



(11) **EP 4 079 902 A1**

(12) **EUROPEAN PATENT APPLICATION**
published in accordance with Art. 153(4) EPC

(43) Date of publication:
26.10.2022 Bulletin 2022/43

(21) Application number: **20903825.6**

(22) Date of filing: **25.11.2020**

(51) International Patent Classification (IPC):
C22C 38/04 (2006.01) **C22C 38/02** (2006.01)
C22C 38/06 (2006.01) **C22C 38/00** (2006.01)
C21D 8/02 (2006.01) **C21D 9/68** (2006.01)

(52) Cooperative Patent Classification (CPC):
**C21D 8/02; C21D 9/68; C22C 38/00; C22C 38/02;
C22C 38/04; C22C 38/06**

(86) International application number:
PCT/KR2020/016838

(87) International publication number:
WO 2021/125603 (24.06.2021 Gazette 2021/25)

(84) Designated Contracting States:
**AL AT BE BG CH CY CZ DE DK EE ES FI FR GB
GR HR HU IE IS IT LI LT LU LV MC MK MT NL NO
PL PT RO RS SE SI SK SM TR**
Designated Extension States:
BA ME
Designated Validation States:
KH MA MD TN

(30) Priority: **18.12.2019 KR 20190169611**

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(54) **HIGH-STRENGTH STEEL SHEET HAVING SUPERIOR WORKABILITY, AND MANUFACTURING METHOD THEREFOR**

(57) The present invention relates to a steel sheet which can be used for automobile parts and the like, and relates to a steel sheet having a superior strength-ductility and strength-hole expansibility balance and superior bending workability, and a method for manufacturing same.

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Description

[Technical Field]

5 **[0001]** The present invention relates to a steel sheet that may be used for automobile parts and the like, and to a steel sheet having high strength characteristics and superior workability and a manufacturing method therefor.

[Background Art]

10 **[0002]** In recent years, the automobile industry is paying attention to ways to reduce material weight and secure occupant stability in order to protect the global environment. In order to meet these requirements for stability and weight reduction, the application of a high strength steel sheet is rapidly increasing. In general, it has been known that as the strength of the steel sheet increases, the workability of the steel sheet decreases. Therefore, in the steel sheet for automobile parts, a steel sheet having superior workability represented by ductility, bendability, and hole expandability
15 while having high strength characteristics is required.

[0003] As a technique for improving workability of a steel sheet, a method of utilizing tempered martensite is disclosed in Patent Documents 1 and 2. Since the tempered martensite made by tempering hard martensite is softened martensite, there is a difference in strength between the tempered martensite and the existing untempered martensite (fresh mar-
20 tensite). Therefore, when fresh martensite is suppressed and the tempered martensite is formed, the workability may be increased.

[0004] However, by the techniques disclosed in Patent Documents 1 and 2, a balance (TSXE1) of tensile strength and elongation does not satisfy 22,000 MPa% or more, meaning that it is difficult to secure a steel sheet having superior strength and ductility.

[0005] Meanwhile, transformation induced plasticity (TRIP) steel using transformation-induced plasticity of retained austenite was developed in order to obtain both high strength and superior workability for automobile member steel sheets. Patent Document 3 discloses TRIP steel having superior strength and workability.

[0006] Patent Document 3 discloses improving high ductility and workability by including polygonal ferrite, retained austenite, and martensite, but it can be seen that Patent Document 3 uses bainite as a main phase, and thus, the high strength is not secured and the balance (TSXE1) of the tensile strength and elongation also does not satisfy 22,000
30 MPa% or more.

[0007] That is, the demand for a steel sheet having superior workability, such as ductility, bendability, and hole expandability while having high strength, is not satisfied.

(Related Art Document)

35 **[0008]**

(Patent Document 1) Korean Patent Laid-Open Publication No. 10-2006-0118602

(Patent Document 2) Japanese Patent Laid-Open Publication No. 2009-019258

40 (Patent Document 3) Korean Patent Laid-Open Publication No. 10-2014-0012167

[Disclosure]

[Technical Problem]

45 **[0009]** The present invention provides a high-strength steel sheet having superior ductility, bendability, and hole expandability by optimizing a composition and microstructure of the steel sheet and a method for manufacturing the same.

[0010] An object of the present invention is not limited to the abovementioned contents. Additional problems of the present invention are described in the overall content of the specification, and those of ordinary skill in the art to which
50 the present invention pertains will have no difficulty in understanding the additional problems of the present invention from the contents described in the specification of the present invention.

[Technical Solution]

55 **[0011]** In an aspect of the present invention, a high-strength steel sheet having superior workability according may include: by wt%, C: 0.25 to 0.75%, Si: 4.0% or less, Mn: 0.9 to 5.0%, Al: 5.0% or less, P: 0.15% or less, S: 0.03% or less, N: 0.03% or less, a balance of Fe, and unavoidable impurities, and includes, as microstructures, ferrite which is a soft structure, and tempered martensite, bainite, and retained austenite which are hard structures, and may satisfy the

following [Relational Expression 1].

[Relational Expression 1]

$$0.4 \leq [H]_F / [H]_{TM+B+\gamma} \leq 0.9$$

[0012] In the above Relational Expression 1, $[H]_F$ and $[H]_{TM+B+\gamma}$ are nanohardness values measured using a nanoindenter, $[H]_F$ is an average nanohardness value H_v of the ferrite which is the soft structure, $[H]_{TM+B+\gamma}$ is the average nanohardness value H_v of the tempered martensite, the bainite, and the retained austenite which are the hard structures.

[0013] The steel sheet may further include one or more of the following (1) to (9).

- (1) one or more of Ti: 0 to 0.5%, Nb: 0 to 0.5%, and V: 0 to 0.5%
- (2) one or more of Cr: 0 to 3.0% and Mo: 0 to 3.0%
- (3) one or more of Cu: 0 to 4.5% and Ni: 0 to 4.5%
- (4) B: 0 to 0.005%
- (5) one or more of Ca: 0 to 0.05%, REM: 0 to 0.05% excluding Y, and Mg: 0 to 0.05%
- (6) one or more of W: 0 to 0.5% and Zr: 0 to 0.5%
- (7) one or more of Sb: 0 to 0.5% and Sn: 0 to 0.5%
- (8) one or more of Y: 0 to 0.2% and Hf: 0 to 0.2%
- (9) Co: 0 to 1.5%

[0014] A total content (Si+Al) of Si and Al may be 1.0 to 6.0 wt%

[0015] The microstructure of the steel sheet includes, by volume fraction, 30 to 70% of tempered martensite, 10 to 45% of bainite, 10 to 40% of retained austenite, 3 to 20% of ferrite, and unavoidable microstructure components.

[0016] A balance B_{T-E} of tensile strength and elongation expressed by the following [Relational Expression 2] is 22,000 (MPa%) or more, a balance B_{T-H} of tensile strength and a hole expansion ratio expressed by the following [Relational Expression 3] is 7×10^6 (MPa $^{2\%^{1/2}}$) or more, and bendability B_R expressed by the following [Relational Expression 4] may satisfy a range of 0.5 to 3.0.

[Relational Expression 2]

$$B_{T-E} = [\text{Tensile Strength (TS, MPa)}] * [\text{Elongation (El, \%)}]$$

[Relational Expression 3]

$$B_{T-H} = [\text{Tensile Strength (TS, MPa)}]^2 * [\text{Hole Expansion Ratio (HER, \%)}]^{1/2}$$

[Relational Expression 4]

$$B_R = R/t$$

[0017] In the above Relational Expression 4, R is a minimum bending radius (mm) at which cracks do not occur after a 90° bending test, and t is a thickness (mm) of the steel sheet.

[0018] In another aspect of the present invention, a manufacturing method for high-strength steel sheet having superior workability may include: heating and hot rolling a steel slab including, by wt%, C: 0.25 to 0.75%, Si: 4.0% or less, Mn: 0.9 to 5.0%, Al: 5.0% or less, P: 0.15% or less, S: 0.03% or less, N: 0.03% or less, a balance of Fe, and unavoidable impurities; coiling the hot-rolled steel sheet; performing a hot-rolled annealing heat treatment on the coiled steel sheet in a temperature within a range of 650 to 850°C for 600 to 1700 seconds; cold rolling the hot-rolled annealing heat-treated steel sheet; heating (primary heating) the cold-rolled steel sheet to a temperature within a range of Ac1 or higher

and less than Ac₃, and maintaining (primary maintaining) the cold-rolled steel sheet for 50 seconds or more; cooling (primary cooling) the primarily heated steel sheet to a temperature within a range of 100 to 300°C at an average cooling rate of 1°C/s or more; heating (secondary heating) the primarily cooled steel sheet to a temperature within a range of 300 to 500°C at an average temperature increase rate of 5°C/s or more, and maintaining (secondarily maintaining) the primarily cooled steel sheet for 50 seconds or more; and cooling (secondarily cooling) the secondarily heated steel sheet to room temperature.

[0019] The steel slab may further include one or more of the following (1) to (9).

- (1) one or more of Ti: 0 to 0.5%, Nb: 0 to 0.5%, and V: 0 to 0.5%
- (2) one or more of Cr: 0 to 3.0% and Mo: 0 to 3.0%
- (3) one or more of Cu: 0 to 4.5% and Ni: 0 to 4.5%
- (4) B: 0 to 0.005%
- (5) one or more of Ca: 0 to 0.05%, REM: 0 to 0.05% excluding Y, and Mg: 0 to 0.05%
- (6) one or more of W: 0 to 0.5% and Zr: 0 to 0.5%
- (7) one or more of Sb: 0 to 0.5% and Sn: 0 to 0.5%
- (8) one or more of Y: 0 to 0.2% and Hf: 0 to 0.2%
- (9) Co: 0 to 1.5%

[0020] A total content (Si+Al) of Si and Al included in the steel slab may be 1.0 to 6.0 wt%.

[0021] The steel slab may be heated to a temperature within a range of 1000 to 1350°C, and may be subjected to finish hot rolling in a temperature within a range of 800 to 1000°C.

[0022] The hot-rolled steel sheet may be coiled at a temperature within a range of 300 to 600°C.

[0023] A reduction ratio of the cold rolling may be 30 to 90%.

[0024] The cooling rate of the secondary cooling may be 1°C/s.

[Advantageous Effects]

[0025] According to an aspect of the present disclosure, it is possible to provide a steel sheet particularly suitable for automobile parts because the steel sheet has superior strength as well as superior workability such as ductility, bending workability, and hole expansion ratio.

[Best Mode]

[0026] The present invention relates to a high strength steel sheet having superior workability and a method for manufacturing the same, and exemplary embodiments in the present invention will hereinafter be described. Exemplary embodiments in the present invention may be modified into several forms, and it is not to be interpreted that the scope of the present invention is limited to exemplary embodiments described below. The present exemplary embodiments are provided in order to further describe the present invention in detail to those skilled in the art to which the present invention pertains.

[0027] The inventors of the present invention recognized that, in a transformation induced plasticity (TRIP) steel including bainite, tempered martensite, retained austenite, and ferrite, when controlling a ratio of specific components included in the retained austenite and the ferrite to a certain range while promoting stabilization of the retained austenite, it is possible to simultaneously secure workability and strength of a steel sheet by reducing a difference in hardness between phases of the retained austenite and the ferrite. Based thereon, the present inventors have reached the present invention by devising a method capable of improving ductility and workability of the high strength steel sheet.

[0028] Hereinafter, a high strength steel sheet having superior workability according to an aspect of the present invention will be described in more detail.

[0029] To an aspect of the present invention, a high-strength steel sheet having superior workability according may include: by wt%, C: 0.25 to 0.75%, Si: 4.0% or less, Mn: 0.9 to 5.0%, Al: 5.0% or less, P: 0.15% or less, S: 0.03% or less, N: 0.03% or less, a balance of Fe, and unavoidable impurities, and includes, as microstructures, ferrite which is a soft structure, and tempered martensite, bainite, and retained austenite which are hard structures, and may satisfy the following [Relational Expression 1].

[Relational Expression 1]

$$0.4 \leq [H]_F / [H]_{TM+B+\gamma} \leq 0.9$$

[0030] In the above Relational Expression 1, $[H]_F$ and $[H]_{TM+B+\gamma}$ are nanohardness values measured using a nanoindenter, $[H]_F$ is an average nanohardness value H_v of the ferrite which is the soft structure, $[H]_{TM+B+\gamma}$ is the average nanohardness value H_v of the tempered martensite, the bainite, and the retained austenite which are the hard structures.

[0031] Hereinafter, compositions of steel according to the present invention will be described in more detail. Hereinafter, unless otherwise indicated, % indicating a content of each element is based on weight.

[0032] The high strength steel sheet having superior workability according to an aspect of the present invention includes, by weight, C: 0.25 to 0.75%, Si: 4.0% or less, Mn: 0.9 to 5.0%, Al: 5.0% or less, P: 0.15% or less, S: 0.03% or less, N: 0.03% or less, a balance of Fe, and unavoidable impurities. In addition, the high strength steel sheet may further include one or more of Ti: 0.5% or less (including 0%), Nb: 0.5% or less (including 0%), V: 0.5% or less (including 0%), Cr: 3.0% or less (including 0%), Mo: 3.0% or less (including 0%), Cu: 4.5% or less (including 0%), Ni: 4.5% or less (including 0%), B: 0.005% or less (including 0%), Ca: 0.05% or less (including 0%), REM: 0.05% or less (including 0%) excluding Y, Mg: 0.05% or less (including 0%), W: 0.5% or less (including 0%), Zr: 0.5% or less (including 0%), Sb: 0.5% or less (including 0%), Sn: 0.5% or less (including 0%), Y: 0.2% or less (including 0%), Hf: 0.2% or less (including 0%), Co: 1.5% or less (including 0%). In addition, a total content (Si+Al) of Si and Al may be 1.0 to 6.0%.

Carbon(C): 0.25 to 0.75%

[0033] Carbon (C) is an essential element for securing strength of a steel sheet, and is also an element for stabilizing the retained austenite that contributes to the improvement in ductility of the steel sheet. Accordingly, the present invention may include 0.25% or more of carbon (C) to achieve such an effect. A preferable content of carbon (C) may exceed 0.25%, may be 0.27% or more, and may be 0.30% or more. The more preferable content of carbon (C) may be 0.31% or more. On the other hand, when the content of carbon (C) exceeds a certain level, cold rolling may become difficult due to an excessive increase in strength. Therefore, an upper limit of the content of carbon (C) of the present disclosure may be limited to 0.75%. The content of carbon (C) may be 0.70% or less, and the more preferable content of carbon (C) may be 0.67% or less.

Silicon (Si): 4.0% or less (excluding 0%)

[0034] Silicon (Si) is an element that contributes to improvement in strength by solid solution strengthening, and is also an element that improves workability by strengthening ferrite and homogenizing a structure. In addition, silicon (Si) is an element contributing to a generation of the retained austenite by suppressing precipitation of cementite. Therefore, in the present invention, silicon (Si) may be necessarily added to achieve such an effect. The preferable content of silicon (Si) may be 0.02% or more, and the more preferable content of silicon (Si) may be 0.05% or more. However, when the content of silicon (Si) exceeds a certain level, a problem of plating defects, such as non-plating, may be induced during plating, and weldability of a steel sheet may be lowered, so the present invention may limit the upper limit of the silicon (Si) content to 4.0%. The preferable upper limit of the content of silicon (Si) may be 3.8%, and the more preferable upper limit of the content of silicon (Si) may be 3.5%.

Aluminum (Al): 5.0% or less (excluding 0%)

[0035] Aluminum (Al) is an element that performs deoxidation by combining with oxygen in steel. In addition, aluminum (Al) is also an element for stabilizing the retained austenite by suppressing precipitation of cementite like silicon (Si). Therefore, in the present invention, aluminum (Al) may be necessarily added to achieve such an effect. A preferable content of aluminum (Al) may be 0.05% or more, and a more preferable content of aluminum (Al) may be 0.1% or more. On the other hand, when aluminum (Al) is excessively added, inclusions in a steel sheet increase, and the workability of the steel sheet may be lowered, so the present invention may limit the upper limit of the content of aluminum (Al) to 5.0%. The preferable upper limit of the content of aluminum (Al) may be 4.75%, and the more preferable upper limit of the content of aluminum (Al) may be 4.5%.

[0036] Meanwhile, the total content (Si+Al) of silicon (Si) and aluminum (Al) is preferably 1.0 to 6.0%. Since silicon (Si) and aluminum (Al) are components that affect microstructure formation in the present invention, and thus, affect ductility, bending formability, and hole expansion ratio, the total content of silicon (Si) and aluminum (Al) is preferably 1.0 to 6.0%. The more preferable total content (Si+Al) of silicon (Si) and aluminum (Al) may be 1.5% or more, and may be 4.0% or less.

Manganese (Mn): 0.9 to 5.0%

[0037] Manganese (Mn) is a useful element for increasing both strength and ductility. Therefore, in the present disclosure, a lower limit of a content of manganese (Mn) may be limited to 0.9% in order to achieve such an effect. A

preferable lower limit of the content of manganese (Mn) may be 1.0%, and a more preferable lower limit of the content of manganese (Mn) may be 1.1%. On the other hand, when manganese (Mn) is excessively added, the bainite transformation time increases and a concentration of carbon (C) in the austenite becomes insufficient, so there is a problem in that the desired austenite fraction may not be secured. Therefore, an upper limit of the content of manganese (Mn) of the present disclosure may be limited to 5.0%. A preferable upper limit of the content of manganese (Mn) may be 4.7%, and a more preferable upper limit of the content of manganese (Mn) may be 4.5%.

Phosphorus (P): 0.15% or less (including 0%)

[0038] Phosphorus (P) is an element that is included as an impurity and deteriorates impact toughness. Therefore, it is preferable to manage the content of phosphorus (P) to 0.15% or less.

Sulfur(S): 0.03% or less (including 0%)

[0039] Sulfur (S) is an element that is included as an impurity to form MnS in a steel sheet and deteriorate ductility. Therefore, the content of sulfur (S) is preferably 0.03% or less.

Nitrogen (N): 0.03% or less (including 0%)

[0040] Nitrogen (N) is an element that is contained as an impurity and forms nitride during continuous casting to causes cracks of slab. Therefore, the content of nitrogen (N) is preferably 0.03% or less.

[0041] Meanwhile, the steel sheet of the present invention has an alloy composition that may be additionally included in addition to the above-described alloy components, which will be described in detail below.

One or more of titanium (Ti): 0 to 0.5%, niobium (Nb): 0 to 0.5%, and vanadium (V): 0 to 0.5%

[0042] Titanium (Ti), niobium (Nb), and vanadium (V) are elements that make precipitates and refine crystal grains, and are elements that also contribute to the improvement in strength and impact toughness of a steel sheet, and therefore, in the present invention, one or more of titanium (Ti), niobium (Nb), and vanadium (V) may be added to achieve such an effect. However, when the content of titanium (Ti), niobium (Nb), and vanadium (V) exceed a certain level, respectively, excessive precipitates are formed to lower impact toughness and increase manufacturing cost, so the present invention may limit the content of titanium (Ti), niobium (Nb), and vanadium (V) to 0.5% or less, respectively.

One or more of chromium (Cr): 0 to 3.0% and molybdenum (Mo): 0 to 3.0%

[0043] Since chromium (Cr) and molybdenum (Mo) are elements that not only suppress austenite decomposition during alloying treatment, but also stabilize austenite like manganese (Mn), the present invention may add one or more of chromium (Cr) and molybdenum (Mo) to achieve such an effect. However, when the content of chromium (Cr) and molybdenum (Mo) exceeds a certain level, the bainite transformation time increases and the concentration of carbon (C) in austenite becomes insufficient, so the desired retained austenite fraction may not be secured. Therefore, the present invention may limit the content of chromium (Cr) and molybdenum (Mo) to 3.0% or less, respectively.

One or more of Cu: 0 to 4.5% and Ni: 0 to 4.5%

[0044] Copper (Cu) and nickel (Ni) are elements that stabilize austenite and suppress corrosion. In addition, copper (Cu) and nickel (Ni) are also elements that are concentrated on a surface of a steel sheet to prevent hydrogen from intruding into the steel sheet, to thereby suppress hydrogen delayed destruction. Accordingly, in the present invention, one or more of copper (Cu) and nickel (Ni) may be added to achieve such an effect. However, when the content of copper (Cu) and nickel (Ni) exceeds a certain level, not only excessive characteristic effects, but also an increase in manufacturing cost is induced, so the present invention may limit the content of copper (Cu) and nickel (Ni) to 4.5% or less, respectively.

Boron (B): 0 to 0.005%

[0045] Boron (B) is an element that improves hardenability to increase strength, and is also an element that suppresses nucleation of grain boundaries. Therefore, in the present invention, boron (B) may be added to achieve such an effect. However, when the content of boron (B) exceeds a certain level, not only excessive characteristic effects, but also an increase in manufacturing cost is induced, so the present invention may limit the content of boron (B) to 0.005% or less.

One or more of calcium (Ca): 0 to 0.05%, Magnesium (Mg): 0 to 0.05%, and rare earth element (REM) excluding yttrium (Y): 0 to 0.05%

[0046] Here, the rare earth element (REM) is scandium (Sc), yttrium (Y), and a lanthanide element. Since calcium (Ca), magnesium (Mg), and the rare earth element (REM) excluding yttrium (Y) are elements that contribute to the improvement in ductility of a steel sheet by spheroidizing sulfides, in the present invention, one or more of calcium (Ca), magnesium (Mg), and the rare earth element (REM) excluding yttrium (Y) may be added to achieve such an effect. However, when the content of calcium (Ca), magnesium (Mg), and the rare earth element (REM) excluding yttrium (Y) exceeds a certain level, not only excessive characteristic effects, but also an increase in manufacturing cost are induced, so the present invention may limit the content of calcium (Ca), magnesium (Mg), and the rare earth element (REM) excluding yttrium (Y) to 0.05% or less, respectively.

One or more of tungsten (W): 0 to 0.5% and zirconium (Zr): 0 to 0.5%

[0047] Since tungsten (W) and zirconium (Zr) are elements that increase strength of a steel sheet by improving hardenability, in the present invention, one or more of tungsten (W) and zirconium (Zr) may be added to achieve such an effect. However, when the content of tungsten (W) and zirconium (Zr) exceeds a certain level, not only excessive characteristic effects, but also an increase in manufacturing cost are induced, so the present invention may limit the content of tungsten (W) and zirconium (Zr) to 0.5% or less, respectively.

One or more of antimony (Sb): 0 to 0.5% and tin (Sn): 0 to 0.5%

[0048] Since antimony (Sb) and tin (Sn) are elements that improve plating wettability and plating adhesion of a steel sheet, in the present invention, one or more of antimony (Sb) and tin (Sn) may be added to achieve such an effect. However, when the content of antimony (Sb) and tin (Sn) exceeds a certain level, brittleness of a steel sheet increases, and thus, cracks may occur during hot working or cold working, so the present invention may limit the content of antimony (Sb) and tin (Sn) to 0.5% or less, respectively.

One or more of yttrium (Y): 0 to 0.2% and hafnium (Hf): 0 to 0.2%

[0049] Since yttrium (Y) and hafnium (Hf) are elements that improve corrosion resistance of a steel sheet, in the present invention, one or more of the yttrium (Y) and hafnium (Hf) may be added to achieve such an effect. However, when the content of yttrium (Y) and hafnium (Hf) exceeds a certain level, the ductility of the steel sheet may deteriorate, so the present invention may limit the content of yttrium (Y) and hafnium (Hf) to 0.2% or less, respectively.

Cobalt (Co): 0 to 1.5%

[0050] Since cobalt (Co) is an element that promotes bainite transformation to increase a TRIP effect, in the present invention, cobalt (Co) may be added to achieve such an effect. However, when the content of cobalt (Co) exceeds a certain level, since weldability and ductility of a steel sheet may deteriorate, the present invention may limit the content of cobalt (Co) to 1.5% or less.

[0051] The high strength steel sheet having superior workability according to an aspect of the present disclosure may include a balance of Fe and other unavoidable impurities in addition to the components described above. However, in a general manufacturing process, unintended impurities may inevitably be mixed from a raw material or the surrounding environment, and thus, these impurities may not be completely excluded. Since these impurities are known to those skilled in the art, all the contents are not specifically mentioned in the present specification. In addition, additional addition of effective components other than the above-described components is not entirely excluded.

[0052] The high strength steel sheet having superior workability according to an aspect of the present invention may include, as microstructures, ferrite which is a soft structure, and tempered martensite, bainite, and retained austenite which are hard structures. Here, the soft structure and the hard structure may be interpreted as a concept distinguished by a relative hardness difference.

[0053] As a preferred example, the microstructure of the high strength steel sheet having superior workability according to an aspect of the present invention may include, by volume fraction, 30 to 70% of tempered martensite, 10 to 45% of bainite, 10 to 40% of retained austenite, 3 to 20% of ferrite, and unavoidable microstructure components. As the unavoidable structure of the present invention, fresh martensite, perlite, martensite austenite constituent (M-A), and the like may be included. When the fresh martensite or the pearlite is excessively formed, the workability of the steel sheet may be lowered or the fraction of the retained austenite may be lowered.

[0054] The high strength steel sheet having superior workability according to an aspect of the present invention, as

shown in the following [Relational Expression 1], a ratio of an average nanohardness value ($[H]_F$, Hv) of the soft structure (ferrite) to an average nanohardness value ($[H]_{TM+B+\gamma}$, Hv) of the hard structure (tempered martensite, bainite, and retained austenite) may satisfy a range of 0.4 to 0.9.

[Relational Expression 1]

$$0.4 \leq [H]_F / [H]_{TM+B+\gamma} \leq 0.9$$

[0055] The nanohardness values of the hard and soft structures may be measured using a nanoindenter (FISCHER-SCOPE HM2000). Specifically, after electropolishing the surface of the steel sheet, the hard and soft structures are randomly measured at 20 points or more under the condition of an indentation load of 10,000 μ N, and the average nanohardness value of the hard and soft structures may be calculated based on the measured values.

[0056] In addition, in the high strength steel sheet having superior workability according to an aspect of the present invention, since a balance B_{T-E} of tensile strength and elongation expressed by the following [Relational Expression 2] is 22,000 (MPa%) or more, a balance B_{T-H} of tensile strength and hole expansion ratio expressed by the following [Relational Expression 3] is 7×10^6 (MPa $^2\%$ $^{1/2}$) or more, and bendability B_R expressed by the following [Relational Expression 4] satisfies a range of 0.5 to 3.0, it may have an superior balance of strength and ductility, a balance of strength and a hole expansion ratio, and superb bending formability.

[Relational Expression 2]

$$B_{T-E} = [\text{Tensile Strength (TS, MPa)}] * [\text{Elongation (El, \%)}]$$

[Relational Expression 3]

$$B_{T-H} = [\text{Tensile Strength (TS, MPa)}]^2 * [\text{Hole Expansion Ratio (HER, \%)}]^{1/2}$$

[Relational Expression 4]

$$B_R = R/t$$

[0057] In the above Relational Expression 4, R is a minimum bending radius (mm) at which cracks do not occur after a 90° bending test, and t is a thickness (mm) of the steel sheet.

[0058] In the present invention, it is important to stabilize retained austenite of a steel sheet because it is intended to simultaneously secure superb ductility and bending formability as well as high strength properties. In order to stabilize the retained austenite, it is necessary to concentrate carbon (C) and manganese (Mn) in the ferrite, bainite, and tempered martensite of the steel sheet into austenite. However, when carbon (C) is concentrated into austenite by using ferrite, the strength of the steel sheet may be insufficient due to the low strength characteristics of ferrite, and excessive inter-phase hardness difference may occur, thereby reducing the hole expansion ratio (HER). Therefore, it is intended to concentrate carbon (C) and manganese (Mn) into austenite by using the bainite and tempered martensite.

[0059] When the content of silicon (Si) and aluminum (Al) in the retained austenite is limited to a certain range, carbon (C) and manganese (Mn) may be concentrated in large amounts from bainite and tempered martensite into retained austenite, thereby effectively stabilizing the retained austenite. In addition, by limiting the content of silicon (Si) and aluminum (Al) in austenite to a certain range, it is possible to increase the content of silicon (Si) and aluminum (Al) in ferrite. As the content of silicon (Si) and aluminum (Al) in the ferrite increases, the hardness of the ferrite increases, so it is possible to effectively reduce an inter-phase hardness difference between ferrite which is a soft structure and tempered martensite, bainite, and retained austenite which are a hard structure.

[0060] When the ratio of the average nanohardness value ($[H]_F$, Hv) of the soft structure (ferrite) to the average

nanohardness value ($[H]_{TM+B+\gamma}$, Hv) of the hard structure (tempered martensite, bainite, and retained austenite) is greater than a certain level, the inter-phase hardness difference between the soft structure (ferrite) and the hard structure (tempered martensite, bainite, and retained austenite) is lowered, so it is possible to secure a desired balance (TSXE1) of tensile strength and elongation, a balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio, and bendability (R/t). On the other hand, when the ratio of the average nanohardness value ($[H]_F$, Hv) of the soft structure (ferrite) to the average nanohardness value ($[H]_{TM+B+\gamma}$, Hv) of the hard structure (tempered martensite, bainite, and retained austenite) is excessive, the ferrite is excessively hardened and the workability is rather lowered, so the desired balance (TSXE1) of tensile strength and elongation, the balance of tensile strength and hole expansion ratio ($TS^2 \times HER^{1/2}$), and the bendability (R/t) may not all be secured. Therefore, the present invention may limit the ratio of the average nanohardness value ($[H]_F$, Hv) of the soft structure to the average nanohardness value ($[H]_{TM+B+\gamma}$, Hv) of the hard structure (tempered martensite, bainite, and retained austenite) to a range of 0.4 to 0.9.

[0061] A steel sheet including retained austenite has superb ductility and bending formability due to transformation-induced plasticity that occurs during transformation from austenite to martensite during processing. When the fraction of the retained austenite is less than a certain level, the balance (TSXE1) of tensile strength and elongation may be less than 22,000 MPa%, or the bendability (R/t) may exceed 3.0. Meanwhile, when the fraction of the retained austenite exceeds a certain level, local elongation may be lowered. Accordingly, in the present invention, the fraction of the retained austenite may be limited to a range of 10 to 40 vol% in order to obtain a steel sheet having a balance (TSXE1) of tensile strength and elongation and superb bendability (R/t).

[0062] Meanwhile, both untempered martensite (fresh martensite) and tempered martensite are microstructures that improve the strength of the steel sheet. However, compared with the tempered martensite, fresh martensite has a characteristic of greatly reducing the ductility and the hole expansion ratio of the steel sheet. This is because the microstructure of the tempered martensite is softened by the tempering heat treatment. Therefore, in the present invention, it is preferable to use tempered martensite to provide a steel sheet having a balance of strength and ductility, a balance of strength and hole expansion ratio, and superb workability. When the fraction of the tempered martensite is less than a certain level, it is difficult to secure the balance (TSXE1) of tensile strength and elongation of 22,000 MPa% or more or the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio of 7×10^6 (MPa^{2%})^{1/2} or more, and when the fraction of the tempered martensite exceeds a certain level, ductility and workability is lowered, and the balance (TSXE1) of tensile strength and elongation is less than 22,000 MPa%, or bendability (R/t) exceeds 3.0, which is not preferable. Therefore, in the present invention, the fraction of the tempered martensite may be limited to 30 to 70 vol% to obtain a steel sheet having the balance (TSXE1) of tensile strength and elongation, the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio, and superb bendability (R/t).

[0063] In order to improve the balance (TS×EI) of tensile strength and elongation, the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio, and the bendability (R/t), it is preferable that bainite is appropriately included as the microstructure. As long as a fraction of bainite is a certain level or more, it is possible to secure the balance (TS×EI) of tensile strength and elongation of 22,000 MPa% or more, the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio of 7×10^6 (MPa^{2%})^{1/2} or more and the bendability (R/t) of 0.5 to 3.0. On the other hand, when the fraction of bainite is excessive, the decrease in the fraction of tempered martensite is necessarily accompanied, so the present invention may not secure the desired balance (TS×EI) of tensile strength and elongation, the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio, and bendability (R/t). Accordingly, the present invention may limit the fraction of bainite to a range of 10 to 45 vol%.

[0064] Since ferrite is an element contributing to improvement in ductility, the present invention may secure the desired balance (TS×EI) of tensile strength and elongation, as long as the fraction of ferrite is a certain level or more. However, when the fraction of ferrite is excessive, the inter-phase hardness difference increases and the hole expansion ratio (HER) may decrease, so the present invention may not secure the desired balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio. Accordingly, the present invention may limit the fraction of ferrite to a range of 3 to 20 vol%.

[0065] Hereinafter, an example of a method for manufacturing a steel sheet of the present invention will be described in detail.

[0066] A manufacturing method for high-strength steel sheet according to an aspect of the present invention may include: preparing a steel slab having a predetermined component; heating and hot rolling the steel slab; coiling the hot-rolled steel sheet; performing a hot-rolled annealing heat treatment on the coiled steel sheet in a temperature within a range of 650 to 850°C for 600 to 1700 seconds; cold rolling the hot-rolled annealing heat-treated steel sheet; heating (primary heating) the cold-rolled steel sheet to a temperature within a range of Ac1 or higher and less than Ac3, and maintaining (primary maintaining) the cold-rolled steel sheet for 50 seconds or more; cooling (primary cooling) the primarily heated steel sheet to a temperature within a range of 100 to 300°C at an average cooling rate of 1°C/s or more; heating (secondary heating) the primarily cooled steel sheet to a temperature within a range of 300 to 500°C at an average temperature increase rate of 5°C/s or more, and maintaining (secondarily maintaining) the primarily cooled steel sheet for 50 seconds or more; and cooling (secondarily cooling) the secondarily heated steel sheet to room temperature.

Preparation and heating of steel slab

[0067] A steel slab having a predetermined component is prepared. Since the steel slab according to the present invention includes an alloy composition corresponding to an alloy composition of the steel sheet described above, the description of the alloy compositions of the slab is replaced by the description of the alloy composition of the steel sheet described above.

[0068] The prepared steel slab may be heated to a temperature within a certain temperature range, and the heating temperature of the steel slab at this time may be in the range of 1000 to 1350°C. This is because, when the heating temperature of the steel slab is less than 1000°C, the steel slab may be hot rolled in the temperature range below the desired finish hot rolling temperature range, and when the heating temperature of the steel slab exceeds 1350°C, the temperature reaches a melting point of steel, and thus, the steel slab is melted.

Hot rolling and coiling

[0069] The heated steel slab may be hot rolled, and thus, provided as a hot-rolled steel sheet. During the hot rolling, the finish hot rolling temperature is preferably in the range of 800 to 1000°C. When the finish hot rolling temperature is less than 800°C, an excessive rolling load may be a problem, and when the finish hot rolling temperature exceeds 1000°C, grains of the hot-rolled steel sheet are coarsely formed, which may cause a deterioration in physical properties of the final steel sheet.

[0070] The hot-rolled steel sheet after the hot rolling has been completed may be cooled at an average cooling rate of 10°C/s or more, and may be coiled at a temperature of 300 to 600°C. When the coiling temperature is less than 300°C, the coiling is not easy, and when the coiling temperature exceeds 600°C, a surface scale is formed to the inside of the hot-rolled steel sheet, which may make pickling difficult.

Hot-rolled annealing heat treatment

[0071] It is preferable to perform a hot rolling annealing heat treatment process in order to facilitate pickling and cold rolling, which are subsequent processes after the coiling. The hot rolling annealing heat treatment may be performed in a temperature within a range of 650 to 850°C for 600 to 1700 seconds. When the hot rolling annealing heat treatment temperature is less than 650°C or the hot rolling annealing heat treatment time is less than 600 seconds, the strength of the hot rolling annealing heat-treated steel sheet increases, and thus, subsequent cold rolling may not be easy. On the other hand, when the hot rolling annealing heat treatment temperature exceeds 850°C or the hot rolling annealing heat treatment time exceeds 1700 seconds, the pickling may not be easy due to a scale formed deep inside the steel sheet.

Pickling and cold rolling

[0072] After the hot rolling annealing heat treatment, in order to remove the scale generated on the surface of the steel sheet, the pickling may be performed, and the cold rolling may be performed. Although the conditions of the pickling and cold rolling are not particularly limited in the present invention, the cold rolling is preferably performed at a cumulative reduction ratio of 30 to 90%. When the cumulative reduction ratio of the cold rolling exceeds 90%, it may be difficult to perform the cold rolling in a short time due to the high strength of the steel sheet.

[0073] The cold-rolled steel sheet may be manufactured as a non-plated cold-rolled steel sheet through the annealing heat treatment process, or may be manufactured as a plated steel sheet through a plating process to impart corrosion resistance. As the plating, plating methods such as hot-dip galvanizing, electro-galvanizing, and hot-dip aluminum plating may be applied, and the method and type are not particularly limited.

Annealing heat treatment

[0074] In the present invention, in order to simultaneously secure the strength and workability of the steel sheet, the annealing heat treatment process is performed.

[0075] The cold-rolled steel sheet is heated (primarily heated) to a temperature within a range of Ac1 or higher and less than Ac3 (two-phase region), and held (primarily held) in the temperature range for 50 seconds or more. The primary heating or primary maintaining temperature is Ac3 or higher (single-phase region), the desired ferrite structure may not be realized, so the desired level of $[H]_F/[H]_{TM+B+\gamma}$, and the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio may be implemented. In addition, when the primary heating or primary maintaining temperature is in a temperature range less than Ac1, there is a fear that sufficient heating is not made, and thus, the microstructure desired by the present invention may not be implemented even by subsequent heat treatment. The average temperature increase rate of the primary heating may be 5°C/s or more.

[0076] When the primary maintaining time is less than 50 seconds, the structure may not be sufficiently homogenized and the physical properties of the steel sheet may be lowered. The upper limit of the primary maintaining time is not particularly limited, but the primary heating time is preferably limited to 1200 seconds or less in order to prevent the decrease in toughness due to the coarsening of grains.

[0077] After the primary maintaining, the primarily heated steel sheet may be cooled (primarily cooled) to a primary cooling stop temperature of 100 to 300°C at a primary cooling rate of an average cooling rate of 1°C/s or more. The upper limit of the primary cooling rate does not need to be particularly specified, but is preferably limited to 100°C/s or less. When the primary cooling stop temperature exceeds 100°C, the tempered martensite is excessively formed and the amount of retained austenite formed is insufficient, so $[H]_F/[H]_{TM+B+\gamma}$, the balance (TSXE1) of tensile strength and elongation, and the bendability (R/t) may be lowered. On the other hand, when the primary cooling stop temperature exceeds 300°C, the bainite is excessively formed and the amount of tempered martensite formed is insufficient, so the balance (TSXE1) of tensile strength and elongation of the steel sheet, and the balance (TS²XHER^{1/2}) of tensile strength and hole expansion ratio of the steel sheet may be lowered.

[0078] After the primary cooling, the primarily cooled steel sheet may be heated (secondarily heated) to a secondary heating temperature of 300 to 500°C at a secondary heating rate of an average temperature increase rate of 5°C/s or more, and may be maintained (secondarily maintained) for 50 seconds or more in the temperature range. The upper limit of the secondary temperature increase rate does not need to be particularly specified, but is preferably limited to 100°C/s or less. When the secondary heating or secondary maintaining temperature is less than 300°C, or the maintaining time is less than 50 seconds, the tempered martensite is excessively formed and the control of Si and Al content in the retained austenite is insufficient, so the desired fraction of the retained austenite is difficult to obtain. As a result, $[H]_F/[H]_{TM+B+\gamma}$, the balance (TSXE1) of tensile strength and elongation and the bending workability (R/t) may be lowered. On the other hand, when the secondary heating or maintaining temperature exceeds 500°C or the secondary maintaining time is 172,000 seconds or more, it is difficult to secure the fraction of the retained austenite because the control of Si and Al content in the retained austenite is insufficient. As a result, $[H]_F/[H]_{TM+B+\gamma}$, and the balance (TSXE1) of tensile strength and elongation may be lowered.

[0079] After the secondary maintaining, it is preferable to cool (secondarily cool) the secondarily heated steel sheet to room temperature at an average cooling rate of 1°C/s or more.

[0080] The high strength steel sheet having superior workability manufactured by the above-described manufacturing method may include, as a microstructure, tempered martensite, bainite, retained austenite, and ferrite, and as a preferred example, may include, by the volume fraction, 30 to 70% of tempered martensite, 10 to 45% of bainite, 10 to 40% of retained austenite, 3 to 20% of ferrite, and unavoidable structures.

[0081] In addition, in the high-strength steel sheet having superior workability manufactured by the above-described manufacturing method, as in the following [Relational Expression 1], the ratio of the average nanohardness value ($[H]_F$, Hv) of the soft structure (ferrite) to the average nanohardness value ($[H]_{TM+B+\gamma}$, Hv) of the hard structures (tempered martensite, bainite, and retained austenite) may satisfy the range of 0.4 to 0.9, the balance B_{T-E} of tensile strength and elongation expressed by the following [Relational Expression 2] is 22,000 (MPa%) or more, and the balance B_{T-H} of tensile strength and elongation expressed by the following [Relational Expression 3] may be 7×10^6 (MPa²%^{1/2}) or more, and the bendability B_A expressed by the following [Relational Expression 4] may satisfy the range of 0.5 to 3.0.

[Relational Expression 1]

$$0.4 \leq [H]_F / [H]_{TM+B+\gamma} \leq 0.9$$

[Relational Expression 2]

$$B_{T-E} = [\text{Tensile Strength (TS, MPa)}] * [\text{Elongation (EL, \%)}]$$

[Relational Expression 3]

$$B_{T-H} = [\text{Tensile Strength (TS, MPa)}]^2 * [\text{Hole Expansion Ratio (HER, \%)}]^{1/2}$$

[Relational Expression 4]

$$B_R = R/t$$

[0082] In the above Relational Expression 4, R is a minimum bending radius (mm) at which cracks do not occur after a 90° bending test, and t is a thickness (mm) of the steel sheet.

[Mode for Invention]

[0083] Hereinafter, a high strength steel sheet having superior workability and a method for manufacturing same according to an aspect of the present invention will be described in more detail. It should be noted that the following examples are only for the understanding of the present invention, and are not intended to specify the scope of the present invention. The scope of the present invention is determined by matters described in claims and matters reasonably inferred therefrom.

(Inventive Example)

[0084] A steel slab having a thickness of 100 mm having alloy compositions (a balance of Fe and unavoidable impurities) shown in Table 1 below was prepared, heated at 1200°C, and then was subjected to finish hot rolling at 900°C. Thereafter, the steel slab was cooled at an average cooling rate of 30°C/s, and coiled at a coiling temperature of Tables 2 and 3 to manufacture a hot-rolled steel sheet having a thickness of 3 mm. The hot-rolled steel sheet was subjected to hot rolling annealing heat treatment under the conditions of Tables 2 and 3. Thereafter, after removing a surface scale by pickling, cold rolling was performed to a thickness of 1.5 mm.

[0085] Thereafter, the heat treatment was performed under the annealing heat treatment conditions disclosed in Tables 2 to 5 to manufacture the steel sheet.

[0086] The microstructure of the thus prepared steel sheet was observed, and the results were shown in Tables 6 and 7. Among the microstructures, ferrite (F), bainite (B), tempered martensite (TM), and pearlite (P) were observed through SEM after nital-etching a polished specimen cross section. The fractions of bainite and tempered martensite, which are difficult to distinguish among them, were calculated using an expansion curve after evaluation of dilatation. Meanwhile, since fresh martensite (FM) and retained austenite (retained γ) are also difficult to distinguish, a value obtained by subtracting the fraction of retained austenite calculated by X-ray diffraction method from the fraction of martensite and retained austenite observed by the SEM was determined as the fraction of the fresh martensite.

[0087] Meanwhile, $[H]_F/[H]_{TM+B+\gamma}$, a balance (TSXE1) of tensile strength and elongation, a balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio, and bendability (R/t) were observed, and the results were shown in Tables 8 and 9.

[0088] Tensile strength (TS) and elongation (EI) were evaluated through a tensile test, and the tensile strength (TS) and the elongation (EI) were measured by evaluating the specimens collected in accordance with JIS No. 5 standard based on a 90° direction with respect to a rolling direction of a rolled sheet. The bendability (R/t) was evaluated by a V-bending test, and calculated by collecting a specimen based on the 90° direction with respect to the rolling direction of the rolled sheet and was determined and calculated as a value obtained by dividing a minimum bending radius R (mm), at which cracks do not occur after a 90° bending test, by a thickness t (mm) of a sheet. The hole expansion ratio (HER) was evaluated through the hole expansion test, and was calculated by the following [Relational Expression 5] by, after forming a punching hole (die inner diameter of 10.3mm, clearance of 12.5%) of 10 mm $\varnothing$, inserting a conical punch having an apex angle of 60° into a punching hole in a direction in which a burr of a punching hole faces outward, and then compressing and expanding a peripheral portion of the punching hole at a moving speed of 20 mm/min.

[Relational Expression 5]

$$\text{Hole Expansion Ratio (HER, \%)} = \left\{ (D - D_0) / D_0 \right\} \times 100$$

[0089] In the above Relational Expression 5, D is a hole diameter (mm) when cracks penetrate through the steel plate along the thickness direction, and D_0 is the initial hole diameter (mm).

[0090] Nanohardness values of hard and soft structures were measured using the nanoindentation method. Specifically, after electropolishing surfaces of each specimen, the hard and soft structures were randomly measured at 20

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points or more under the condition of an indentation load of 10,000 μN using a nanoindenter (FISCHERSCOPE HM2000), and the average nanohardness value of the hard and soft structures was calculated based on the measured values.

[Table 1]

Steel Type	Chemical Component (wt%)									
	C	Si	Mn	P	S	Al	N	Cr	Mo	Others
A	0.33	1.77	2.13	0.008	0.0011	0.45	0.0025	0.53		
B	0.34	2.52	2.26	0.013	0.0009	0.51	0.0033	0.31	0.24	
C	0.33	2.28	2.15	0.009	0.0011	0.44	0.0024		0.53	
D	0.35	2.47	3.58	0.012	0.0012	0.63	0.0023		0.49	
E	0.41	1.59	2.45	0.009	0.0008	0.49	0.0034			
F	0.55	1.67	2.33	0.008	0.0010	0.88	0.0031			
G	0.67	1.72	1.14	0.010	0.0011	0.95	0.0026			
H	0.34	1.88	2.26	0.011	0.0009	1.19	0.0034			
I	0.37	1.23	1.68	0.009	0.0010	2.47	0.0027			
J	0.33	0.04	2.71	0.008	0.0011	4.46	0.0031			Ti: 0.05
K	0.41	2.26	2.54	0.011	0.0011	0.49	0.0025			Nb: 0.04
L	0.43	2.41	2.37	0.013	0.0013	0.33	0.0022			V: 0.06
M	0.35	1.37	1.90	0.009	0.0008	0.50	0.0036			Ni: 0.31
N	0.33	1.45	2.29	0.010	0.0011	0.61	0.0027			Cu: 0.35
O	0.36	1.32	2.63	0.013	0.0007	0.59	0.0031			B: 0.0024
P	0.30	1.49	2.91	0.009	0.0009	0.52	0.0020			Ca: 0.002
Q	0.34	1.92	2.58	0.011	0.0010	0.47	0.0022			REM: 0.001
R	0.41	1.37	2.35	0.008	0.0011	0.54	0.0036			Mg: 0.002
S	0.42	1.46	2.07	0.010	0.0010	0.45	0.0029			W: 0.14
T	0.39	1.65	2.23	0.011	0.0012	0.63	0.0030			Zr: 0.12
U	0.37	1.41	2.61	0.009	0.0013	0.56	0.0034			Sb: 0.02
V	0.34	1.55	2.88	0.008	0.0010	0.49	0.0027			Sn: 0.03
W	0.34	1.32	2.42	0.012	0.0009	0.52	0.0023			Y: 0.01
X	0.27	3.68	1.90	0.011	0.0010	0.47	0.0032			Hf: 0.02
Y	0.36	2.23	2.43	0.009	0.0011	0.44	0.0035			Co: 0.31
XA	0.23	1.45	2.58	0.011	0.0009	0.51	0.0028			
XB	0.78	1.59	2.27	0.008	0.0010	0.39	0.0024			
XC	0.33	0.02	2.42	0.009	0.0013	0.03	0.0025			
XD	0.35	4.17	2.83	0.011	0.0009	0.04	0.0037			
XE	0.41	0.03	2.35	0.009	0.0010	5.27	0.0026			
XF	0.42	1.48	0.76	0.010	0.0008	0.45	0.0031			
XG	0.37	1.59	5.25	0.0011	0.0010	0.50	0.0026			
XH	0.36	2.35	2.47	0.009	0.0009	0.46	0.0033	3.27		
XI	0.34	2.26	2.26	0.010	0.0008	0.53	0.0029		3.28	

[Table 2]

Specimen No.	Steel type	Coiling temperature of hot-rolled steel sheet (°C)	Annealing temperature of hot-rolled steel sheet (°C)	Annealing time of hot-rolled steel sheet (s)	Primary average heating rate (°C/s)	Primary maintaining temperature section (°C)	Primary maintaining time (s)
1	A	550	750	1300	10	Two-phase region	120
2	A	500	900	1100	Poor pickling		
3	A	550	600	1200	Occurrence of fracture during cold rolling		
4	A	450	700	1800	Poor pickling		
5	A	550	750	500	Occurrence of fracture during cold rolling		
6	A	500	700	1100	10	Single-phase region	120
7	A	500	650	1300	10	Two-phase region	120
8	B	550	800	1000	10	Two-phase region	120
9	B	500	700	1200	10	Two-phase region	120
10	B	450	750	1300	10	Two-phase region	120
11	C	500	650	1000	10	Two-phase region	120
12	C	550	700	1200	10	Two-phase region	120
13	C	500	700	1100	10	Two-phase region	120
14	C	500	750	1300	10	Two-phase region	120
15	C	450	800	600	10	Two-phase region	120
16	C	500	750	1200	10	Two-phase region	120
17	C	550	700	1700	10	Two-phase region	120
18	D	500	750	1200	10	Two-phase region	120
19	E	500	700	900	10	Two-phase region	120
20	F	500	850	1200	10	Two-phase region	120
21	G	350	700	1500	10	Two-phase region	120
22	H	450	750	1300	10	Two-phase region	120

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(continued)

Specimen No.	Steel type	Coiling temperature of hot-rolled steel sheet (°C)	Annealing temperature of hot-rolled steel sheet (°C)	Annealing time of hot-rolled steel sheet (s)	Primary average heating rate (°C/s)	Primary maintaining temperature section (°C)	Primary maintaining time (s)
23	I	500	700	1400	10	Two-phase region	120
24	J	450	700	1000	10	Two-phase region	120
25	K	550	750	900	10	Two-phase region	120

[Table 3]

Specimen No.	Steel type	Coiling temperature of hot-rolled steel sheet (°C)	Annealing temperature of hot-rolled steel sheet (°C)	Annealing time of hot-rolled steel sheet (s)	Primary average heating rate (°C/s)	Primary maintaining temperature section (°C)	Primary maintaining time (s)
26	L	550	700	1300	10	Two-phase region	120
27	M	500	750	1100	10	Two-phase region	120
28	N	550	750	1200	10	Two-phase region	120
29	O	450	700	900	10	Two-phase region	120
30	P	550	700	1500	10	Two-phase region	120
31	Q	500	700	1300	10	Two-phase region	120
32	R	450	800	900	10	Two-phase region	120
33	S	500	750	1000	10	Two-phase region	120
34	T	550	700	1500	10	Two-phase region	120
35	U	500	800	1300	10	Two-phase region	120
36	V	500	700	1400	10	Two-phase region	120
37	W	550	750	1200	10	Two-phase region	120
38	X	450	700	1300	10	Two-phase region	120
39	Y	550	750	1400	10	Two-phase region	120

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(continued)

Specimen No.	Steel type	Coiling temperature of hot-rolled steel sheet (°C)	Annealing temperature of hot-rolled steel sheet (°C)	Annealing time of hot-rolled steel sheet (s)	Primary average heating rate (°C/s)	Primary maintaining temperature section (°C)	Primary maintaining time (s)
40	XA	500	800	1200	10	Two-phase region	120
41	XB	550	750	1300	10	Two-phase region	120
42	XC	550	700	1200	10	Two-phase region	120
43	XD	500	750	900	10	Two-phase region	120
44	XE	500	750	1200	10	Two-phase region	120
45	XF	500	700	1000	10	Two-phase region	120
46	XG	450	700	1300	10	Two-phase region	120
47	XH	500	800	1600	10	Two-phase region	120
48	XI	550	700	1300	10	Two-phase region	120

[Table 4]

Specimen No.	Steel type	Primary average cooling rate (°C/s)	Primary cooling stop temperature (°C)	Secondary average heating rate (°C/s)	Secondary maintaining temperature (°C)	Secondary maintaining time (s)	Secondary average cooling rate (°C/s)
1	A	20	220	15	400	300	10
2	A	Poor pickling					
3	A	Occurrence of fracture during cold rolling					
4	A	Poor pickling					
5	A	Occurrence of fracture during cold rolling					
6	A	20	220	15	400	300	10
7	A	0.5	200	15	400	300	10
8	B	20	200	15	400	300	10
9	B	20	180	15	400	300	10
10	B	20	240	15	350	600	10
11	C	20	140	15	400	300	10
12	C	20	80	15	450	300	10
13	C	20	330	15	400	300	10
14	C	20	200	15	270	300	10

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(continued)

Specimen No.	Steel type	Primary average cooling rate (°C/s)	Primary cooling stop temperature (°C)	Secondary average heating rate (°C/s)	Secondary maintaining temperature (°C)	Secondary maintaining time (s)	Secondary average cooling rate (°C/s)
15	C	20	180	15	530	300	10
16	C	20	240	15	400	40	10
17	C	20	200	15	350	172,000	10
18	D	20	220	15	400	300	10
19	E	20	180	15	400	300	10
20	F	20	220	15	450	300	10
21	G	20	280	15	300	300	10
22	H	20	200	15	450	300	10
23	I	20	240	15	500	500	10
24	J	20	120	15	350	300	10
25	K	20	240	15	400	300	10

[Table 5]

Specimen No.	Steel type	Primary average cooling rate (°C/s)	Primary cooling stop temperature (°C)	Secondary average heating rate (°C/s)	Secondary maintaining temperature (°C)	Secondary maintaining time (s)	Secondary average cooling rate (°C/s)
26	L	20	240	15	400	300	10
27	M	20	200	15	400	300	10
28	N	20	200	15	450	300	10
29	O	20	240	15	350	600	10
30	P	20	180	15	400	300	10
31	Q	20	220	15	400	300	10
32	R	20	200	15	400	300	10
33	S	20	200	15	450	300	10
34	T	20	240	15	450	300	10
35	U	20	220	15	400	300	10
36	V	20	200	15	450	300	10
37	W	20	200	15	400	300	10
38	X	20	240	15	450	600	10
39	Y	20	220	15	400	300	10
40	XA	20	180	15	300	300	10
41	XB	20	200	15	400	300	10
42	XC	20	180	15	400	300	10
43	XD	20	200	15	450	600	10
44	XE	20	220	15	400	300	10

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(continued)

Specimen No.	Steel type	Primary average cooling rate (°C/s)	Primary cooling stop temperature (°C)	Secondary average heating rate (°C/s)	Secondary maintaining temperature (°C)	Secondary maintaining time (s)	Secondary average cooling rate (°C/s)
45	XF	20	200	15	450	300	10
46	XG	20	200	15	400	300	10
47	XH	20	220	15	400	300	10
48	XI	20	200	15	400	300	10

[Table 6]

Specimen No.	Steel type	Ferrite (vol.%)	Bainite (vol.%)	Tempered martensite (vol.%)	Fresh martensite (vol.%)	Retained austenite (vol.%)	Perlite (vol.%)
1	A	11	15	55	0	19	0
2	A	Poor pickling					
3	A	Occurrence of fracture during cold rolling					
4	A	Poor pickling					
5	A	Occurrence of fracture during cold rolling					
6	A	1	21	61	0	17	0
7	A	22	13	48	0	5	12
8	B	5	18	61	0	16	0
9	B	12	15	54	1	18	0
10	B	8	16	56	0	20	0
11	C	10	22	47	0	21	0
12	C	11	6	78	0	5	0
13	C	9	64	7	1	19	0
14	C	10	13	72	0	5	0
15	C	13	18	62	1	6	0
16	C	9	13	74	0	4	0
17	C	11	19	63	2	5	0
18	D	10	17	54	0	19	0
19	E	13	14	58	0	15	0
20	F	7	22	52	1	18	0
21	G	6	43	36	0	15	0
22	H	18	16	49	0	17	0
23	I	11	18	57	0	14	0
24	J	9	16	56	1	18	0
25	K	10	19	55	0	16	0

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[Table 7]

Specimen No.	Steel type	Ferrite (vol.%)	Bainite (vol.%)	Tempered martensite (vol.%)	Fresh martensite (vol.%)	Retained austenite (vol.%)	Perlite (vol.%)
26	L	8	16	57	1	18	0
27	M	10	15	58	1	16	0
28	N	5	17	63	0	15	0
29	O	7	22	57	0	14	0
30	P	9	21	50	1	19	0
31	Q	11	17	57	0	15	0
32	R	7	14	45	0	34	0
33	S	10	18	50	1	21	0
34	T	9	19	53	0	19	0
35	U	12	15	56	0	17	0
36	V	8	17	59	0	16	0
37	W	11	20	55	0	14	0
38	X	8	21	54	0	17	0
39	Y	9	18	58	0	15	0
40	XA	10	16	61	0	13	0
41	XB	7	14	18	19	42	0
42	XC	8	21	66	2	3	0
43	XD	6	11	43	22	18	0
44	XE	5	16	44	20	15	0
45	XF	7	18	62	0	5	8
46	XG	6	14	50	14	16	0
47	XH	8	17	43	19	13	0
48	XI	7	13	49	16	15	0

[Table 8]

Specimen No.	Steel type	$[H]_F / [H]_{TM+B+\gamma}$	B_{T-E} (MPa%)	B_{T-H} (MPa ^{2%} ^{1/2})	B_R [R/t]
1	A	0.67	30,636	10,606,040	1.95
2	A	Poor pickling			
3	A	Occurrence of fracture during cold rolling			
4	A	Poor pickling			
5	A	Occurrence of fracture during cold rolling			
6	A	0.21	28,891	6,393,193	2.42
7	A	0.94	18,879	8,475,048	2.54
8	B	0.72	30,518	10,517,710	1.67
9	B	0.77	28,718	11,559,177	1.33
10	B	0.60	31,085	9,926,629	1.58

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(continued)

Specimen No.	Steel type	$[H]_F / [H]_{TM+B+\gamma}$	B_{T-E} (MPa%)	B_{T-H} (MPa ^{2%} ^{1/2})	B_R [R/t]
11	C	0.66	30,639	12,510,645	1.87
12	C	0.95	17,278	8,765,217	5.60
13	C	0.76	20,377	5,355,715	2.62
14	C	0.92	15,481	7,284,461	5.36
15	C	0.95	18,648	9,179,726	2.04
16	C	0.93	16,193	11,431,072	5.93
17	C	0.96	19,615	8,696,192	2.24
18	D	0.79	28,996	10,252,535	1.29
19	E	0.69	30,498	8,626,877	0.67
20	F	0.71	31,811	9,272,433	2.89
21	G	0.58	29,064	11,749,404	1.28
22	H	0.81	32,308	8,878,145	1.92
23	I	0.43	29,967	10,849,320	2.34
24	J	0.87	30,882	9,012,837	2.07
25	K	0.78	28,332	8,648,407	1.88

[Table 9]

Specimen No.	Steel type	$[H]_F / [H]_{TM+B+\gamma}$	B_{T-E} (MPa%)	B_{T-H} (MPa ^{2%} ^{1/2})	B_R [R/t]
26	L	0.75	30,157	8,058,963	1.70
27	M	0.80	28,294	11,607,916	1.85
28	N	0.62	31,608	9,754,084	1.48
29	O	0.67	33,459	10,481,642	2.16
30	P	0.72	30,238	12,929,614	2.33
31	Q	0.70	32,795	9,497,041	2.59
32	R	0.77	29,766	8,331,161	2.48
33	S	0.69	31,547	10,596,615	1.77
34	T	0.71	30,303	11,373,489	2.15
35	U	0.64	32,056	10,905,303	2.19
36	V	0.45	31,529	9,571,137	1.99
37	W	0.59	29,818	9,439,702	1.52
38	X	0.67	27,874	11,311,612	1.71
39	Y	0.54	30,397	8,337,602	1.24
40	XA	0.51	19,508	6,015,204	2.08
41	XB	0.83	21,075	5,643,588	6.56
42	XC	0.97	14,997	9,026,852	4.94
43	XD	0.78	26,963	7,986,381	4.57
44	XE	0.66	27,856	10,871,201	6.49

(continued)

Specimen No.	Steel type	$[H]_F / [H]_{TM+B+\gamma}$	B_{T-E} (MPa%)	B_{T-H} (MPa ^{2%} ^{1/2})	B_R [R/t]
45	XF	0.94	15,420	8,540,137	2.21
46	XG	0.80	24,351	9,258,527	4.86
47	XH	0.74	26,759	10,845,197	6.55
48	XI	0.71	27,882	11,049,872	4.73

[0091] As shown in Tables 1 to 9 above, it could be seen that the specimens satisfying the conditions presented in the present invention simultaneously provide superior strength and workability since the value of $[H]_F/[H]_{TM+B+\gamma}$ satisfies the range of 0.4 to 0.9, the balance (TSXE1) of tensile strength and elongation is 22,000 MPa% or more, the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio is 7×10^6 (MPa^{2%}^{1/2}) or more, and the bendability (R/t) satisfies the range of 0.5 to 3.0.

[0092] It could be seen that, in specimens 2 to 5, since the alloy composition range of the present invention overlaps, but the hot rolling annealing temperature and time are outside the range of the present invention, the pickling failure occurs or the fracture occurs during the cold rolling.

[0093] In specimen 6, the amount of ferrite formed was insufficient because the primary heating or maintaining temperature in the annealing heat treatment process after the cold rolling exceeded the range limited by the present invention. As a result, it could be seen that, in specimen 6, $[H]_F/[H]_{TM+B+\gamma}$ is less than 0.4, and the balance of tensile strength and hole expansion ratio ($TS^2 \times HER^{1/2}$) is less than 7×10^6 (MPa^{2%}^{1/2}).

[0094] In specimen 7, the primary cooling rate in the annealing heat treatment after the cold rolling did not reach the range limited by the present invention, so the ferrite was excessively formed and the retained austenite was formed less. As a result, it could be seen that, in specimen 7, $[H]_F/[H]_{TM+B+\gamma}$ exceeds 0.9, and the balance (TSXE1) of tensile strength and elongation is less than 22,000 MPa%.

[0095] In specimen 12, the primary cooling stop temperature was low, so that tempered martensite was excessively formed and less retained austenite was formed. As a result, it could be seen that, in specimen 12, $[H]_F/[H]_{TM+B+\gamma}$ exceeds 0.9, the balance (TSXE1) of tensile strength and elongation is less than 22,000 MPa%, and the bendability (R/t) exceeds 3.0.

[0096] In Specimen 13, the primary cooling stop temperature was high, so that bainite was excessively formed and less tempered martensite was formed. As a result, it could be seen that, in specimen 13, the balance (TSXE1) of tensile strength and elongation is less than 22,000 MPa% and the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion ratio is less than 7×10^6 (MPa^{2%}^{1/2}).

[0097] In specimen 14, the secondary heating or maintaining temperature was low, so that tempered martensite was excessively formed and less retained austenite was formed. As a result, it could be seen that, in specimen 14, $[H]_F/[H]_{TM+B+\gamma}$ exceeds 0.9, the balance (TSXE1) of tensile strength and elongation is less than 22,000 MPa%, and the bendability (R/t) exceeds 3.0.

[0098] It could be seen that, in specimen 15, the secondary heating or maintaining temperature is high, so the amount of retained austenite formed is insufficient, $[H]_F/[H]_{TM+B+\gamma}$ exceeds 0.9, and the balance (TSXE1) of tensile strength and elongation is less than 22,000 MPa%.

[0099] In specimen 16, the secondary maintaining time was insufficient, so that tempered martensite was excessively formed and less retained austenite was formed. As a result, it could be seen that, in specimen 16, $[H]_F/[H]_{TM+B+\gamma}$ exceeds 0.9, the balance (TSXE1) of tensile strength and elongation is less than 22,000 MPa%, and the bendability (R/t) exceeds 3.0.

[0100] It could be seen that, in specimen 17, the secondary maintaining time was excessive, so the amount of retained austenite formed was insufficient, $[H]_F/[H]_{TM+B+\gamma}$ exceeded 0.9, and the balance (TSXE1) of tensile strength and elongation was less than 22,000 MPa%.

[0101] Specimens 40 to 48 may satisfy the manufacturing conditions presented in the present invention, but may be outside the alloy composition range. In these cases, it could be seen that the condition of the $[H]_F/[H]_{TM+B+\gamma}$, the condition of the balance (TSXE1) of tensile strength and elongation, the condition of the balance ($TS^2 \times HER^{1/2}$) of tensile strength and hole expansion, and the condition of bendability (R/t) of the present invention are not all satisfied. Meanwhile, it could be seen that, in specimen 42, when the total content of aluminum (Al) and silicon (Si) is less than 1.0%, the conditions of $[H]_F/[H]_{TM+B+\gamma}$, the balance (TSXE1) of tensile strength and elongation, and the bendability (R/t) are not satisfied.

[0102] While the present invention has been described in detail through exemplary embodiment, other types of exemplary embodiments are also possible. Therefore, the technical spirit and scope of the claims set forth below are not limited to exemplary embodiments.

Claims

1. A high-strength steel sheet having superior workability, comprising:

by wt%, C: 0.25 to 0.75%, Si: 4.0% or less, Mn: 0.9 to 5.0%, Al: 5.0% or less, P: 0.15% or less, S: 0.03% or less, N: 0.03% or less, a balance of Fe, and unavoidable impurities; and as microstructures, ferrite which is a soft structure, and tempered martensite, bainite, and retained austenite which are hard structures, wherein the high-strength steel sheet satisfies the following [Relational Expression 1]

[Relational Expression 1]

$$0.4 \leq [H]_F / [H]_{TM+B+\gamma} \leq 0.9$$

in the above Relational Expression 1, $[H]_F$ and $[H]_{TM+B+\gamma}$ are nanohardness values measured using a nanoindenter, $[H]_F$ is an average nanohardness value Hv of the ferrite which is the soft structure, $[H]_{TM+B+\gamma}$ is the average nanohardness value Hv of the tempered martensite, the bainite, and the retained austenite which are the hard structures.

2. The high-strength steel sheet of claim 1, including one or more of the following (1) to (9):

- (1) one or more of Ti: 0 to 0.5%, Nb: 0 to 0.5%, and V: 0 to 0.5%;
- (2) one or more of Cr: 0 to 3.0% and Mo: 0 to 3.0%;
- (3) one or more of Cu: 0 to 4.5% and Ni: 0 to 4.5%;
- (4) B: 0 to 0.005%;
- (5) one or more of Ca: 0 to 0.05%, REM: 0 to 0.05% excluding Y, and Mg: 0 to 0.05%;
- (6) one or more of W: 0 to 0.5% and Zr: 0 to 0.5%;
- (7) one or more of Sb: 0 to 0.5% and Sn: 0 to 0.5%;
- (8) one or more of Y: 0 to 0.2% and Hf: 0 to 0.2%; and
- (9) Co: 0 to 1.5%.

3. The high-strength steel sheet of claim 1, wherein a total content (Si+Al) of Si and Al is 1.0 to 6.0 wt%.

4. The high-strength steel sheet of claim 1, wherein the microstructure of the steel sheet includes, by volume fraction, 30 to 70% of tempered martensite, 10 to 45% of bainite, 10 to 40% of retained austenite, 3 to 20% of ferrite, and unavoidable microstructure components.

5. The high-strength steel sheet of claim 1, wherein a balance B_{T-E} of tensile strength and elongation expressed by the following [Relational Expression 2] is 22,000 (MPa%) or more, a balance B_{T-H} of tensile strength and a hole expansion ratio expressed by the following [Relational Expression 3] is 7×10^6 (MPa^{2%}^{1/2}) or more, and bendability B_R expressed by the following [Relational Expression 4] is 0.5 to 3.0,

[Relational Expression 2]

$$B_{T-E} = [\text{Tensile Strength (TS, MPa)}] * [\text{Elongation (El, \%)}]$$

[Relational Expression 3]

$$B_{T-H} = [\text{Tensile Strength (TS, MPa)}]^2 * [\text{Hole Expansion Ratio (HER, \%)}]^{1/2}$$

[Relational Expression 4]

$$B_R = R/t$$

where R is a minimum bending radius (mm) at which cracks do not occur after a 90° bending test, and t is a thickness (mm) of the steel sheet.

6. A manufacturing method for high-strength steel sheet having superior workability, comprising:

heating and hot rolling a steel slab including, by wt%, C: 0.25 to 0.75%, Si: 4.0% or less, Mn: 0.9 to 5.0%, Al: 5.0% or less, P: 0.15% or less, S: 0.03% or less, N: 0.03% or less, a balance of Fe, and unavoidable impurities; coiling the hot-rolled steel sheet; coiling up the hot-rolled steel sheet into a coil; performing a hot-rolled annealing heat treatment on the coiled steel sheet in a temperature within a range of 650 to 850°C for 600 to 1700 seconds; cold rolling the hot-rolled annealing heat-treated steel sheet; heating (primary heating) the cold-rolled steel sheet to a temperature within a range of Ac1 or higher and less than Ac3, and maintaining (primary maintaining) the cold-rolled steel sheet for 50 seconds or more; cooling (primary cooling) the primarily heated steel sheet to a temperature within a range of 100 to 300°C at an average cooling rate of 1°C/s or more; heating (secondary heating) the primarily cooled steel sheet to a temperature within a range of 300 to 500°C at an average temperature increase rate of 5°C/s or more, and maintaining (secondarily maintaining) the primarily cooled steel sheet for 50 seconds or more; and cooling (secondarily cooling) the secondarily heated steel sheet to room temperature.

7. The manufacturing method of claim 6, wherein the steel slab further includes one or more of the following (1) to (9) :

- (1) one or more of Ti: 0 to 0.5%, Nb: 0 to 0.5%, and V: 0 to 0.5%;
- (2) one or more of Cr: 0 to 3.0% and Mo: 0 to 3.0%;
- (3) one or more of Cu: 0 to 4.5% and Ni: 0 to 4.5%;
- (4) B: 0 to 0.005%;
- (5) one or more of Ca: 0 to 0.05%, REM: 0 to 0.05% excluding Y, and Mg: 0 to 0.05%;
- (6) one or more of W: 0 to 0.5% and Zr: 0 to 0.5%;
- (7) one or more of Sb: 0 to 0.5% and Sn: 0 to 0.5%;
- (8) one or more of Y: 0 to 0.2% and Hf: 0 to 0.2%; and
- (9) Co: 0 to 1.5%.

8. The manufacturing method of claim 6, wherein a total content (Si+Al) of Si and Al included in the steel slab is 1.0 to 6.0 wt%.

9. The manufacturing method of claim 6, wherein the steel slab is heated to a temperature within a range of 1000 to 1350°C, and is subjected to finish hot rolling in a temperature within a range of 800 to 1000°C.

10. The manufacturing method of claim 6, wherein the hot-rolled steel sheet is coiled at a temperature within a range of 300 to 600°C.

11. The manufacturing method of claim 6, wherein a reduction ratio of the cold rolling is 30 to 90%.

12. The manufacturing method of claim 6, wherein a cooling rate of the secondary cooling is 1°C/s or more.

INTERNATIONAL SEARCH REPORT

International application No.

PCT/KR2020/016838

A. CLASSIFICATION OF SUBJECT MATTER C22C 38/04(2006.01)i; C22C 38/02(2006.01)i; C22C 38/06(2006.01)i; C22C 38/00(2006.01)i; C21D 8/02(2006.01)i; C21D 9/68(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 38/04(2006.01); C21D 8/02(2006.01); C21D 9/46(2006.01); C21D 9/48(2006.01); C22C 38/00(2006.01); C22C 38/38(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: 가공성(workability), 고강도(high strength), 템퍼드 마르텐사이트(tempered martensite), 변태유기소성강(TRIP steel), 소둔(annealing)																		
C. DOCUMENTS CONSIDERED TO BE RELEVANT <table border="1"> <thead> <tr> <th>Category*</th> <th>Citation of document, with indication, where appropriate, of the relevant passages</th> <th>Relevant to claim No.</th> </tr> </thead> <tbody> <tr> <td>X</td> <td>JP 2013-019047 A (KOBEL STEEL LTD.) 31 January 2013 (2013-01-31) See paragraphs [0053]-[0067], [0076]-[0077] and [0092], claims 1, 3 and 9 and table 4.</td> <td>1-12</td> </tr> <tr> <td>X</td> <td>JP 2012-237042 A (JFE STEEL CORP.) 06 December 2012 (2012-12-06) See paragraphs [0029]-[0038] and claims 1-2.</td> <td>1-12</td> </tr> <tr> <td>A</td> <td>KR 10-2019-0007055 A (NIPPON STEEL & SUMITOMO METAL CORPORATION) 21 January 2019 (2019-01-21) See paragraphs [0129]-[0142] and claim 1.</td> <td>1-12</td> </tr> <tr> <td>A</td> <td>WO 2018-221307 A1 (KABUSHIKI KAISHA KOBE SEIKO SHO (KOBEL STEEL, LTD.)) 06 December 2018 (2018-12-06) See paragraphs [0056]-[0073].</td> <td>1-12</td> </tr> <tr> <td>A</td> <td>KR 10-2018-0096637 A (ARCELORMITTAL) 29 August 2018 (2018-08-29) See claims 1 and 6.</td> <td>1-12</td> </tr> </tbody> </table>	Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.	X	JP 2013-019047 A (KOBEL STEEL LTD.) 31 January 2013 (2013-01-31) See paragraphs [0053]-[0067], [0076]-[0077] and [0092], claims 1, 3 and 9 and table 4.	1-12	X	JP 2012-237042 A (JFE STEEL CORP.) 06 December 2012 (2012-12-06) See paragraphs [0029]-[0038] and claims 1-2.	1-12	A	KR 10-2019-0007055 A (NIPPON STEEL & SUMITOMO METAL CORPORATION) 21 January 2019 (2019-01-21) See paragraphs [0129]-[0142] and claim 1.	1-12	A	WO 2018-221307 A1 (KABUSHIKI KAISHA KOBE SEIKO SHO (KOBEL STEEL, LTD.)) 06 December 2018 (2018-12-06) See paragraphs [0056]-[0073].	1-12	A	KR 10-2018-0096637 A (ARCELORMITTAL) 29 August 2018 (2018-08-29) See claims 1 and 6.	1-12
Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.																
X	JP 2013-019047 A (KOBEL STEEL LTD.) 31 January 2013 (2013-01-31) See paragraphs [0053]-[0067], [0076]-[0077] and [0092], claims 1, 3 and 9 and table 4.	1-12																
X	JP 2012-237042 A (JFE STEEL CORP.) 06 December 2012 (2012-12-06) See paragraphs [0029]-[0038] and claims 1-2.	1-12																
A	KR 10-2019-0007055 A (NIPPON STEEL & SUMITOMO METAL CORPORATION) 21 January 2019 (2019-01-21) See paragraphs [0129]-[0142] and claim 1.	1-12																
A	WO 2018-221307 A1 (KABUSHIKI KAISHA KOBE SEIKO SHO (KOBEL STEEL, LTD.)) 06 December 2018 (2018-12-06) See paragraphs [0056]-[0073].	1-12																
A	KR 10-2018-0096637 A (ARCELORMITTAL) 29 August 2018 (2018-08-29) See claims 1 and 6.	1-12																
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Date of the actual completion of the international search 23 February 2021	Date of mailing of the international search report 24 February 2021																	
Name and mailing address of the ISA/KR Korean Intellectual Property Office Government Complex-Daejeon Building 4, 189 Cheongsaro, Seo-gu, Daejeon 35208 Facsimile No. +82-42-481-8578	Authorized officer Telephone No.																	

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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.

PCT/KR2020/016838

Patent document cited in search report	Publication date (day/month/year)	Patent family member(s)	Publication date (day/month/year)
JP 2013-019047 A	31 January 2013	CN 102828106 A	19 December 2012
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