafter, the present disclosure will be described in more detail through exemplary embodiments. However, an exemplary embodiment below is intended to describe the present disclosure in more detail through illustration thereof, but not to limit right scope of the present disclosure, because the right scope thereof is determined by the contents written in the appended claims and reasonably inferred therefrom.

(Exemplary Embodiment)

[0124] A steel slab having a composition illustrated in Table 1 below, which is 400 mm in thickness, was reheated to a temperature of 1045°C, and then rough rolling was started at a temperature of 1020°C, thereby manufacturing a bar. A cumulative reduction ratio of 52% during rough rolling was equally applied to an entirety of steel grades.

[0125] A thickness of a bar having been rough rolled was 192 mm, while a grain size of a central portion after rough rolling before finish rolling, as illustrated in Table 2, was 66 μm to 82 μm. A reduction ratio of three final passes during rough rolling was within a range of 7.9% to 14.1%. A strain rate during rolling was within a range of 1.22/s to 1.68/s.

[0126] After rough rolling, finish rolling was performed at the reduction ratio per pass of 4.2% to 5.6% and at the cumulative reduction ratio of 50% at a temperature equal to a difference between a finish rolling temperature and an Ar_3 temperature, illustrated in Table 2 below to obtain a steel sheet having a thickness illustrated in Table 3 below, and then the steel sheet was cooled to a temperature of 241°C to 378°C at a cooling rate of the central portion of 3.8°C/sec to 5.0°C/sec.

[0127] In terms of the steel sheet manufactured as illustrated above, a microstructure, yield strength, a Kca value (a brittle crack arrestability coefficient), and a crack tip opening displacement (CTOD) value (a brittle crack initiation resistance) were examined, and results thereof were illustrated in Tables 3 and 4 below.

[0128] Surface properties illustrated in Table 3 below were measured to determine whether a star crack in a surface portion is generated by hot shortness occurring depending on a Cu to N addition ratio.

[0129] In addition, the Kca value in Table 4 below is a value evaluated by performing an ESSO test on the steel sheet. The CTOD value was a result in which a FCAW (1.0 kJ/mm) welding process is performed to carry out structure analysis and a CTOD test on the HAZ.

[Table 1]

Steel Grade	Steel Composition (wt%)										
	C	Si	Mn	Ni	Cu	Ti	Nb	Al	P(pp m)	S(pp m)	Cu/Ni wt%
Inventive Steel 1	0.07 8	0.21	1.82	0.73	0.32	0.023	0.03 2	0.03 0	63	18	0.44
Inventive Steel 2	0.06 9	0.19	1.72	0.66	0.39	0.012	0.02 2	0.03 1	72	15	0.59
Inventive Steel 3	0.05 7	0.22	2.05	0.57	0.26	0.017	0.02 7	0.02 5	56	16	0.46
Inventive Steel 4	0.07 2	0.21	1.83	0.62	0.33	0.022	0.01 9	0.03 5	49	13	0.53
Inventive Steel 5	0.08 4	0.17	1.58	1.06	0.49	0.016	0.03 3	0.04 0	66	12	0.46
Inventive Steel 6	0.07 1	0.23	1.93	0.36	0.19	0.018	0.02 8	0.02 5	43	23	0.53
Inventive Steel 7	0.06 6	0.18	1.82	0.79	0.36	0.019	0.01 2	0.02 0	39	31	0.46
Comparative Steel 1	0.13	0.19	1.88	0.63	0.29	0.021	0.02 9	0.02 1	79	13	0.46
Comparative Steel 2	0.06 7	0.45	1.96	0.59	0.22	0.011	0.03 8	0.03 2	88	9	0.37
Comparative Steel 3	0.07 1	0.19	2.44	0.88	0.32	0.013	0.02 1	0.02 9	65	23	0.36

(continued)

Steel Grade	Steel Composition (wt%)										
	C	Si	Mn	Ni	Cu	Ti	Nb	Al	P(pp m)	S(pp m)	Cu/Ni wt%
Comparative Steel 4	0.082	0.18	2.02	1.62	0.52	0.022	0.026	0.034	55	19	0.32
Comparative Steel 5	0.07 2	0.27	1.89	0.62	0.44	0.043	0.04 9	0.03 0	48	22	0.71
Inventive Steel 8	0.06 2	0.18	1.93	0.59	0.63	0.015	0.02 9	0.03 1	65	16	1.07
Comparative Steel 6	0.04 2	0.22	1.36	0.55	0.21	0.019	0.03 3	0.02 7	43	18	0.38

[Table 2]

Exemp lary Embod iment No.	Steel Gra de	Grain Size of Central Portion after Rough Rolling before Finish Rolling (μm)	Average Reducti on Ratio of Three final passes during Rough Rolling (%)	Average Strain Rate of Three final passes during Rough Rolling (/s)	Avera ge Reduc tion Ratio per Pass durin g Finis h Rolli ng (%)	Finish Rollin g Temper ature -Ar_3 Temper ature (°C)	Cool ing Rate of Cent ral Port ion (°C/ sec)	Cooli ng End Tempe ratur e (°C)
Inventive Example 1	Inventive Steel 1	78	11.3	1.61	4.2	35	4.1	324
Inventive Example 2	Inventive Steel 2	66	9.6	1.35	4.5	43	4.4	285
Inventive Example 3	Inventive Steel 3	79	10.3	1.44	5.1	29	4.3	296
Inven tive Examp le 4	Inven tive Steel 4	75	8.9	1.46	5.3	41	3.8	335
Inven tive Examp le 5	Inven tive Steel 5	69	13.2	1.67	4.8	23	3.9	342
Inventive Example 6	Inventive Steel 6	73	14.1	1.32	4.2	15	4.3	312
Comparative Example 1	Inventive Steel 7	79	12.2	1.22	4.7	93	4.8	256
Comparative Example 2	Comparative Steel 1	76	10.3	1.68	5.6	28	4.4	330
Comparative Example 3	Comparative Steel 2	68	13.1	1.55	5.2	26	4.2	351
Comparative Example 4	Comparative Steel 3	77	7.9	1.54	5.3	39	4.1	241
Compa rative Example 5	Comparative Steel 4	81	10.1	1.39	4.5	18	4.6	378
Comparative Example 6	Comparative Steel 5	82	10.9	1.41	4.9	13	4.7	312

(continued)

Exemp lary Embod iment No.	Steel Gra de	Grain Size of Central Portion after Rough Rolling before Finish Rolling (μm)	Average Reducti on Ratio of Three final passes during Rough Rolling (%)	Average Strain Rate of Three final passes during Rough Rolling (/s)	Avera ge Reduc tion Ratio per Pass durin g Finis h Rolli ng (%)	Finish Rollin g Temper ature -Ar_3 Temper ature (°C)	Cool ing Rate of Cent ral Port ion (°C/ sec)	Cooli ng End Tempe ratur e (°C)
Inventive Examp	Inventive Steel 8	73	11.3	1.43	5.1	46	4.1	333
Comparative Example 7	Comparative Steel 6	72	9.8	1.52	5.3	53	5.0	316

[Table 3]

Exemplary Embod iment No.	Steel Grade	Surface Prope rties	Steel Sheet Thic knes s (mm)	Microstructure Phase Fraction of Central Portion (area%)				Phase Fraction of Ferrit e of Surface Portion (%)
				Acicular Ferrite (AF)	Granular Bainite (GB)	Upper Bainite (Average Grain Size, μm)	Residual Phase Fraction (Ferrite/Pearlite/One or more of MA	
Inventive Example 1	Inventive Steel 1	None	85	37.9	45.4	13.9(12. 2)	2.8	45
Inventive Example 2	Inventive Steel 2	None	80	49.3	41.9	7.2(9.5)	1.6	39
Inventive Example 3	Inventive Steel 3	None	75	36.5	45.1	16.8(9.6)	1.6	53
Inventive Example 4	Inventive Steel 4	None	95	39.8	53.2	5.1(8.6)	1.9	26
Inventive Example 5	Inventive Steel 5	None	90	42.6	45.6	9.2(11.3)	2.6	51
Inventive Example 6	Inventive Steel 6	None	100	29.7	48.8	15.6(11. 6)	5.9	66
Comparative Example 1	Inventive Steel 7	None	90	32.6	46.6	12.4(12. 2)	8.4	0
Comparative Example 2	Comparative Steel 1	None	90	20.5	39.3	32(18.3)	8.2	59
Comparative Example 3	Comparative Steel 2	None	85	51.2	31	11.3(9.7)	6.5	52
Compa rativ e Examp le 4	Comparative Steel 3	None	80	16.8	32.1	43.8(17. 1)	7.3	36
Comparative Example 5	Comparative Steel 4	None	90	13.2	30.6	49.8(19. 0)	6.4	61
Comparative Example 6	Comparative Steel 5	None	75	37.5	41.6	13.7(13. 3)	7.2	59
Inventive Example 7	Inventive Steel 8	Occur	90	53.9	38.7	6.1(9.6)	1.3	31

(continued)

Exemplary Embod iment No.	Steel Grade	Surface Prope rties	Steel Sheet Thic knes s (mm)	Microstructure Phase Fraction of Central Portion (area%)				Phase Fraction of Ferrit e of Surface Portion (%)
				Acicular Ferrite (AF)	Granular Bainite (GB)	Upper Bainite (Average Grain Size, μm)	Residual Phase Fraction (Ferrite/Pearlite/One or more of MA	
Comparative Example 7	Comparative Steel 6	None	90	13.2	20.3	6.2(13.2)	60.3	33

[Table 4]

Exempl ary Embodi ment No.	Steel Grade	Yield Strengt h (Mpa)	Kca ($N/mm^{1.5}$, @-10°C)	MA Fraction in HAZ (%)	CTOD Value in HAZ (mm)
Inventive Example 1	Inventive Steel 1	512	7120	1.8	0.56
Inventive Example 2	Inventive Steel 2	489	6988	1.9	0.49
Inventive Example 3	Inventive Steel 3	504	7286	2.6	0.55
Inventive Example 4	Inventive Steel 4	518	7057	2.3	0.43
Inventive Example 5	Inventive Steel 5	485	7516	2.1	0.68
Inventive Example 6	Inventive Steel 6	523	7030	3.2	0.56
Comparative Example 1	Inventive Steel 7	501	5549	3.9	0.47
Comparative Example 2	Comparative Steel 1	579	4256	6.8	0.12
Comparative Example 3	Comparative Steel 2	496	6775	7.8	0.16
Comparative Example 4	Comparative Steel 3	577	4356	3.1	0.22
Comparative Example 5	Comparative Steel 4	562	4150	2.1	0.59
Comparative Example 6	Comparative Steel 5	532	6554	7.2	0.12
Inventive Example 7	Inventive Steel 8	516	7211	1.3	0.54
Comparative Example 7	Comparative Steel 6	435	5026	2.1	0.62

[0130] As illustrated in Tables 1 to 4, in the case of Comparative Example 1, the difference between the finish rolling temperature during finish rolling and the Ar_3 temperature, proposed in an exemplary embodiment, was controlled to be 60°C or higher. Rolling was performed at a relatively high temperature, so that sufficient reduction was not applied to the central portion. In addition, cooling was started at a relatively high temperature, so that ferrite of 20% or greater was not generated in a surface portion. Thus, it can be confirmed that the Kca value measured at a temperature of -10°C may not exceed 6000 required in a steel material for shipbuilding of the related art.

[0131] In the case of Comparative Example 2, a C content had a value higher than an upper limit value of a C content of an exemplary embodiment. It can be confirmed that a relatively large amount of coarse upper bainite was generated in the central portion during rough rolling, so the Kca value measured at a temperature of -10° C was 6000 or less. It can be confirmed that a relatively large amount of martensite-austenite (MA) was also generated in the HAZ, so the CTOD value was 0.25 mm or less .

[0132] In the case of Comparative Example 3, a Si content had a value higher than an upper limit value of a Si content of an exemplary embodiment. It can be confirmed that a relatively large amount of Si was added to generate a relatively large amount of an MA structure in the HAZ, so the CTOD value is 0.25 mm or less.

[0133] In the case of Comparative Example 4, a Mn content has a value higher than an upper limit value of a Mn content of an exemplary embodiment. It can be confirmed that due to having a relatively high level of hardenability, a relatively large amount of upper bainite is formed in the central portion, thereby allowing the Kca value to be 6000 or less at a temperature of -10°C. In addition, it can be confirmed that due to a relatively high carbon equivalent (Ceq) value, a relatively small amount of MA phase was present in the HAZ, but the CTOD value is 0.25 or less.

[0134] In the case of Comparative Example 5, an Ni content had a value higher than an upper limit value of an Ni content of an exemplary embodiment. It can be confirmed that due to a relatively high level of hardenability, a relatively large amount of upper bainite was generated in the central portion, thereby allowing the Kca value to be 6000 or less at a temperature of -10°C. However, it can be confirmed that due to a relatively high Ni content, the CTOD value was relatively high.

[0135] In the case of Comparative Example 6, an Nb and Ti content has a value higher than an upper limit value of an Nb and Ti content of an exemplary embodiment. It can be confirmed that an entirety of other conditions satisfies a condition suggested in an exemplary embodiment, but due to a relatively high Nb and Ti content, a relatively large amount of the MA structure is generated in the HAZ, thereby allowing the CTOD value to be 0.25 mm or less.

[0136] Inventive Example 7 includes a component exceeding a ratio of Cu to Ni suggested in an aspect of the present disclosure. It can be confirmed that despite having other, significantly excellent physical properties, a star crack was generated on a surface, thereby causing a default in surface quality.

[0137] In the case of Comparative Example 7, a C and Mn content has a value lower than a lower limit value of a C and Mn content of an exemplary embodiment. It can be confirmed that due to a relatively low level of hardenability, a fraction of AF+GB in the central portion is significantly low, and a relatively large amount of polygonal ferrite and a pearlite structure of 10% or greater are present, thereby allowing the Kca value to be 6000 or less at a temperature of -10°C.

[0138] On the other hand, in the case of Inventive Examples 1 to 6, satisfying a composition range and a manufacturing range of an exemplary embodiment, AF + GB of a microstructure in the central portion was 70% or greater, a fraction of upper bainite in the central portion was 20% or less, a circle-equivalent diameter of an effective grain of upper bainite of the central portion having a high angle grain boundary of 15° or greater was 15 μm or less, and a fraction of the MA phase in the HAZ was less than 5%.

[0139] It can be confirmed that, in Inventive Examples 1 to 6, yield strength satisfies 460 MPa or greater, the Kca value satisfies a value of 6000 or greater at a temperature of -10°C, and the CTOD value also represents a relatively high value of 0.25 mm or greater.

[0140] While exemplary embodiments have been shown and described above, it will be apparent to those skilled in the art that modifications and variations could be made without departing from the scope of the present invention as defined by the appended claims.

Claims

1. A high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance, comprising:

by wt%, carbon (C): 0.05% to 0.09%, manganese (Mn): 1.5% to 2.2%, nickel (Ni): 0.3% to 1.2%, niobium (Nb): 0.005% to 0.04%, titanium (Ti): 0.005% to 0.04%, copper (Cu): 0.1% to 0.8%, silicon (Si): 0.05% to 0.3%, aluminum (Al): 0.005% to 0.05%, phosphorus (P): 100 ppm or less, sulfur (S): 40 ppm or less, iron (Fe) as a residual component, and inevitable impurities,
wherein a microstructure of a central portion includes, by area%, a mixed phase of acicular ferrite and granular bainite in an amount of 70% or greater, upper bainite in an amount of 20% or less, and one or more selected from a group consisting of ferrite, pearlite, and martensite-austenite (MA), as residual components; a circle-equivalent diameter of an effective grain of the upper bainite having a high angle grain boundary of 15° or greater measured using an electron backscatter diffraction (EBSD) method being 15 μm or less; a surface portion microstructure in a region at a depth of 2 mm or less, directly below a surface, includes, by area%, ferrite in an amount of 20% or greater and one or more of bainite and martensite as residual components; and a heat affected zone (HAZ) formed during welding includes, by area%, martensite-austenite (MA) in an amount of 5% or less.

2. The high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance of claim 1, comprising a thickness of 50 mm or greater.

3. The high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance of claim 1, wherein, in terms of a Cu and Ni content, a weight ratio of Cu to Ni is 0.8 or less.

4. The high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance of claim 1, wherein welding heat input during welding is 0.5 kJ/mm to 10 kJ/mm.

5. The high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance of claim 4, wherein a welding method during welding includes flux cored arc welding (FCAW) or submerged arc welding (SAW).

6. The high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance of claim 1, comprising yield strength of 460 MPa or greater.

7. The high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance of any of claims 1 to 6, comprising a Kca value measured at a temperature of -10°C of 6000 or greater.

8. The high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance of claim 1, comprising a Charpy fracture transition temperature of -40°C or lower in a 1/2t position in a steel material thickness direction, where t is a steel sheet thickness.

9. A method of manufacturing a high-strength steel material having excellent brittle crack arrestability and welding zone brittle crack initiation resistance, comprising:

rough rolling a slab at a temperature of 900°C to 1100°C after reheating the slab at 1000°C to 1100°C, including, by wt%, C: 0.05% to 0.09%, Mn: 1.5% to 2.2%, Ni: 0.3% to 1.2%, Nb: 0.005% to 0.04%, titanium (Ti): 0.005% to 0.04%, copper (Cu): 0.1% to 0.8%, silicon (Si): 0.05% to 0.3%, aluminum (Al): 0.005% to 0.05%, phosphorus (P): 100 ppm or less, sulfur (S): 40 ppm or less, iron (Fe) as a residual component, and inevitable impurities;
obtaining a steel sheet by finish rolling a bar obtained from the rough rolling a slab, at a temperature in a range of Ar_3 + 60°C to Ar_3°C, based on a temperature of a central portion; and
cooling the steel sheet to 500°C or lower.

10. The method of claim 9, wherein a thickness of the steel sheet having been finish rolled is 50 mm or greater.

11. The method of claim 9, wherein a reduction ratio per pass of three final passes during the rough rolling a slab is 5% or greater, and a total cumulative reduction ratio is 40% or greater.

12. The method of claim 9, wherein a strain rate of three final passes during the rough rolling a slab is 2/sec or lower.

13. The method of claim 9, wherein a grain size of a central portion of a bar thickness before finish rolling after the rough rolling a slab is 150 μm or less.

14. The method of claim 9, wherein a reduction ratio during the finish rolling is set such that a ratio of a slab thickness (mm) to a steel sheet thickness (mm) after the finish rolling is 3.5 or greater.

15. The method of claim 9, wherein a cumulative reduction ratio during the finish rolling is maintained to be 40% or greater, and a reduction ratio per pass, not including skin pass rolling, is maintained to be 4% or greater.

16. The method of claim 9, wherein the cooling the steel sheet is performed at a cooling rate of the central portion of 2°C/s or higher.

17. The method of claim 9, wherein the cooling the steel sheet is performed at an average cooling rate of 3°C/s to 300°C/s.

INTERNATIONAL SEARCH REPORT

International application No.
PCT/KR2016/014088

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/16(2006.01)i, C22C 38/14(2006.01)i, C22C 38/12(2006.01)i, C22C 38/06(2006.01)i, C22C 38/04(2006.01)i, C21D 8/02(2006.01)i, B21B 37/16(2006.01)i

According to International Patent Classification (IPC) or to both national classification and IPC

B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)

C22C 38/16; C21D 1/02; C21D 8/02; C22C 38/00; C21D 8/00; C21D 7/13; C22C 38/04; B21B 3/00; C22C 38/14; C22C 38/12; C22C 38/06; B21B 37/16

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched
Korean Utility models and applications for Utility models: IPC as above
Japanese Utility models and applications for Utility models: IPC as above

Electronic data base consulted during the international search (name of data base and, where practicable, search terms used)
eKOMPASS (KIPO internal) & Keywords: brittle crack propagation resistance, high strength, welding part, ferrite, bainite, martensite, effective grain, rough rolling, finish rolling

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X	KR 10-2012-0075274 A (POSCO) 06 July 2012 See paragraphs [0056]-[0059], claims 1-4, 6-9 and table 3.	9-11,13-17
A		1-8,12
A	KR 10-2011-0075321 A (POSCO) 06 July 2011 See abstract and claims 1-9.	1-17
A	KR 10-2012-0097160 A (HYUNDAI STEEL COMPANY) 03 September 2012 See abstract and claims 1, 5-8.	1-17
A	KR 10-2014-0113975 A (JFE STEEL CORPORATION) 25 September 2014 See abstract and claims 1-4.	1-17
A	KR 10-2015-0112489 A (HYUNDAI STEEL COMPANY) 07 October 2015 See claims 1-8.	1-17

☐ Further documents are listed in the continuation of Box C. ☒ See patent family annex.

* Special categories of cited documents:
"A" document defining the general state of the art which is not considered to be of particular relevance
"E" earlier application or patent but published on or after the international filing date
"L" document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified)
"O" document referring to an oral disclosure, use, exhibition or other means
"P" document published prior to the international filing date but later than the priority date claimed
"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention
"X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone
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