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(54) HOT ROLLED STEEL SHEET HAVING EXCELLENT COLD BENDABILITY AFTER HEATING AND QUENCHING-TEMPERING HEAT TREATMENT, STEEL PIPE, MEMBER AND MANUFACTURING METHODS THEREFOR

(57) A hot rolled steel sheet having excellent cold bendability after heating and quenching-tempering heat treatment, a steel pipe, a member and manufacturing methods therefor are provided. The present invention relates to: a hot rolled steel sheet comprising, by wt%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0.004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and the balance of Fe and other impurities, and satisfying relations 1-3; a steel pipe; and a member.

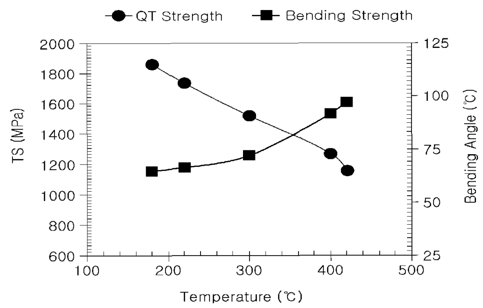

FIG. 1

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Description

Technical Field

[0001] The present disclosure relates to a hot-rolled steel sheet used in automobile body components such as an automobile suspension component, a steel pipe and a member using the same and a method of manufacturing the same, and more particularly, a hot-rolled steel sheet of which a hardness difference (ΔHv) value between a surface portion of a steel sheet and a position portion of t/4 (where t is a thickness of the steel sheet) is less than 15, which is excellent, and which exhibits a high strength of more than 1300 MPa and a maximum bending angle of more than 40 degrees after heating and quenching-tempering heat treatment, a steel pipe and a member using the same and a method of manufacturing the same.

Background Art

[0002] Among vehicle chassis components, a suspension component may be one of the components requiring fatigue durability, and in accordance with the trend of component lightweighting, high strength and thin hot-rolled steel sheets are preferable.

[0003] Meanwhile, such suspension components may be generally manufactured using pipe-shaped materials through hot-forming or cold-forming and heat treatment. However, for eco-friendly manufacturing and cost reductions, a new manufacturing method including manufacturing an ultra-high strength steel pipe on which heat treatment is performed and manufacturing a component through cold-forming has been suggested. This may be to manufacture a steel pipe having a high strength of 1000MPa or more into a component having a desired shape by external bending force in a cold (≤25°C) or warm (≤550°C) state, and in this case, the heat treated steel sheet or steel pipe itself needs to exhibit high resistance against bending cracks or high ductility.

[0004] Meanwhile, as a method to improve bending formability of a steel sheet, research to uniformly control a ferrite and pearlite microstructure of a final cold-rolled steel sheet regardless of whether heat treatment is applied, or controlling a prior austenite size on martensite of 95% or more and the size distribution of carbides retained in martensite have been conducted.

[0005] Reference 1 suggests a method of improving bendability of a steel sheet for hot-forming to address the problem of decreased bendability caused by obtaining high strength of hot-press formed product. According to the method, by controlling a Mn/Si ratio in the range of 0.05 to 2.0 and adding 0.5% or more of silicon (Si) element to a steel used for cold-rolled steel sheet and performing cold-rolled annealing heat treatment, a microstructure of the cold-rolled steel sheet, specifically the pearlite phase, may be uniformly distributed, and by hot-press forming the cold-rolled steel sheet and performing painting heat treatment, a retained austenite phase may be formed in a martensite structure, bendability may improve. That is, the reference suggests adding 0.5% or more of silicon (Si) content to steel used as a cold-rolled steel sheet. To apply the process directly to a steel pipe-shaped chassis component, a large amount of silicon oxide formed in the butt weld or melted zone may need to be effectively discharged during an electric arc welding process, and when it is not possible to effectively discharged oxides, welding defects may occur, which may reduce formability such as flattening or expansion characteristics of the steel pipe, such that there may be limitation in use.

[0006] Reference 2 suggests the method to improve bendability (limit bending radius(R)/thickness(t) ≤2.4) of a high strength cold-rolled steel sheet of 1470MPa or more by controlling the number of fine carbides of 25 to 60 nm size in a martensite single-phase structure including prior austenite (PAGS) having a fine size of 5 to 6 um to 0 to 500,000 or less per 1 mm^2 area. Meanwhile, the improved bendability or extremely low R/t measurement value of the high strength cold-rolled steel sheet appears to be mainly due to a fine prior austenite size, which may be because formation or propagation of cracks caused by external bending force at a prior austenite grain boundary or carbide/martensite interface is delayed. In particular, due to the extremely fine martensite structure, tensile strength of a cold-rolled steel sheet may be high, 1470 MPa or more, but it may be expected to exhibit low elongation of less than 6%, such that formability may be insufficient to manufacture a component having a complex shape. Even in the case of high strength steel, a large amount of silicon (Si) content of more than 1.0% may be added to control a size of carbide precipitating and growing during tempering heat treatment of a martensite structure formed by rapid cooling of 100°C/sec or more, such that it may be difficult to apply the steel as a steel pipe or a steel pipe component.

[0007] Therefore, from the review of the process of manufacturing a steel sheet and a steel component suggested in the references mentioned above, as a steel sheet which may be used in manufacturing an electric arc welding steel pipe or drawn steel pipe having undergone heating and quenching-tempering heat treatment in advance into a stabilizer component having desired shape through external bending force, there has been no suggestion of a hot-rolled steel sheet, a steel pipe and a member and method of manufacturing the same of which a hardness difference (ΔHv) value between a surface portion of the steel sheet and a thickness/4 position portion is less than 15, and which exhibits high strength of 1300MPa or more (elongation of 8% or more) and a maximum bending angle of 40° or more after heat treatment.

[Prior art]

[Patent Reference]

[0008]

(Cited document 1) Korean Registered Patent Publication No. 10-1568549
(Cited document 1) Korean Laid-Open Patent Publication No. 10-2015-0105476

Detailed description of present disclosure

Technical problems to solve

[0009] A preferable aspect of the present disclosure is to provide a hot-rolled steel sheet for cold-forming member exhibiting excellent bendability, of which a hardness difference between a steel sheet surface portion and a thickness/4 position portion is small(decarburization resistance is large), and which has high strength and a maximum bending angle of 40° or more after heating and quenching-tempering heat treatment, and a method of manufacturing the same.

[0010] Also, another preferable aspect of the present disclosure is to provide a steel pipe for a cold-forming member manufactured using a hot-rolled steel sheet exhibiting excellent bendability, of which a hardness difference between a steel sheet surface portion and a thickness/4 position portion is small(decarburization resistance is large), and which has high strength and a maximum bending angle of 40° or more after heating and quenching-tempering heat treatment, and a method of manufacturing the same.

[0011] Further, a preferable aspect of the present disclosure is to provide a cold-forming member manufactured using a steel pipe exhibiting excellent bendability, of which a hardness difference between a steel sheet surface portion and a thickness/4 position portion is small(decarburization resistance is large), and which has high strength and a maximum bending angle of 40° or more after heating and quenching-tempering heat treatment, and a method of manufacturing the same.

[0012] The purpose of present disclosure is not limited to the above aspects. Anyone having ordinary knowledge in the technical field to which the present disclosure belongs may have no difficulty in understanding the additional purpose of the invention from the overall description of the present disclosure.

Solution to Problem

[0013] An aspect of the present disclosure relates to a hot-rolled steel sheet for a cold-forming member having excellent bendability, which has high strength and a maximum bending angle of 40° or more after quenching-tempering heat treatment, including, by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0.004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities, and satisfying relational expressions 1-3, wherein a hardness difference value between a surface portion and a thickness/4 position portion of the steel sheet is less than 15, wherein, by volume%, the hot-rolled steel sheet has a microstructure including 20-65% of ferrite and 35-80% of pearlite, and wherein an average grain size of prior austenite is 15 um or more,

[Relational expression 1]

$$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

[Relational expression 2]

$$(Ni)/(Mn) \geq 0.05 \text{ (weight ratio)}$$

[Relational expression 3]

$$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio)}.$$

[0014] Another aspect of the present disclosure provides a method of manufacturing a hot-rolled steel sheet for a cold-forming member having excellent bendability, which has high strength and a maximum bending angle of 40° or more after quenching-tempering heat treatment, including,

heating a steel slab having a composition as above in a temperature range of 1150-1300°C;
obtaining a hot-rolled steel sheet by hot-rolling the heated slab at an Ar_3 temperature or higher, the hot-rolling including rough-rolling and finishing rolling; and
cooling the hot-rolled steel sheet on a run-out table and coiling the hot-rolled steel sheet at a temperature of 550 to 750°C, wherein the hot-rolled steel sheet has a hardness difference value of less than 15 between a surface portion and a thickness/4 position portion.

[0015] Obtaining a hot-rolling pickling steel sheet by pickling the hot-rolled steel sheet may be further included.

[0016] Another aspect of the present disclosure provides a steel pipe for a cold-forming member having excellent bendability, which has high strength and a maximum bending angle of 40° or more after quenching-tempering heat treatment, by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0.004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities, satisfying relational expressions 1-3, having a hardness difference value of less than 15 between a surface portion and a thickness/4 position portion of the steel pipe, having, by volume%, a microstructure including 20-65% of ferrite and 35-80% of pearlite, and including prior austenite having an average grain size of 15 μm or more,

[Relational expression 1]

$$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

[Relational expression 2]

$$(Ni)/(Mn) \geq 0.05 \text{ (weight ratio)}$$

[Relational expression 3]

$$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio)}.$$

[0017] Another aspect of the present disclosure provides a method of manufacturing a steel pipe for a cold-forming member including

heating a steel slab having a composition as above in a temperature range of 1150-1300°C;
obtaining a hot-rolled steel sheet by hot-rolling the heated slab at an Ar_3 temperature or higher, the hot-rolling including rough-rolling and finishing rolling;
cooling the hot-rolled steel sheet on a run-out table and coiling the hot-rolled steel sheet at a temperature of 550 to 750°C;
obtaining a steel pipe by welding the hot-rolled steel sheet; and
performing annealing heat treatment on the steel pipe at a temperature of Ac_1 -50°C-Ac_3 +150°C for 3-60 minutes.

[0018] Drawing out the annealing-heat treated steel pipe may be further included.

[0019] A reheating step of heating the annealing-heat treated and drawn-out steel pipe to a temperature of Ar_3-970°C at a heating rate of 10°C/sec or more and maintaining the steel pipe for less than 60 seconds; a quenching step of cooling the reheated steel pipe to room temperature at a cooling rate of 20-350°C/sec; and a tempering heat treatment step of heating the quenched steel pipe at a heating rate of 2-20°C/sec to a temperature range of 150-350°C and maintaining the steel pipe at the above temperature may be further included.

[0020] Another aspect of the present disclosure provides a cold-forming member having high strength and excellent bendability including by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0.004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities, satisfying relational expressions 1-3, and having, by volume%, a microstructure including 95% of one or more of martensite and tempered martensite and a balance of 5% or less of retained austenite, wherein an average grain size of prior austenite is 15 um or more, and the number of Fe_3C carbides having an average circle equivalent size of 300 nm or less is 30 or less per unit area (μm^2)

[Relational expression 1]

$$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

[Relational expression 2]

$$(Ni)/(Mn) \geq 0.05 \text{ (weight ratio)}$$

[Relational expression 3]

$$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio)}$$

[0021] Another aspect of the present disclosure provides a method of manufacturing a cold-forming member including,

heating a steel slab having a composition as above in a temperature range of 1150-1300°C;
obtaining a hot-rolled steel sheet by hot-rolling the heated slab at an Ar_3 temperature or higher, the hot-rolling including rough-rolling and finishing rolling;
cooling the hot-rolled steel sheet on a run-out table and coiling the hot-rolled steel sheet at a temperature of 550 to 750°C;
obtaining a steel pipe by welding the hot-rolled steel sheet;
performing annealing heat treatment and drawing out the steel pipe;
performing a reheating step of heating the annealing-heat treated or drawn steel pipe to a temperature of Ar_3-970°C at a heating rate of 10°C/sec or more and maintaining the steel pipe for less than 60 seconds;
performing a quenching step of cooling the reheated steel pipe to room temperature at a cooling rate of 20-350°C/sec more;
a tempering heat treatment step of heating the quenched steel pipe at a heating rate of 2-20°C/sec to a temperature range of 150-350°C and maintaining the steel pipe at the above temperature; and
cold-forming the tempering-heat treated steel pipe into a member.

Advantageous Effects of Invention

[0022] According to an aspect of the present disclosure, a hot-rolled steel sheet and a steel pipe for a forming member, which has an excellent hardness difference of less than 15 between a surface portion of a steel sheet and a thickness/4 position portion, and which exhibits a high strength of 1300MPa or more and a maximum bending angle of 40° or more after heating and quenching-tempering heat treatment may be provided, and also, a components may be manufactured through cold-forming, which has the effect of reducing component manufacturing costs.

Brief Description of Drawings

[0023]

FIG. 1 is a diagram illustrating strength and maximum bending angle change curves of a steel sheet after tempering heat treatment of inventive example 2 in an embodiment of the present disclosure.

FIG. 2 is an image of a shape after cold bending of a quenching-tempering heat treated steel pipe of inventive example 2 in an embodiment of the present disclosure.
FIG. 3(a-b) is an image indicating whether cracks occur in a heat treated steel pipe of an embodiment of the present disclosure, in which (a) indicates inventive example 2 and (b) indicates comparative example 2.

Best Mode for Invention

[0024] Hereinafter, the present disclosure will be described.
[0025] First, a hot-rolled steel sheet for cold-forming member exhibiting excellent bendability, which has a hardness difference of less than 15 between a steel sheet surface portion and a thickness/4 position portion (great decarburization resistance), and which has high strength and a maximum bending angle of 40° or more after heating and quenching-tempering heat treatment will be described.
[0026] The hot-rolled steel sheet for a cold-forming member in the present disclosure may include, by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0.004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities, may satisfy relational expressions 1-3, and may have a microstructure including, by volume%, 20-65% of ferrite and 35-80% of pearlite, wherein hardness in each of a surface portion and a thickness/4 position portion of the steel sheet has a hardness difference value of less than 15, and an average grain size of prior austenite is 15 um or more.
[0027] First, the reasons for limiting an alloy composition and content of a steel sheet and a steel pipe for a cold-forming member in the present disclosure will be described. Hereinafter, "%" indicates "weight%" unless otherwise indicated.

C: 0.20% or more and less than 0.35%

[0028] Carbon (C) may be effective in increasing strength of steel and may increase strength after quenching-tempering heat treatment. When a content thereof is less than 0.20%, it may be difficult to assure sufficient strength of 1300 MPa or more after tempering heat treatment, whereas when the content is 0.35% or more, martensite having excessive hardness or strength may be formed, such that it may be difficult to assure a bending angle of 40° or more when cold bending is performed on a steel sheet material or a steel pipe component after heat treatment. To increase the bending angle, a tempering heating temperature may be increased to 350°C or more, but there may be a limitation in ensuring strength of 1300MPa or more. Accordingly, it may be preferable to limit the carbon (C) content to 0.20% or more and less than 0.35%.

Mn: 0.5-1.3%

[0029] Manganese (Mn) may be an essential element for increasing strength of steel, and may increases strength after quenching heat treatment of steel. When the content thereof is less than 0.5%, it may be difficult to assure sufficient strength of 1300MPa or more after tempering heat treatment, whereas when the content exceeds 1.3%, it may be advantageous to assure strength, but after heat treatment, it may difficult to assure a bending angle of 40° or more for a steel sheet and a steel pipe. In this case, a prior austenite size may be appropriately controlled to improve cold bendability, or a relatively large amount of nickel (Ni) content may be necessary to suppress carbide growth, which may increase manufacturing costs. Also, a segregation region may be formed in and/or outside a continuous casting slab and a hot-rolled steel sheet, which may cause a high frequency of processing defects while manufacturing a steel pipe. Accordingly, it may be preferable to limit the manganese (Mn) content to 0.5-1.3%.

Si: 0.3% or less (excluding 0%)

[0030] Silicon (Si) may be added to improve strength or ductility, and may be added in a range in which there is no problem of surface scaling of a hot-rolled steel sheet and a hot-pickled steel sheet. When the content thereof exceeds 0.3% or more, surface defects may occur due to formation of silicon oxide, such that removal by pickling may be difficult. When oxide discharge is not smoothly performed in a melted zone of a welded zone of a steel pipe, formability of the steel pipe may decrease. Accordingly, a silicon (Si) content may be limited to 0.3% or less.

Relational expression 1

[0031] Mn and Si may need to satisfy relational expression 1 as below.

[Relational expression 1]

$$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

[0032] The Mn/Si ratio may be an important parameter which determines quality of a welded zone of a steel pipe. When the Mn/Si ratio is less than 2, the Si content may be relatively high and silicon oxide may be formed in a molten metal in a welded zone, and in the case in which the element is not forcibly discharged, defects may be formed in the welded zone, such that defects may occur in manufacturing a steel pipe. Accordingly, the Mn/Si ratio may be limited to 2 or more.

P: 0.03% or less (including 0%)

[0033] Phosphorus (P) may segregate to an austenite grain boundary and/or an interphase grain boundary and may cause embrittlement. Accordingly, the content of phosphorus (P) may be maintained as low as possible, and an upper limit thereof may be limited to 0.03%. The preferable phosphorus (P) content may be 0.02% or less.

S: 0.004% or less (including 0%)

[0034] Sulfur (S) may segregate into MnS non-metallic inclusion during continuous casting solidification in steel, and may cause high temperature cracks. Also, since sulfur (S) may deteriorate impact toughness of a heat treatment steel sheet or a steel pipe, it may be necessary to control the content thereof to be as low as possible. Accordingly, in the present disclosure, it may be preferable to maintain the sulfur (S) content to be as low as possible and to limit an upper limit thereof to 0.004%.

Al: 0.04% or less (excluding 0%)

[0035] Aluminum (Al) may be added as a deoxidizing agent. Meanwhile, aluminum (Al) may react with nitrogen (N) in steel and may form AlN precipitate, which may cause slab cracks under a cast cooling conditions in which the precipitates precipitate during manufacturing of a slab, such that quality of a cast or a hot-rolled steel sheet may deteriorate. Also, when Al-rich inclusions or oxides are present in a steel sheet or a steel pipe, fatigue durability of a final component may deteriorate. Accordingly, it may be necessary to maintain the content thereof as low as possible. Accordingly, it may be preferable to limit the aluminum (Al) content to 0.04% or less (excluding 0%).

Cr: 0.3% or less (excluding 0%)

[0036] Chromium (Cr) may delay ferrite transformation of austenite, thereby increasing hardenability and improving heat treatment strength during quenching heat treatment of steel. When more than 0.3% of chromium (Cr) is added to steel including 0.30% or more of carbon (C), excessive hardenability of the steel may occur, and accordingly, the content thereof may be limited to 0.3% or less (excluding 0%).

Ni: 0.1-0.4%

[0037] Nickel (Ni) may simultaneously increase hardenability and toughness of steel. Meanwhile, in the present disclosure, when tensile properties and a maximum bending angle are evaluated after quenching-tempering heat treatment of a steel sheet or a steel pipe including an increased content of nickel (Ni) in a basic component, yield strength may decrease in accordance with the increase of Ni content after heat treatment, Accordingly, the Nickel (Ni) element may promote movement of dislocation included in into martensite or during heating heat treatment, prior austenite size may be suppressed from being excessively coarsened to an average of 26 um or more, and further, during tempering heat treatment, the size of carbide precipitating in a martensite structure may be controlled to less than 300 nm. However, when the content is less than 0.1%, the effect of lowering yield strength and increasing the maximum bending angle may be insufficient, whereas, when the content exceeds 0.4%, despite the above advantages, cost of manufacturing the steel sheet may increase rapidly. Accordingly, the content thereof may be limited to a 0.1- 0.4% range.

·Relational expression 2

[0038] The Mn and Ni may need to satisfy relational expression 2 as below.

[Relational expression 2]

$$(Ni/Mn) \geq 0.05 \text{ (weight ratio)}$$

[0039] The above (Ni/Mn) ratio may be a necessary condition to assure a maximum bending angle of 40° or more while assuring strength of 1300 MPa or more after quenching-tempering heat treatment. When the (Ni/Mn) ratio is less than 0.05, the ratio may be beyond the range of manganese (Mn) or nickel (Ni) content suggested in the present disclosure. Also, when the silicon (Si) content is low or the manganese (Mn) content is high, a bend structure including a high content of manganese (Mn) may be easily formed in a microstructure of a hot-rolled steel sheet, such that bendability may deteriorate after heating and quenching-tempering heat treatment. Meanwhile, during the heating heat treatment process, the nickel (Ni) element may segregate into a pearlitic-band structure or a grain boundary of iron carbide, may prevent complete decomposition or dissolution of carbide and may prevent prior austenite size from growing excessively. Also, during a tempering process, the element may segregate in the vicinity of carbide precipitating in martensite, may suppress growth of carbide and may allow the element to have a fine size. Accordingly, as a method of improving cold bendability of a steel sheet or a steel pipe after heating and quenching-tempering heat treatment, the (Ni/Mn) ratio may be limited to 0.05 or more in the present disclosure.

·Relational expression 3

[0040] The C, Mn, Si, and Ni may need to satisfy relational expression 3 as below.

[Relational expression 3]

$$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio)}$$

[0041] The above (Si+Ni)/(C+Mn) ratio may be a necessary condition to assure a maximum bending angle of 40° or more while assuring strength of 1300 MPa or more after quenching-tempering heat treatment. Generally, cold bendability of a steel sheet or a steel pipe after quenching-tempering heat treatment may appear to be inversely related to tensile strength. Meanwhile, in the present disclosure, relational expression 3 was devised to simultaneously satisfy high strength and high formability after heat treatment. When the (Si+Ni)/(C+Mn) ratio is less than 0.2, the (C+Mn) content may be higher than the (Si+Ni) content. In this case, hardenability may be high and strength after heat treatment may be extremely high, but the maximum bending angle may be extremely low, such that bending cracks may occur frequently during a bending process. Also, when precipitates forming elements are added, a prior austenite (PAGS) size may be less than 15 μm, which is a fine size, during heating treatment, such that tensile strength may be high and bendability may deteriorate.

Ti: 0.05% or less (excluding 0%)

[0042] Titanium (Ti) may form precipitates (TiC, TiCN, TiNbCN, etc.) in a hot-rolled steel sheet and may increase strength of the hot-rolled steel sheet by suppressing growth of austenite grains.

[0043] When the content exceeds 0.05%, quenching-tempering heat treatment may be effective in increasing strength of steel, but when the element is present in the form of coarsening crystals rather than fine precipitates in a hot-rolled steel sheet, toughness may degrade, or the element may act as a starting point for cracks during a cold bending process and may reduce cold-forming properties of a heat treatment steel sheet and a steel pipe component or may reduce fatigue durability of a final component. Accordingly, the content thereof may be limited to 0.05% or less (excluding 0%).

B: 0.0005-0.005% or less (excluding 0%)

[0044] Boron (B) may be beneficial by greatly increasing hardenability of steel even in a low content. When boron is added in an appropriate content, the element may be effective in increasing hardenability by suppressing ferrite formation, but when added in an excess content, austenite recrystallization temperature may increase and weldability may deteriorate. When the boron (B) content is less than 0.0005%, it may be difficult to assure the above effect of steel, and when the content exceeds 0.005%, the above effect may be saturated or it may be difficult to assure appropriate strength and toughness. Accordingly, the content thereof may be limited to 0.0005-0.005% or less. More preferably, limiting the content to 0.003% or less may be effective in simultaneously assuring strength and formability of heat treatment steel.

N: 0.01% or less (excluding 0%)

[0045] Nitrogen (N) may stabilize austenite and may form nitride. When the nitrogen (N) content exceeds 0.01%, coarsened AlN nitride may be formed, which may act as a starting point for formation of fatigue cracks when evaluating durability of a heat treatment steel sheet or a steel pipe component, such that fatigue durability may deterioate. Accordingly, the content thereof may be limited to 0.01% or less (excluding 0%). More preferably, it may be preferable to limit the content to 0.006% or less.

[0046] Also, when boron (B) element is added together, it may be necessary to control the nitrogen (N) content as low as possible to increase the effective boron (B) content.

[0047] The steel of the present disclosure may basically include the above components, and includes a balance of Fe and inevitable impurities, but allowable components as below may be optionally added to the extent that the present disclosure is not impaired.

Mo, Cu, Nb and V

[0048] In the present disclosure, one or more of Mo: 0.01-0.2%, Cu: 0.05-0.2%, Nb: 0.005-0.02% and V: 0.01-0.05% may be included optionally. These elements may increase hardenability of steel or may refine a grain size of prior austenite, thereby refining a lath size forming martensite or a tempered martensite structure of a final component. Accordingly, the element may contribute to increasing tensile strength of steel or further improving a bending angle or bendability.

[0049] Meanwhile, the hot-rolled steel sheet and the steel pipe in the present disclosure described above may have a microstructure including 20-65% of ferrite and 35-80% of pearlite by volume%. When the ferrite fraction is less than 20%, the pearlite content may excessively increases, such that it may be difficult to assure high strength and a bending angle due to the development of a bend structure. Accordingly, it may be preferable to limit the fraction of ferrite to 20% or more. The preferable ferrite fraction may be 20-60%. When the ferrite fraction exceeds 65%, the total amount of hardenability elements added to the hot-rolled steel sheet may be insufficient, and in this case, it may be difficult to assure sufficient strength after heating and quenching-tempering heat treatment.

[0050] Also, hardness in a surface portion and a thickness/4 position portion of the hot-rolled steel sheet in the present disclosure may have a hardness difference (ΔHv) value of less than 15, and the steel sheet may have tensile strength of 320-950 MPa.

[0051] When heating and quenching-tempering heat treatment is performed on the hot-rolled steel sheet, a forming member having, by volume%, a microstructure including 95% of one or more of martensite and tempered martensite, and a balance of 5% or less of retained austenite, and having high strength and a maximum bending angle of 40° or more, may be obtained.

[0052] Next, the method of manufacturing a hot-rolled steel sheet according to an embodiment of the present disclosure will be described.

[0053] The method of manufacturing a hot-rolled steel sheet in the present disclosure may include heating a steel slab having the above composition to a temperature range of 1150-1300°C; obtaining a hot-rolled steel sheet by hot-rolling the heated slab, including rough-rolling and finishing rolling, at a temperature of Ar_3 or more; and cooling the hot-rolled steel sheet on a run-out table and coiling the steel sheet at a temperature of 550-750°C.

Heating steel slab

[0054] First, in the present disclosure, the steel slab formed as above may be heated to a temperature range of 1150-1300°C.

[0055] The heating the steel slab to a temperature range of 1150-1300°C may be to ensure a uniform structure and component distribution in the slab. When the slab heating temperature is lower than 1150°C, precipitates formed on the continuous casting slab may not be solid-solute and component uniformity may not be assured. When the slab heating temperature exceeds 1300°C, excessive increase in decarburization depth and grain growth may occur, such that it may be difficult to assure a target material and surface quality of the hot-rolled steel sheet. Accordingly, the slab heating temperature may be limited to the range of 1150-1300°C.

Obtaining hot-rolled steel sheet

[0056] Thereafter, in the present disclosure, by hot-rolling the heated slab, including rough-rolling and finishing rolling, at a temperature of Ar_3 or more, a hot-rolled steel sheet may be obtained.

[0057] The hot-rolling may be preferably performed as hot-finish rolling at Ar_3 or higher. When the hot-rolling is performed at a temperature below Ar_3, austenite may be partially transformed into ferrite, and deformation resistance of the material for hot-rolling may become uneven, and strip or passing ability, including straightness of the steel sheet, may deteriorate, such that operational defects such as plate breakage may occur. However, when the finishing rolling temperature exceeds 950°C, scale defects may occur, such that it may be preferable to limit the hot-finishing rolling temperature to 950°C or less.

Coiling

[0058] In the present disclosure, the hot-rolled steel sheet obtained through hot-rolling as described above may be cooled on a run-out table and may be coiled at a temperature of 550-750°C.

[0059] The cooling on the run-out table after hot-rolling and coiling in the temperature range of 550-750°C may be to ensure uniform quality of the hot-rolled steel sheet. When the coiling temperature is extremely low below 550°C, a low-temperature transformation phase such as bainite or martensite may be formed in an edge portion of the steel sheet in a width direction, strength of the steel sheet may increase rapidly, and deviation in hot-rolling strength may increase in the width direction.

[0060] When a coiling temperature exceeds 750°C, internal oxidation may be promoted in a surface portion of the steel sheet, and surface defects such as cracks or surface unevenness may occur after hot-rolling picking. Also, coarsening of pearlite may cause deviation in surface hardness of the steel sheet. Accordingly, the coiling temperature after cooling of hot-rolled steel sheet may be preferably limited to 550-750°C.

[0061] In the present disclosure, the hot-rolled steel sheet manufactured as above may be manufactured into a hot-pickled steel sheet by further pickling the steel sheet. As the pickling method, a pickling method generally used in the hot-pickled process may be used, and accordingly, the specific method is not limited.

[0062] Meanwhile, in the present disclosure, in the coiling process, the coiling temperature in the range of 550-750°C may be controlled to be as low as possible, such that decarburization may be reduced, or the winding coil may be inserted in a water cooling bath rather than being naturally cooled, and the winding coil may be maintained at high temperature for a long time, or carrying out over pickling after cooling the winding coil to 200-250°C, such that the coil surface decarburization zone may be removed, it may be preferable to selectively use one of aforementioned processes. By selectively using these processes, hardness difference (ΔHv) between the surface portion and the t/4 position of the steel sheet may be effectively reduced to less than 15.

[0063] The hot-rolled steel sheet manufactured by the manufacturing process as described above in the present disclosure may have a microstructure including 20-65% of ferrite and 35-80% of pearlite by volume%.

[0064] Also, the hardness difference (ΔHv) value between the surface portion and the t/4 position portion of the hot-rolled steel sheet may be controlled to be less than 15, and may have tensile strength of 320-950 MPa.

[0065] Also, the hot-rolled steel sheet in the present disclosure manufactured through the above manufacturing process may have a microstructure including prior austenite having an average grain size of 15 μm or more.

[0066] When a hot-rolled steel sheet including prior austenite having an average grain size of 15 μm or more is gone through heating and quenching-tempering heat treatment for a manufacturing forming member described later, the formed martensite structure may be relatively coarsened and initiation of cracks during bending may be delayed, such that bendability may be improved or a bending angle may be increased.

[0067] In the case of a hot-rolled steel sheet satisfying relational expressions 2-3 described above in the present disclosure, the average grain size of prior austenite may be effectively controlled to 15 um or more, and accordingly, after quenching-tempering and heat treatment, appropriate yield strength or tensile strength and a high bending angle may be assured at the same time.

[0068] When the heating and quenching-tempering heat treatment for manufacturing the above-described forming member is performed using a hot-rolled steel sheet or a steel pipe including a prior austenite structure having an average grain size of less than 15 um, a martensite structure may be formed finely, such that strength may increase after heat treatment, and accordingly, cracks may occur in a short period of time during bending, such that bendability may decrease or the bending angle may decrease. Accordingly, in this case, there may be limitations in shape implementation when manufacturing a component through cold-forming.

[0069] Hereinafter, the method of manufacturing a steel pipe according to the embodiment of the present disclosure will be described.

[0070] A preferable method of manufacturing a steel pipe in the present disclosure may include obtaining a steel pipe by welding a hot-rolled steel sheet manufactured according to the method of manufacturing a hot-rolled steel sheet in the present disclosure; and performing annealing heat treatment on the steel pipe.

Obtaining steel pipe

[0071] By welding a hot-rolled steel sheet manufactured according to the method of manufacturing a hot-rolled steel sheet in the present disclosure described above, a steel pipe may be obtained.

[0072] Using the hot-rolled steel sheet or the hot-pickled steel sheet, for example, by forming pipes through electric arc welding or induction heating welding, a steel pipe may be obtained.

Performing annealing heat treatment on steel pipe

[0073] The steel pipe obtained as described above may go through annealing heat treatment.

[0074] In the present disclosure, using a general cold-forming method including a process of forming a steel pipe using the hot-rolled steel sheet or the hot-pickled steel sheet, for example, through electric arc welding or induction heating welding, and annealing heating and cold drawing, a small-diameter steel pipe may be manufactured.

[0075] The annealing heat treatment may be preferably performed on the steel pipe for 3-60 minutes at a temperature of Ac_1 -50°C-Ac_3 +150°C.

[0076] The annealing heat treatment may include furnace-cooling and air cooling.

[0077] In this case, in the present disclosure, a process of drawing the annealing heat treated steel pipe may be further included. A diameter of the steel pipe may be reduced by cold drawing the steel pipe. The drawing method may include a cold drawing method.

[0078] The steel pipe in the present disclosure, manufactured as described above may have a microstructure including 20-65% of ferrite and 35-80% of pearlite by volume%, and preferably, the microstructure of the steel pipe may include 20-50% of ferrite by volume%.

[0079] Thereafter, in the present disclosure, a reheating step of heating the cooled steel pipe or the drawn steel pipe to a temperature of Ar3-970°C at a heating rate of 10°C/sec or more and maintaining the steel pipe for less than 60 seconds; a quenching step of cooling the reheated steel pipe to room temperature at a cooling rate of 20-350°C/sec; and a tempering heat treatment step of heating the quenched steel pipe at a heating rate of 2-20°C/sec to a temperature range of 150-350°C and maintaining the steel pipe at this temperature may be further included, which will be described in greater detail below.

[0080] In the description below, the method of manufacturing a cold-forming member according to an embodiment of the present disclosure will be described.

[0081] The method of manufacturing a cold-forming member in the present disclosure may include reheating a steel pipe obtained according to the method of manufacturing a steel pipe; performing quenching-tempering heat treatment on the reheated steel pipe; and manufacturing a member by cold-forming the quenching-tempering heat treated steel pipe.

[0082] Before manufacturing the steel pipe into a chassis component such as a stabilizer by cold bending, first, heat treatment may be performed as below.

Reheating steel pipe

[0083] In the present disclosure, the annealing heat treated steel pipe or the drawn steel pipe may be reheated to manufacture a member for cold-forming use.

[0084] The reheating temperature may be Ar_3-970°C. In other words, a steel pipe having a specific length may be heated to a temperature in the target range at a heating rate of 10°C/sec or more while being passed through a high-frequency induction heating furnace at a moving rate of less than 100 mpm, and may be maintained under the condition of 60 seconds or less. In this case, the movement rate may be varied under the condition of less than 100mpm such that the inner wall of the steel pipe having a thickness of 3-6mm may have a uniform temperature. When heating swiftly to the target temperature at the heating rate of 10°C/sec or more, since the depth of a decarburization layer formed on the outer or inner wall of the steel pipe may be reduced, there may be an advantage of improving durability of a final component.

Performing quenching-tempering heat treatment on steel pipe

[0085] Thereafter, in the present disclosure, the reheated steel pipe may go through quenching-tempering heat treatment.

[0086] In the case of quenching cooling, the steel pipe may be cooled to room temperature by spraying water or oil to a specific length of the steel pipe heated to a temperature below 970°C. The cooling rate of the entire steel pipe may be varied from 20 to 350°C/sec, and is not limited to a specific range as long as the entire inner wall of the steel pipe may have a martensite structure. In this case, the cooling rate may be determined by appropriately controlling the amount of water or oil sprayed and the steel pipe movement rate.

[0087] The heated and quenched steel pipe may go through tempering heat treatment to provide toughness.

[0088] Under the tempering heat treatment temperature conditions, in the microstructure of steel, a tempered martensite structure formed in corresponding to a prior austenite grain size of 15 um or more may be formed as a main phase. Accordingly, during bending, resistance against grain boundary cracks along a prior austenite grain boundary may increase or plastic deformation may sufficiently occur without cracks, such that a bending angle of 40° or more may be exhibited.

[0089] When the heating temperature during tempering is less than 150°C, formation of the tempered martensite structure may be insufficient, or dislocation density may be high in martensite, and accordingly, strength may be relatively high after heat treatment, such that the bending angle may be low. When the heating temperature during tempering exceeds 350°C, a high bending angle may be assured due to the effect of excessive tempering of the martensite structure, but it may be difficult to assure strength of 1300MPa or more. Accordingly, the tempering heat treatment temperature may be preferably limited to a range of 150-350°C. More preferably, heat treatment may be performed in the temperature range of 200-250°C to avoid temper brittleness.

[0090] Meanwhile, under the heat treatment heating rate conditions, the steel microstructure may have Fe_3C carbides having various sizes in tempered martensite grains and at grain boundaries. When the tempering heat treatment heating rate is less than 2°C/sec, the heating rate may be extremely slow and excessive Fe_3C growth may occur due to the excessive tempering softening effect, such that strength may be low after heat treatment. Also, under the conditions as above, the heating rate may be low, such that steel pipe productivity may be low, which may be uneconomical. When the heating rate exceeds 20°C, Fe_3C growth in martensite may be suppressed and strength after heat treatment may tend to be excessively high, such that it may be difficult to ensure appropriate strength and a high bending angle at the same time. In other words, when the Fe_3C size is fine, crack sites including grain boundaries between Fe_3C and tempered martensite may be less, but resultant strength may be too high to ensure a high bending angle. Accordingly, the tempering heat treatment heating rate may be limited to the range of 2-20°C/sec, and it may be preferable to select the heating rate considering the heating temperature range. Considering the above circumstances, in the embodiment of the present disclosure, the number of Fe_3C carbides having an average circle equivalent size of 300 nm or less may be limited to 30 or less per unit area (μm^2).

<u>Cold-forming</u>

[0091] A member may be manufactured by cold-forming the steel pipe having undergone the reheating and quenching-tempering heat treatment as described above.

[0092] The forming of the steel pipe may be performed by cold-forming. For example, by room temperature forming may be performed on the heat treated steel pipe using a cold-forming device having dies of various bending radii (R). An example of the member may include a suspension component such as a stabilizer.

[0093] In the cold-forming of the steel pipe, it may be preferable to obtain a member by charging a steel pipe of a specific length into a die having a bending radius of 30-60R and performing minimum-maximum bending. In the present disclosure, the bending radius may refer to the degree to which a linear steel pipe is bent (curvature) and the radius of a curve formed by a circle, curvature radius. Accordingly, the bending radius 30R may indicate that the curvature radius may be small but a curvature or bending angle may be large, and the bending radius 60R may indicate that the curvature radius may be large but the curvature or bending angle may be small. Also, a bending radius of 60R may indicate bending relatively gently. Meanwhile, in the present disclosure, when a member may be manufactured by adjusting a bending rate and a friction coefficient between a die-steel pipe to a range in which bending cracks do not occur during the process of manufacturing the member, the specific range of the bending rate and the friction coefficient is not limited.

[0094] Meanwhile, the bending angle may be measured on a flat sheet rather than a steel pipe after heat treatment, and in the present disclosure, a maximum bending angle of the quenching-tempering heat treated flat sheet material having various thicknesses according to the VDA 238-100 standard test of the three-point bending test may be evaluated.

[0095] According to the method of manufacturing the member in the present disclosure, the member having both high strength and excellent cold bending formability after heat treatment, which may have tensile strength of 1300 MPa or more and a bending angle of 40° or more after heat treatment, or in which no cracks occur even at bending radius less than R50 may be manufactured.

[0096] As described above, in the present disclosure, to manufacture the member having both high strength and excellent cold bending formability after heat treatment, which may have tensile strength of 1300 MPa or more and a bending angle of 40° or more after heat treatment, or in which no cracks occur even at bending radius less than R50, it may be preferable to limit the average grain size of prior austenite to 15 um or more and the number of Fe_3C carbides having an average circle equivalent size of 300 nm or less to 30 or less per unit area (μm^2). Accordingly, appropriate strength, toughness or plastic deformation characteristics of tempered martensite structure steel including carbides of appropriate size in the tempered martensite structure may be assured, so that when bending a steel sheet or a steel pipe, resistance against grain boundary cracks along the grain boundary of prior austenite may increase, or sufficient plastic deformation may occur before cracks occur.

[0097] As described above, in the present disclosure, by performing heating and quenching-tempering heat treatment on the drawn steel pipe, strength of 1300-1800 MPa may be assured, and the ultra-high strength steel pipe may be componentized into a complex shape such as a stabilizer by cold-forming. The shape/performance of the cold-forming component of ultra-high strength steel pipe may have no significant difference as compared to the component manufactured through the general hot-forming process, but the shape of ultra-high strength steel pipe may be relatively simplified, and also, component manufacturing costs may be significantly reduced.

Mode for Invention

[0098] Hereinafter, the present disclosure will be described in greater detail through embodiments. However, it is important to note that the embodiment is only intended to further describe the present disclosure and is not intended to limit the scope of rights of present disclosure. This is because the scope of rights in present disclosure is determined by matters stated in the claims and matters reasonably inferred therefrom.

(Embodiment)

[0099]

[Table 1]

Steel type	C	Si	Mn	P	S	S.Al	Cr	Mo	Ni
Inventiv e steel 1	0.200	0.200	0.500	0.01	0.002	0.010	0.300	-	0.150
Inventiv e steel 2	0.335	0.167	0.998	0.0073	0.0007	0.035	0.090	0.10	0.180
Inventiv e steel 3	0.345	0.103	1.020	0.014	0.005	0.037	0.120	-	0.211
Comparat ive steel 1	0.335	0.098	0.990	0.014	0.005	0.025	-	-	0.001
Comparat ive steel 2	0.348	0.105	1.020	0.014	0.006	0.03	-	-	0.009
Inventiv e steel 4	0.337	0.103	1.010	0.014	0.005	0.037	0.150	-	0.207
Comparat ive steel 3	0.337	0.300	1.000	0.015	0.005	0.035	-	-	0.001
Comparat ive steel 4	0.340	0.303	0.900	0.014	0.005	0.025	-	-	0.001
Comparat ive steel 5	0.345	0.150	1.400	0.01	0.002	0.030	0.140	0.10	0.001
Comparat ive steel 6	0.385	0.200	1.400	0.01	0.002	0.030	0.150	-	0.001
Comparat ive steel 7	0.263	0.258	1.311	0.0071	0.0009	0.038	0.150	-	0.001
Comparat ive steel 8	0.338	0.164	1.28	0.0064	0.0008	0.036	0.140	0.10	0.001
Comparat ive steel 9	0.386	0.2	1.28	0.0063	0.0012	0.039	0.130	-	0.002
Comparat ive steel 10	0.041	0.099	1.3	0.01	0.002	0.03	0.190	0.09	0.001

[Table 2]

Steel type	Cu	Ti	Nb	V	B	N	Relationa 1 expressio n1	Relation al expressi on 2	Relation al expressi on 3
Inventive steel 1	0.10 0	0.020 0	-	-	0.002 5	0.004 0	2.5	0.300	0.500
Inventive steel 2	-	0.030 0	-	-	0.001 9	0.003 2	6.0	0.180	0.260
Inventive steel 3	-	0.005 4	-	-	0.002 4	0.004 2	9.9	0.207	0.230
Comparati ve steel 1	-	0.005 2	-	0.01 1	0.002 1	0.003 8	10.1	0.001	0.075
Comparati ve steel 2	-	0.005 2	0.008 5	0.01 2	0.002 3	0.001 6	9.7	0.009	0.083

(continued)

Steel type	Cu	Ti	Nb	V	B	N	Relationa l expressio n1	Relation al expressi on 2	Relation al expressi on 3
Inventive steel 4	-	0.0054	0.0088	0.012	0.0021	0.0018	9.8	0.205	0.230
Comparative steel 3	-	0.0056	-	0.011	-	0.0020	3.3	0.001	0.225
Comparative steel 4	-	0.0049	-	0.011	-	0.0024	3.0	0.001	0.245
Comparative steel 5	-	0.0300	-	-	0.0020	0.0060	9.3	0.001	0.087
Comparative steel 6	-	0.0300	-	-	0.0022	0.0070	7.0	0.001	0.113
Comparative steel 7	0.012	0.0370	0.0012	-	0.0031	0.0044	5.1	0.001	0.165
Comparative steel 8	0.024	0.0290	0.0020	-	0.0017	0.0037	7.8	0.001	0.102
Comparative steel 9	-	0.0300	-	-	0.0023	0.0047	6.4	0.002	0.121
Comparative steel 10	0.01	0.0290	0.0010	-	0.0019	0.0041	13.1	0.001	0.075

[0100] Using steel composed as in Table 1-2 above, a hot-rolled steel sheet having a thickness of 3.6 mm was manufactured by performing hot-rolling under the conditions in Table 3 below, and pickling was performed. Specifically, a slab or a lab manufacturing ingot having the composition in Table 1-2 above was homogenized by heating in the range of 1200 ± 20 °C for 200 minutes, and subsequently, the individual slab or ingot was rough-rolled and finishing rolled and was coiled at a temperature of 550-750°C, thereby manufacturing a hot-rolled steel sheet having a thickness of 3.6 mm.

[Table 3]

Steel type	Finishing rolling temperatu re (°C)	Coiling temperatu re (°C)	YS (MPa)	TS (MPa)	EL (%)	Average ΔHv	perlite fractio n (volume %)	Note
Inventive steel 1	880	640	332	524	31.2	8	35	Inventive example 1
Inventive steel 2	880	610	389	607	26.4	10	40	Inventive steel 2
Inventive steel 3	880	640	339	545	30.7	13	44	Inventive example 3
Inventive steel 3	880	580	369	594	29.3	14	57	Inventive example 3-1
Comparative steel 1	880	620	338	534	31.7	8	39	Comparative example 1
Comparati ve steel 2	880	660	387	597	27.2	22	47	Comparati ve example 2
Inventive steel 4	880	640	377	575	28.4	7	42	Inventive example 4

(continued)

Steel type	Finishing rolling temperatu re (°C)	Coiling temperatu re (°C)	YS (MP a)	TS (MP a)	EL (%)	Avera ge ΔHv	perlite fractio n (volume %)	Note
Inventive steel 4	870	580	440	657	26.1	8	76	Inventive example 4-1
Comparati ve steel 3	880	580	462	665	28.0	14	72	Comparati ve example 3
Comparati ve steel 4	880	570	459	672	25.9	18	73	Comparati ve example 4
Comparati ve steel 5	880	640	377	625	22.7	16	40	Comparati ve example 5
Comparati ve steel 6	880	640	367	620	23.1	15	38	Comparati ve example 6
Comparati ve steel 7	880	700	436	631	24.0	17	41	Comparati ve example 7
Comparati ve steel 8	860	640	521	756	23.1	16	47	Comparati ve example 8
Comparati ve steel 9	860	640	443	690	21.3	15	56	Comparati ve example 9
Comparati ve steel 10	880	650	458	700	22.0	16	63	Comparati ve example 10

[0101] For the hot-rolled steel sheet manufactured as above, yield strength (YS), tensile strength (TS), elongation (EL), Vickers hardness difference between surface portion and t/4, and a microstructure fraction were measured, and the results were listed in Table 3 above. The microstructure other than pearlite was ferrite. Meanwhile, in the embodiment, yield strength (YS), tensile strength (TS), and elongation (EL) of the hot-rolled steel sheet were measured using JIS 5 standards on samples taken in a direction parallel to the rolling direction, and the perlite fraction was measured using an image analysis program on nital-etched samples under the condition of an optical microscope at X500 magnification.

[0102] Also, a steel pipe having a diameter of 28 mm was manufactured using the hot-rolled steel sheets using electric resistance welding, and thereafter, a drawn steel pipe having a diameter of 22.2 mm was manufactured by performing annealing heat treatment and cold drawing. The steel pipe went through the heating-quenching-tempering heat treatment under the conditions in Table 4 below, and a member was manufactured through cold-forming.

[0103] In this case, in the quenching heat treatment, the steel pipe was heated to a temperature of 930-970 °C, and the steel pipe was cooled to 200 °C or less, and cooling was performed by inserting the steel pipe in water or oil or spraying water or oil to the steel pipe to cool the steel pipe to room temperature as much as possible. Also, tempering heat treatment was performed by heating the steel pipe to a temperature of 200-300°C at a heating rate in the range of 2-20°C/s, and cooling was performed.

[Table 4]

Steel type	Heating temperature(° C)	Cooling rate (°C/s)	Tempering temperature (° C)	Tempering rate (°C/s)	Note
Inventive steel 1	970	50	300	2	Inventive example 1
Inventive steel 2	950	40	220	3	Inventive example 2
Inventive steel 3	950	45	250	3	Inventive example 3

(continued)

Steel type	Heating temperature(° C)	Cooling rate (°C/s)	Tempering temperature (° C)	Tempering rate (°C/s)	Note
Inventive steel 3	950	40	220	15	Inventive example 3-1
Comparativ e steel 1	930	40	230	3	Comparativ e example 1
Comparativ e steel 2	930	40	230	3	Comparativ e example 2
Inventive steel 4	930	50	230	3	Inventive example 4
Inventive steel 4	950	50	200	5	Inventive example 4-1
Comparativ e steel 3	930	40	230	3	Comparativ e example 3
Comparativ e steel 4	930	40	230	3	Comparativ e example 4
Comparativ e steel 5	930	51	230	3	Comparativ e example 5
Comparative steel 6	930	50	230	3	Comparative example 6
Comparativ e steel 7	950	30	350	5	Comparativ e example 7
Comparativ e steel 8	930	42	300	7	Comparativ e example 8
Comparativ e steel 9	930	43	220	5	Comparativ e example 9
Comparativ e steel 10	930	40	250	5	Comparativ e example 10

[0104] After the quenching-tempering heat treatment, tensile properties and three-point bending tests were performed on the steel pipe, the microstructure was observed, and the results are listed in Table 5 below. Specifically, after the heating and quenching-heat treatment of the steel pipe, tensile properties of the steel sheet and a maximum bending angle were measured through a three-point bending test.

[0105] As for a grain average size of prior austenite, the cross-sectional surface of the same sample in which the optical microstructure was observed was polished and etched using picric acid, the grain size of at least 10 or more was measured at X500 magnification, and the average of the results was calculated. As for the number of Fe_3C carbide per unit area, using the same sample, the number of Fe3C carbides present within an area of width 2.9um × length 3.1um was measured at a magnification of X5,000-X10,000 using a scanning electron microscope, and the result was determined as the number of carbides per unit area. Specifically, the number of Fe_3C carbides present in tempered martensite grains was measured. After the heating and quenching-tempering heat treatment, the microstructure was measured in detail as above to determine the correlation between the microstructure changes and the bending angle.

[0106] Meanwhile, cold-forming was performed on the steel pipe by simply performing Zig-Zag forming using dies having various bending radii or using a cold-forming device. Bending of the steel pipe was performed under various conditions using a cold-forming device, whether cracks occurred and the minimum bending radius without cracks were observed, and the results are listed in Table 6 below. Here, mechanical properties of the heat treatment steel sheet were values measured by processing and performing heat treatment on samples taken in a direction parallel to the rolling direction according to the JIS 5 standard, and as for the maximum bending test value, to exclude sample edge properties while following the VDA 238-100 standard test, the edges of both the long sides of the heat treated sample were ground off.

[Table 5]

Steel type	YS (MPa)	TS (MPa)	EL (%)	YR	Maximum bending angle α (°)	TS×α	Prior austenite Average size (μm)	Number of Fe_3C carbide (/ μm^2)	Note
Inventive steel 1	1210	1327	9.8	0.9 1	63	83601	19	24	Inventive example 1
Inventive steel 2	1439	1733	8.8	0.8 3	61	104847	21	26	Inventive example 2
Inventive steel 3	1330	1626	9.4	0.8 2	49	79837	21	26	Inventive example 3
Inventive steel 3	1371	1667	8.4	0.8 2	53	88351	21	11	Inventive example 3-1
Comparative steel 1	1361	1647	9.1	0.8 3	38	61763	18	35	Comparative example 1
Comparative steel 2	1436	1726	8.9	0.83	38	65588	13	29	Comparative example 2
Inventive steel 4	1396	1685	9.3	0.8 3	46	77510	15	27	Inventive example 4
Inventive steel 4	1404	1684	9.1	0.8 3	51	85884	15	12	Inventive example 4-1
Comparative steel 3	1371	1672	6.2	0.8 2	38	63703	17	38	Comparative example 3
Comparative steel 4	1397	1691	6.3	0.8 3	39	65780	16	29	Comparative example 4
Comparative steel 5	1464	1819	8.7	0.8 0	37	67303	15	21	Comparative example 5
Comparative steel 6	1432	1805	8.6	0.7 9	36	64980	16	22	Comparative example 6
Comparative steel 7	1291	1365	9.3	0.9 5	53	72345	14	29	Comparative example 7
Comparative steel 8	1458	1616	8.8	0.9 0	43	69488	14	29	Comparative example 8
Comparative steel 9	1451	1889	10.0	0.7 7	39	73681	14	36	Comparative example 9
Comparative steel 10	1480	1829	8.6	0.8 2	38	69502	13	32	Comparative example 10

[Table 6]

Steel type	Minimum bending radius of steel pipe	Determination		Note
Inventive steel 1	45 (45R)	○		Inventive example 1
Inventive steel 2	48 (48R)	○		Inventive example 2
Inventive steel 3	48 (48R)	○		Inventive example 3
Inventive steel 3	48 (48R)	○		Inventive example 3-1
Comparative steel 1	55 (55R)	X	Cracks occurred	Comparative example 1

(continued)

Steel type	Minimum bending radius of steel pipe	Determination		Note
Comparative steel 2	55 (55R)	X	Cracks occurred	Comparative example 2
Inventive steel 4	50 (50R)	○		Inventive example 4
Inventive steel 4	50 (50R)	○		Inventive example 4-1
Comparative steel 3	55 (55R)	X	Cracks occurred	Comparative example 3
Comparative steel 4	55 (55R)	X	Cracks occurred	Comparative example 4
Comparative steel 5	55 (55R)	X	Cracks occurred	Comparative example 5
Comparative steel 6	55 (55R)	X	Cracks occurred	Comparative example 6
Comparative steel 7	50 (50R)	X	Cracks occurred	Comparative example 7
Comparative steel 8	50 (50R)	X	Cracks occurred	Comparative example 8
Comparative steel 9	50 (50R)	X	Cracks occurred	Comparative example 9
Comparative steel 10	50 (50R)	X	Cracks occurred	Comparative example 10

[0107] As indicated in Table 1-6 above, in inventive example 1-4, inventive example 3-1 and inventive example 4-1 manufactured using inventive steel 1-4 satisfying the steel composition and relational expressions 1-3, no bending cracks occurred even when the maximum bending angle exceeded 40° or the bending radius was 50R or less.

[0108] Also, inventive example 1-4, inventive example 3-1 and inventive example 4-1 had 1200-1400MPa of yield strength, 1300-1700MP of tensile strength, a yield ratio of 0.8 or more, and the maximum bending angle was 40° or more, which indicates that bending formability was excellent.

[0109] Also, in the case of hot-rolled steel sheets before heat treatment, the hardness difference value between the surface portion and the t/4 in the present disclosure examples was relatively small, less than 15, as compared to comparative examples 1-6. It may indicate that the hardness difference depending on the position in the thickness direction of the hot-rolled steel sheet was small, or that decarburization occurred less in the surface portion.

[0110] Conversely, in comparative examples 1-10, manufactured using comparative steel 1-10 not satisfying at least one of the alloy components and relational expressions 1-3 of the present disclosure, the maximum bending angle of the steel sheet after heat treatment was relatively less than 40° or the bending radius without cracks in the steel pipe was 55R or more.

[0111] Meanwhile, FIG. 1 is a diagram illustrating strength and maximum bending angle change curves of a steel sheet after tempering heat treatment of inventive example 2 in an embodiment of the present disclosure, and FIG. 2 is an image of a shape after cold bending of a quenching-tempering heat treated steel pipe of inventive example 2 in an embodiment of the present disclosure. FIG. 3(a-b) is an image indicating whether cracks occur in a heat treated steel pipe of an embodiment of the present disclosure, in which (a) indicates inventive example 2 and (b) indicates comparative example 2.

[0112] In other words, changes in strength and the maximum bending angle of the heat treated steel sheet of various embodiments of the inventive examples and the comparative examples were observed. As a representative example, the result of strength and the maximum bending angle of the steel sheet after tempering heat treatment of inventive example 2 were provided in FIG. 1.

[0113] Also, a bending test was performed on heat treated steel pipes of inventive steel and comparative steel of various embodiments using a cold-forming device having a 30-60R bending radius die, and as a representative example, results of the final shape of the steel pipe according to the bending test of inventive example 2 were provided in FIG. 2.

[0114] Also, when various steel pipes are cold bent, in the case in which the bending radius was small, cracks occurred in the surface portion of the heat treated steel pipe. As representative examples according to the presence or absence of cracks, inventive example 2 and comparative example 2 were provided in FIG. 3.

[0115] As indicated in FIG. 1, tensile strength of 1300MPa or more and a high bending angle of 40° or more were obtained at a tempering temperature of 200-250°C.

[0116] Also, as indicated in Figures 2-3, in the case of a heat treated steel pipe manufactured as an invention example satisfying the conditions of the present disclosure under the condition of bending radius 50R or less, the member may be manufactured without cracks occurring during cold-forming.

[0117] As described above, the reason why steel pipe bending cracks did not occur or the flat sheet 3-point bending angle was measured high after QT heat treatment in the present disclosure is that, since the steel type in the present disclosure had prior austenite grains having a relatively coarsening size, initiation of cracks was delayed due to external bending force, or during tempering heating, the Fe_3C carbide size was not sufficiently grown in the coarsened tempered martensite grain due to the application of a high heating rate. Also, the effect of delaying growth of Fe_3C is believed to be further promoted when nickel (Ni) element segregates at the interfacial surface adjacent to Fe_3C during tempering heating.

[0118] While example embodiments have been indicated 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 disclosure as defined by the appended claims.

Claims

1. A hot-rolled steel sheet for a cold-forming member, the hot-rolled steel sheet comprising:

   by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0. 004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities, and satisfying relational expressions 1-3, wherein a hardness difference value between a surface portion and a thickness/4 position portion of the steel sheet is less than 15, wherein, by volume%, the hot-rolled steel sheet has a microstructure including 20-65% of ferrite and 35-80% of pearlite, and
   wherein an average grain size of prior austenite is 15 um or more,

   [Relational expression 1]

   $$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

   [Relational expression 2]

   $$(Ni)/(Mn) \geq 0.05 \text{ (weight ratio)}$$

   [Relational expression 3]

   $$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio)}.$$

2. The hot-rolled steel sheet of claim 1, wherein one or more of Mo: 0.01-0.2%, Cu: 0.05-0.2%, Nb: 0.005-0.02% and V: 0.01-0.05% are included.

3. The hot-rolled steel sheet of claim 1, wherein the hot-rolled steel sheet has tensile strength of 320 to 950 MPa.

4. A method of manufacturing a hot-rolled steel sheet for a cold-forming member, the method comprising:

   heating a slab including, by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0.004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities and satisfying relational expressions 1-3 in a temperature range of 1150-1300°C;
   obtaining a hot-rolled steel sheet by hot-rolling the heated slab at an Ar_3 temperature or higher, the hot-rolling including rough-rolling and finishing rolling; and
   cooling the hot-rolled steel sheet on a run-out table and coiling the hot-rolled steel sheet at a temperature of 550 to 750°C,
   wherein the hot-rolled steel sheet has a hardness difference value of less than 15 between a surface portion and a thickness/4 position portion,

[Relational expression 1]

$$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

[Relational expression 2]

$$(Ni)/(Mn) \geq 0.05 \text{ (weight ratio)}$$

[Relational expression 3]

$$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio)}.$$

5. The method of claim 4, wherein one or more of Mo: 0.01-0.2%, Cu: 0.05-0.2%, Nb: 0.005-0.02% and V: 0.01-0.05% are included.

6. A steel pipe for a cold-forming member, the steel pipe comprising:
by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0. 004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities, satisfying relational expressions 1-3, having a hardness difference value of less than 15 between a surface portion and a thickness/4 position portion of the steel pipe, having, by volume%, a microstructure including 20-65% of ferrite and 35-80% of pearlite, and including prior austenite having an average grain size of 15 μm or more,

[Relational expression 1]

$$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

[Relational expression 2]

$$(Ni)/(Mn) \geq 0.05 \text{ (weight ratio)}$$

[Relational expression 3]

$$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio)}.$$

7. The steel pipe of claim 6, wherein one or more of Mo: 0.01-0.2%, Cu: 0.05-0.2%, Nb: 0.005-0.02% and V: 0.01-0.05% are included.

8. A method of manufacturing a steel pipe for a cold-forming member, the method comprising:

heating a slab including, by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0.004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities and satisfying relational expressions 1-3 in a temperature range of 1150-1300°C;
obtaining a hot-rolled steel sheet by hot-rolling the heated slab at an Ar_3 temperature or higher, the hot-rolling including rough-rolling and finishing rolling;
cooling the hot-rolled steel sheet on a run-out table and coiling the hot-rolled steel sheet at a temperature of 550 to 750°C;
obtaining a steel pipe by welding the hot-rolled steel sheet; and
performing annealing heat treatment on the steel pipe at a temperature of Ac_1 -50°C-Ac_3 +150°C for 3-60 minutes,

[Relational expression 1]

$$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

[Relational expression 2]

$$(Ni)/(Mn) \geq 0.05 \text{ (weight ratio)}$$

[Relational expression 3]

$$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio).}$$

9. The method of claim 8, wherein one or more of Mo: 0.01-0.2%, Cu: 0.05-0.2%, Nb: 0.005-0.02% and V: 0.01-0.05% are included.

10. The method of claim 8, further comprising:
drawing out the annealing-heat treated steel pipe.

11. The method of claim 10, further comprising:

a reheating step of heating the annealing-heat treated and drawn-out steel pipe to a temperature of Ar_3-970°C at a heating rate of 10°C/sec or more and maintaining the steel pipe for less than 60 seconds;
a quenching step of cooling the reheated steel pipe to room temperature at a cooling rate of 20-350°C/sec; and
a tempering heat treatment step of heating the quenched steel pipe at a heating rate of 2-20°C/sec to a temperature range of 150-350°C and maintaining the steel pipe at the above temperature.

12. A cold-forming member, comprising:
by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0. 004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities, satisfying relational expressions 1-3, and having, by volume%, a microstructure including 95% of one or more of martensite and tempered martensite and a balance of 5% or less of retained austenite, wherein an average grain size of prior austenite is 15 um or more, and the number of Fe_3C carbides having an average circle equivalent size of 300 nm or less is 30 or less per unit area (μm^2),

[Relational expression 1]

$$(Mn/Si) \geq 2 \text{ (weight ratio)}$$

[Relational expression 2]

$$(Ni)/(Mn) \geq 0.05 \text{ (weight ratio)}$$

[Relational expression 3]

$$(Si+Ni)/(C+Mn) \geq 0.2 \text{ (weight ratio)}$$

13. The cold-forming member of claim 12, wherein one or more of Mo: 0.01-0.2%, Cu: 0.05-0.2%, Nb: 0.005-0.02% and V: 0.01-0.05% are included.

14. The cold-forming member of claim 12, wherein the forming member has tensile strength of 1300MPa or more and a maximum bending angle of 40° or more.

15. A method of manufacturing a cold-forming member, the method comprising:

heating a slab including, by weight%, 0.20% or greater and less than 0.3% of C, 0.5-1.3% of Mn, 0.3% or less of Si (excluding 0%), 0.03% or less of P (including 0%), 0.004% or less of S (including 0%), 0.04% or less of Al (excluding 0%), 0.3% or less of Cr, 0.1-0.4% of Ni, 0.05% of Ti (including 0%), 0.0005-0.0050% of B, 0.01% or less of N (excluding 0%) and a balance of Fe and inevitable impurities and satisfying relational expressions 1-3 in a temperature range of 1150-1300°C;
obtaining a hot-rolled steel sheet by hot-rolling the heated slab at an Ar_3 temperature or higher, the hot-rolling including rough-rolling and finishing rolling;
cooling the hot-rolled steel sheet on a run-out table and coiling the hot-rolled steel sheet at a temperature of 550 to 750°C;
obtaining a steel pipe by welding the hot-rolled steel sheet;
performing annealing heat treatment and drawing out the steel pipe;
performing a reheating step of heating the annealing-heat treated or drawn steel pipe to a temperature of Ar_{3}-970°C at a heating rate of 10°C/sec or more and maintaining the steel pipe for less than 60 seconds;
performing a quenching step of cooling the reheated steel pipe to room temperature at a cooling rate of 20-350°C/sec more;
a tempering heat treatment step of heating the quenched steel pipe at a heating rate of 2-20°C/sec to a temperature range of 150-350°C and maintaining the steel pipe at the above temperature; and
cold-forming the tempering-heat treated steel pipe into a member.

16. The method of claim 12, wherein one or more of Mo: 0.01-0.2%, Cu: 0.05-0.2%, Nb: 0.005-0.02% and V: 0.01-0.05% are included.

17. The method of claim 15, wherein the forming member has tensile strength of 1300MPa or more and a maximum bending angle of 40° or more.

18. The method of claim 15, wherein annealing heat treatment is performed on the steel pipe for 3-60 minutes at a temperature of Ac_1 -50°C-Ac_3 +150°C.

19. The method of claim 15, wherein, during cold-forming the heat treated steel pipe, the cold-forming is performed in a range of bending radius of 30-60R.

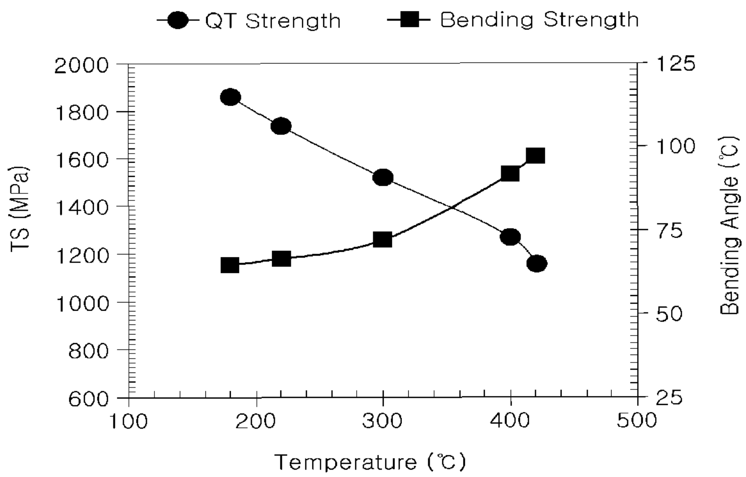

QT Strength
Bending Strength
TS (MPa)
Bending Angle (℃)
Temperature (℃)
2000
1800
1600
1400
1200
1000
800
600
125
100
75
50
25
100
200
300
400
500

FIG. 1

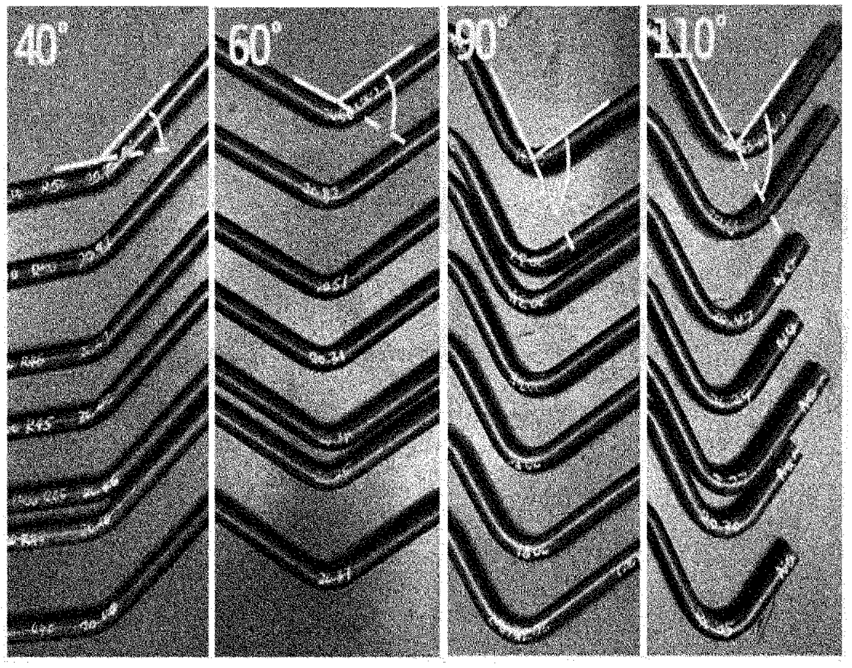
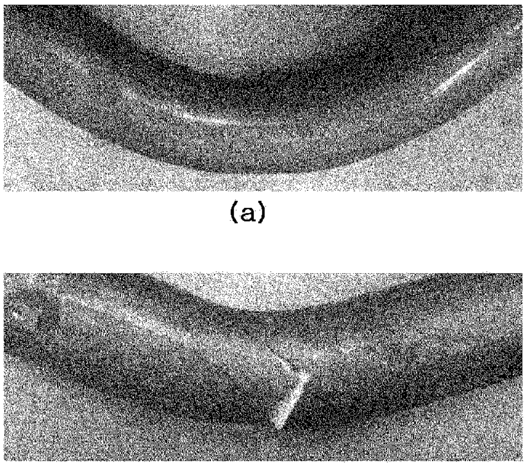


FIG. 2



(a)

(b)

FIG. 3

INTERNATIONAL SEARCH REPORT

International application No.
PCT/KR2022/020724

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/50(2006.01)i; **C22C 38/54**(2006.01)i; **C22C 38/00**(2006.01)i; **C22C 38/44**(2006.01)i; **C22C 38/42**(2006.01)i; **C22C 38/48**(2006.01)i; **C22C 38/46**(2006.01)i; **C21D 1/18**(2006.01)i; **C21D 8/02**(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 38/50(2006.01); C21D 8/02(2006.01); C21D 9/46(2006.01); C22C 38/00(2006.01); C22C 38/06(2006.01); C22C 38/42(2006.01); C22C 38/44(2006.01)

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: 냉간 성형(cold forming), 열연강판(hot rolling steel plate), 권취(winding), 인발(drawing), 퀜칭(quenching), 템퍼링(tempering)

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
Y	JP 2015-117406 A (NIPPON STEEL & SUMITOMO METAL) 25 June 2015 (2015-06-25) See claims 1-2.	4-5,8-10
A		1-3,6-7,11-19
Y	KR 10-2019-0078327 A (POSCO) 04 July 2019 (2019-07-04) See paragraph [0004]; and claims 4, 9-12 and 16-20.	4-5,8-10
Y	KR 10-2019-0077800 A (POSCO) 04 July 2019 (2019-07-04) See claims 3, 6-8 and 12-15.	4-5,8-10
A	KR 10-2012-0087619 A (HYUNDAI STEEL COMPANY) 07 August 2012 (2012-08-07) See claim 6.	1-19
A	JP 2011-102434 A (JFE STEEL CORP.) 26 May 2011 (2011-05-26) See claim 2.	1-19

☐ 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
- "D" document cited by the applicant in the international application
- "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
- "Y" document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art
- "&" document member of the same patent family

Date of the actual completion of the international search	Date of mailing of the international search report
22 March 2023	**22 March 2023**

Name and mailing address of the ISA/KR	Authorized officer
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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/KR2022/020724

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		EP 3733909 A1	04 November 2020
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		CN 111511953 B	10 December 2021
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		JP 7186229 B2	08 December 2022
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Form PCT/ISA/210 (patent family annex) (July 2022)

REFERENCES CITED IN THE DESCRIPTION

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Patent documents cited in the description

- KR 101568549 **[0008]**
- KR 1020150105476 **[0008]**