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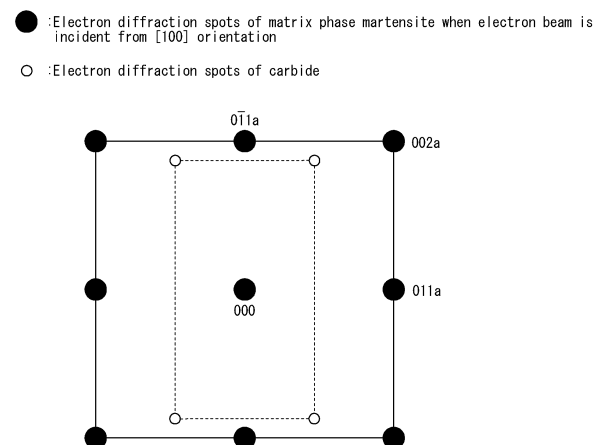
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(54) **HIGH-STRENGTH STEEL SHEET, MEMBER, AND MANUFACTURING METHODS THEREFOR**

(57) Provided is a high strength steel sheet having high strength and excellent component strength, ductility, stretch flangeability, bendability, and warm workability. The high strength steel sheet has a chemical composition containing, in mass%, C: 0.030 % to 0.500 %, Si: 0.01 % to 2.50 %, Mn: 0.10 % to 5.00 %, P: 0.100 % or less, S: 0.0200 % or less, Al: 1.000 % or less, N: 0.0100 % or less, and O: 0.0100 % or less, with the balance being Fe and inevitable impurity. At a 1/4 sheet thickness position, the steel microstructure is such that area fractions are 60 % or more martensite, 40 % or less ferrite, and 20 % or less retained austenite, and the lattice constant a_M of martensite satisfies the following Expression: $1.00005 \leq a_M/a_R \leq 1.00500$, where a_R is the lattice constant of the martensite at room temperature after heat treatment of the high strength steel sheet at 500 °C for 30 min.

FIG. 1



Description

TECHNICAL FIELD

[0001] The present disclosure relates to a high strength steel sheet, a member, and methods of producing same.

BACKGROUND

[0002] Higher strength steel sheets for automobiles are being developed to both reduce CO₂ emissions by reducing vehicle weight and improve crashworthiness by reducing automotive body weight. New laws and regulations concerning automobiles are also being introduced one after another.

[0003] For the purpose of increasing automotive body strength, high strength cold-rolled steel sheets having a tensile strength (TS) of 980 MPa or more are increasingly being applied to major structural components that form the framework of automobile cabins. Patent Literature (PTL) 1 describes a high strength cold-rolled steel sheet having a tensile strength of 1310 MPa or more. Further, in order to increase the ratio of high strength steel sheet use in automobiles, in addition to the application of hot stamping, attention is being paid to warm working, which involves forming at lower heating temperature. PTL 2 describes a high strength steel sheet for warm working that has excellent workability even at a low heating temperature from 50 °C to 200°C.

CITATION LIST

Patent Literature

[0004]

PTL 1: WO 2019/181950 A1

PTL 2: WO 2017/131053 A1

SUMMARY

(Technical Problem)

[0005] Conventionally, there has been no high strength steel sheet that has excellent ductility, stretch flangeability, and bendability at room temperature, and excellent warm workability. For example, although the high strength steel sheet of PTL 1 has excellent bendability and ductility at room temperature, no consideration is given to warm workability. Further, although the high strength steel sheet of PTL 2 has excellent warm workability (total elongation), the breakdown of uniform elongation and local elongation at warm temperatures is unclear. Further, stretch flangeability and bendability at room temperature were not considered in PTL 2.

[0006] In view of the circumstances described above, it would be helpful to provide a high strength steel sheet having a tensile strength of 980 MPa or more and excellent component strength, ductility, stretch flangeability, bendability, and warm workability, a member, and methods of producing same.

(Solution to Problem)

[0007] The inventors conducted extensive studies to address the technical problems described above, and as a result discovered that the following configuration can be adopted, and thus arrived at the present disclosure. Primary features of the present disclosure are as follows.

[1] A high strength steel sheet comprising a chemical composition containing (consisting of), in mass%,

C: 0.030 % or more and 0.500 % or less,

Si: 0.01 % or more and 2.50 % or less,

Mn: 0.10 % or more and 5.00 % or less,

P: 0.100 % or less,

S: 0.0200 % or less,

Al: 1.000 % or less,

N: 0.0100 % or less, and

O: 0.0100 % or less,

with the balance being Fe and inevitable impurity, wherein
 at a 1/4 sheet thickness position, the steel microstructure is such that an area fraction of martensite is 60 % or more,
 an area fraction of ferrite is 40 % or less,
 an area fraction of retained austenite is 20 % or less, and
 a lattice constant a_M of the martensite, in nm, satisfies the following Expression 1,

$$\text{Expression 1: } 1.00005 \leq a_M/a_R \leq 1.00500$$

where, a_R is the lattice constant of the martensite, in nm, at room temperature after heat treatment of the high strength steel sheet at 500 °C for 30 min.

[2] The high strength steel sheet according to [1], wherein the number of martensitic blocks containing carbide having a major axis length of 200 nm or more is 50 % or less of the number of martensitic blocks containing carbide.

[3] The high strength steel sheet according to [1] or [2], the chemical composition further containing, in mass%, at least one element selected from the group consisting of:

Ti: 0.200 % or less,
 Nb: 0.200 % or less,
 V: 0.200 % or less,
 Ta: 0.10 % or less,
 W: 0.10 % or less,
 B: 0.0100 % or less,
 Cr: 1.00 % or less,
 Mo: 1.00 % or less,
 Ni: 1.00 % or less,
 Co: 0.010 % or less,
 Cu: 1.00 % or less,
 Sn: 0.200 % or less,
 Sb: 0.200 % or less,
 Ca: 0.0100 % or less,
 Mg: 0.0100 % or less,
 REM: 0.0100 % or less,
 Zr: 0.100 % or less,
 Te: 0.100 % or less,
 Hf: 0.10 % or less, and
 Bi: 0.200 % or less.

[4] The high strength steel sheet according to any one of [1] to [3], further comprising a coated or plated layer on at least one side.

[5] A member formed using the high strength steel sheet according to any one of [1] to [4].

[6] The member according to [5], wherein the member is for an automobile frame structural component or reinforcing component.

[7] A method of producing a high strength steel sheet, the method comprising:

rough rolling a steel slab having the chemical composition according to [1] or [3] at an average strain rate of 1×10^{-4} /s or more and 1×10^{-1} /s or less with a total rolling reduction of 50 % or more to obtain a rough rolled sheet,
 then finish rolling the rough rolled sheet to obtain a hot-rolled sheet;
 cold rolling the hot-rolled sheet to obtain a cold-rolled sheet;
 annealing the cold-rolled sheet by heating to a heating temperature of 800 °C or more, and cooling under a condition that a residence time t_1 in a temperature range T_1 from M_s temperature to 700 °C is 1000 s or less; and
 partially tempering the cold-rolled sheet by cooling from the M_s temperature to 50 °C or less under a condition that a partial tempering parameter S , in μm^2 , satisfies the following Expression 2,

$$\text{Expression 2:}$$

$$0.01 \leq S \leq 30$$

Expression 3:

$$S = \sum_{t=1}^{t_E} f(t) D(T) \times 10^{12}$$

Expression 4:

$$f(t) = 1 - \exp(-1.1 \times 10^{-2}(Ms - T_{min}))$$

Expression 5:

$$D(T) = 0.394 \times 10^{-6} \exp\left(-\frac{80220}{8.314 \times (T+273)}\right)$$

where, in Expressions 2 to 5,

a time when the temperature of the cold-rolled sheet first reaches the Ms temperature after the annealing is defined as $t = 0$, in s, and $t = t_E$, in s, is the time when the partial tempering is completed and the temperature of the cold-rolled sheet is 50 °C,

T , in °C, is an average temperature of the cold-rolled sheet at time $t-1$ to t , in s,

T_{min} , in °C, is the lowest temperature of T at time 0 to t , in s, and

Ms , in °C, indicates the Ms temperature of the high strength steel sheet.

[8] The method of producing a high strength steel sheet according to [7], where, in the partial tempering, cooling is stopped at a cooling stop temperature that is room temperature or more and the Ms temperature or less, followed by reheating to a reheating temperature and then cooling to 50 °C or less.

[9] The method of producing a high strength steel sheet according to [7] or [8], where, in the annealing, coating or plating treatment is further applied to at least one side of the cold-rolled sheet.

[10] A method of producing a member, the method comprising at least one of a forming process or a joining process applied to the high strength steel sheet according to any one of [1] to [4] to obtain the member.

(Advantageous Effect)

[0008] According to the present disclosure, a high strength steel sheet having a tensile strength of 980 MPa or more and excellent component strength, ductility, stretch flangeability, bendability, and warm workability, a member, and methods of producing same can be provided.

BRIEF DESCRIPTION OF THE DRAWINGS

[0009] In the accompanying drawings:

FIG. 1 is a diagram illustrating an example of an electron diffraction pattern of martensite with carbide;

FIG. 2 is a schematic diagram of thermal hysteresis of a partial tempering process of continuously cooling from the Ms temperature to 50 °C, and a schematic diagram of $f(t)$; and

FIG. 3 is a schematic diagram of thermal hysteresis when reheating is carried out in partial tempering, and a schematic diagram of $f(t)$.

DETAILED DESCRIPTION

[0010] The following describes embodiments of the present disclosure. However, the present disclosure is not limited to the following embodiments. The high strength steel sheet according to the present embodiment has the chemical composition and steel microstructure described below. Hereinafter, "high strength steel sheet" is also referred to simply as "steel sheet".

[0011] The high strength steel sheet according to the present embodiment has a tensile strength of 980 MPa or more and excellent component strength, ductility, stretch flangeability, bendability, and warm workability.

[0012] High strength means that the tensile strength (TS) is 980 MPa or more, as determined by the tensile test described below.

[0013] Excellent component strength means that the yield ratio (YR) is 60 % or more, as determined by the tensile test described below.

[0014] Excellent ductility means that the total elongation (T-El) is 6.0 % or more, as determined by the tensile test described below.

[0015] Excellent stretch flangeability means that the hole expansion ratio (λ) is 30 % or more, as determined by the hole expanding test described below.

[0016] Excellent bendability means that the limit bending radius (R/t) is 5.0 or less, as determined by the bend test of a sample having a ground end face as described below.

[0017] Excellent warm workability means that when the uniform elongation (U-El_{RT}) and the local elongation (L-El_{RT}) are determined by the tensile test at room temperature described below, and the uniform elongation (U-El₂₀₀) and the local elongation (L-El₂₀₀) are determined by the tensile test at 200 °C, then U-El₂₀₀/U-El_{RT} is 1.10 or more and L-El₂₀₀/L-El_{RT} is 1.30 or more.

[0018] When warm working, from the viewpoint of strain dispersion capability, the uniform elongation of the steel sheet needs to be higher than at room temperature, and in order to suppress fracture of the worked portion, the local elongation of the steel sheet needs to be higher than at room temperature. In warm working, it is assumed that a press mold and a steel sheet are heated to a defined temperature for working, but end face portions of the steel sheet tend to cool down, and therefore it is assumed that the working is essentially carried out at room temperature. Therefore, in addition to excellent warm workability, the steel sheet is required to have good ductility, stretch flangeability, and bendability at room temperature. Further, high strength steel sheets used for automobile reinforcing components and frame structural components are required to have excellent component strength (high impact absorbed energy in a crash). This can be achieved by increasing the yield strength (YS) of a steel sheet or by increasing the yield ratio (YR = Yield strength YS / Tensile strength TS × 100 [%]) of the steel sheet. Conventionally, there were no high strength steel sheets having good ductility, stretch flangeability, and bendability at room temperature and excellent workability at warm temperatures. In conventionally used steel sheets containing a large amount of retained austenite phase, very hard fresh martensite is formed by working by punching, and low stretch flangeability is expected in terms of void formation, growth, and consolidation. As described below, as a result of extensive studies, the inventors arrived at the idea of appropriately adjusting a partial tempering parameter S. This allows a ratio a_M/a_R between the lattice constant a_M of martensite at room temperature and the lattice constant a_R at room temperature after heat treatment of a high strength steel sheet including martensite at 500 °C for 30 min to be set to an appropriate value. By setting a_M/a_R to an appropriate value, it is possible to provide the high strength steel sheet having good ductility, stretch flangeability, and bendability at room temperature and excellent workability at warm temperatures.

(Chemical composition)

[0019] First, appropriate ranges for a chemical composition of the high strength steel sheet and reasons for such limitations are described. In the following description, "%" representing the content of a component element in the steel sheet means "mass%" unless otherwise specified.

((C: 0.030 % or more and 0.500 % or less))

[0020] C is one of the important basic components of a steel sheet, and particularly in the high strength steel sheet according to the present embodiment, C affects the area fraction of martensite and the lattice constant of martensite.

[0021] When C content is too low, the area fraction of martensite decreases, making it difficult to achieve a TS of 980 MPa or more. Further, the lattice constant a_M of martensite decreases, and warm workability decreases. Further, the area fraction of ferrite increases. The C content of the steel sheet is therefore 0.030 % or more. The C content is preferably 0.050 % or more. The C content is more preferably 0.090 % or more.

[0022] On the other hand, when the C content is too high, the retained austenite increases excessively and the hardness of the martensite formed from the retained austenite during punching increases significantly. As a result, crack propagation during hole expansion is promoted, resulting in a lower hole expansion ratio and decreased stretch flangeability. Further, stress-induced transformation of retained austenite decreases YR, resulting in lower component strength. The C content is therefore 0.500 % or less. The C content is preferably 0.400 % or less. The C content is more preferably 0.350 % or less.

((Si: 0.01 % or more and 2.50 % or less))

[0023] Si increases the strength of the steel sheet by inhibiting the precipitation of cementite in martensite and by solid solution strengthening. To obtain this effect, Si content is 0.01 % or more. Si content is preferably 0.05 % or more. Si content is more preferably 0.10 % or more.

[0024] On the other hand, when the Si content is too high, carbide precipitation during bainite transformation is significantly suppressed, resulting in an excessive increase in retained austenite and a large increase in the hardness of martensite formed from the retained austenite during punching. As a result, crack propagation during hole expansion is

promoted, resulting in a lower hole expansion ratio and decreased stretch flangeability. Further, stress-induced transformation of retained austenite decreases YR, resulting in lower component strength. The Si content is therefore 2.50 % or less. The Si content is preferably 2.00 % or less. The Si content is more preferably 1.50 % or less.

((Mn: 0.10 % or more and 5.00 % or less))

[0025] Mn is one of the important basic components of a steel sheet, and particularly in the high strength steel sheet according to the present embodiment, Mn affects the area fraction of martensite.

[0026] When Mn content is too low, the area fraction of martensite decreases, making it difficult to achieve a TS of 980 MPa or more. Further, the area fraction of martensite decreases, making it difficult to obtain high warm workability. The Mn content is therefore 0.10 % or more. The Mn content is preferably 0.90 % or more. The Mn content is more preferably 1.80 % or more.

[0027] On the other hand, when the Mn content is too high, austenite stabilizes and retained austenite increases excessively. The hardness of martensite formed from retained austenite during punching is then greatly increased. As a result, crack propagation during hole expansion is promoted, resulting in a lower hole expansion ratio and decreased stretch flangeability. Further, stress-induced transformation of retained austenite decreases YR, resulting in lower component strength. The Mn content is therefore 5.00 % or less. The Mn content is preferably 4.20 % or less. The Mn content is more preferably 3.60 % or less.

<<P: 0.100 % or less>>

[0028] P segregates at prior austenite grain boundaries, embrittling grain boundaries, thereby decreasing steel sheet ultimate deformability, resulting in a decrease in λ . Further, bendability also decreases. P content therefore needs to be 0.100 % or less. A lower limit of the P content is not particularly specified. P is a solid-solution-strengthening element able to increase steel sheet strength, and therefore the P content is preferably 0.001 % or more. The P content is preferably 0.070 % or less.

<<S: 0.0200 % or less>>

[0029] S exists as a sulfide and decreases ultimate deformability of the steel sheet, resulting in a decrease in λ . Further, bendability also decreases. S content therefore needs to be 0.0200 % or less. A lower limit of the S content is not particularly specified. In view of production technology constraints, the S content is preferably 0.0001 % or more. The S content is preferably 0.0050 % or less.

((Al: 1.000 % or less))

[0030] Al provides sufficient deoxidation and decreases inclusions in steel.

[0031] However, when Al content is too high, a large amount of ferrite is formed, resulting in a lower hole expansion ratio and decreased stretch flangeability. The Al content is therefore 1.000 % or less. The Al content is preferably 0.500 % or less. The Al content is more preferably 0.100 % or less.

[0032] On the other hand, for stable deoxidation, the Al content is preferably 0.010 % or more. The Al content is more preferably 0.015 % or more. The Al content is even more preferably 0.020 % or more.

((N: 0.0100 % or less))

[0033] N exists as a nitride and decreases ultimate deformability of the steel sheet, resulting in a decrease in λ . Further, bendability also decreases. N content is therefore 0.0100 % or less. A lower limit of the N content is not particularly specified. In view of production technology constraints, the N content is preferably 0.0001 % or more. The N content is preferably 0.0050 % or less.

<<O: 0.0100 % or less>>

[0034] O exists as an oxide and decreases ultimate deformability of the steel sheet, resulting in a decrease in λ . Further, bendability also decreases. O content is therefore 0.0100 % or less. A lower limit of the O content is not particularly specified. In view of production technology constraints, the O content is preferably 0.0001 % or more. The O content is preferably 0.0050 % or less.

[0035] The high strength steel sheet according to the present embodiment has a chemical composition that consists of the above components, with the balance being Fe and inevitable impurity. Examples of inevitable impurity include Zn, Pb,

As, Ge, Sr, and Cs. It is acceptable for these impurities to be present for a total content of 0.100 % or less.

[0036] The chemical composition of the high strength steel sheet according to the present embodiment may further contain, in mass%, at least one element, alone or in combination, selected from the group consisting of: Ti: 0.200 % or less, Nb: 0.200 % or less, V: 0.200 % or less, Ta: 0.10 % or less, W: 0.10 % or less, B: 0.0100 % or less, Cr: 1.00 % or less, Mo: 1.00 % or less, Ni: 1.00 % or less, Co: 0.010 % or less, Cu: 1.00 % or less, Sn: 0.200 % or less, Sb: 0.200 % or less, Ca: 0.0100 % or less, Mg: 0.0100 % or less, REM: 0.0100 % or less, Zr: 0.100 % or less, Te: 0.100 % or less, Hf: 0.10 % or less, and Bi: 0.200 % or less.

[0037] When Ti content, Nb content, and V content are respectively 0.200 % or less, coarse precipitates and inclusions do not form in large amounts, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. Therefore, when Ti is contained in the steel sheet, the Ti content is preferably 0.200 % or less. When Nb is contained in the steel sheet, the Nb content is preferably 0.200 % or less. When V is contained in the steel sheet, the V content is preferably 0.200 % or less. Lower limits of Ti, Nb, and V content are not particularly specified. Ti, Nb, and V content increases strength of the steel sheet by forming fine carbides, nitrides, and carbonitrides during hot rolling or continuous annealing, and therefore the content of each of Ti, Nb, and V is respectively more preferably 0.001 % or more. The Ti content is more preferably 0.100 % or less. The Nb content is more preferably 0.100 % or less. The V content is more preferably 0.100 % or less.

[0038] When Ta content and W content are respectively 0.10 % or less, coarse precipitates and inclusions do not form in large amounts, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. Therefore, Ta content is preferably 0.10 % or less. W content is preferably 0.10 % or less. Lower limits of Ta and W content are not particularly specified. Ta and W content increases strength of the steel sheet by forming fine carbides, nitrides, and carbonitrides during hot rolling or continuous annealing, and therefore the content of each of Ta and W is respectively more preferably 0.01 % or more. The Ta content is more preferably 0.08 % or less. The W content is more preferably 0.08 % or less.

[0039] When B content is 0.0100 % or less, cracks are not caused inside the steel sheet during casting or hot rolling and the ultimate deformability of the steel sheet is not decreased, and therefore λ does not decrease. Further, bendability does not decrease. The B content is therefore preferably 0.0100 % or less. A lower limit of the B content is not particularly specified. B is an element that improves hardenability by segregating to austenite grain boundaries during annealing, and therefore the B content is more preferably 0.0003 % or more. The B content is more preferably 0.0080 % or less.

[0040] When Cr content, Mo content, and Ni content are respectively 1.00 % or less, the amount of coarse precipitates and inclusions do not increase, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. Therefore, Cr content is preferably 1.00 % or less. Mo content is preferably 1.00 % or less. Ni content is preferably 1.00 % or less. Lower limits of Cr, Mo, and Ni content are not particularly specified. Cr, Mo, and Ni are elements that improve hardenability, and therefore the content of each of Cr, Mo, and Ni is respectively more preferably 0.01 % or more. The Cr content is more preferably 0.80 % or less. The Mo content is more preferably 0.80 % or less. The Ni content is more preferably 0.80 % or less.

[0041] When Co content is 0.010 % or less, the amount of coarse precipitates and inclusions does not increase, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. The Co content is therefore preferably 0.010 % or less. A lower limit of the Co content is not particularly specified. Co is an element that improves hardenability, and therefore the Co content is more preferably 0.001 % or more. The Co content is more preferably 0.008 % or more.

[0042] When Cu content is 1.00 % or less, the amount of coarse precipitates and inclusions does not increase, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. The Cu content is therefore preferably 1.00 % or less. A lower limit of the Cu content is not particularly specified. Cu is an element that improves hardenability, and therefore the Cu content is more preferably 0.01 % or more. The Cu content is more preferably 0.80 % or less.

[0043] When Sn content is 0.200 % or less, cracks are not caused inside the steel sheet during casting or hot rolling and the ultimate deformability of the steel sheet is not decreased, and therefore λ does not decrease. Further, bendability does not decrease. The Sn content is therefore preferably 0.200 % or less. A lower limit of the Sn content is not particularly specified. Sn is an element that improves hardenability, and therefore the Sn content is more preferably 0.001 % or more. The Sn content is more preferably 0.100 % or less.

[0044] When Sb content is 0.200 % or less, the amount of coarse precipitates and inclusions does not increase, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. The Sb content is therefore preferably 0.200 % or less. A lower limit of the Sb content is not particularly specified. Sb is an element that controls surface layer softening thickness and enables strength adjustment, and therefore the Sb content is more preferably 0.001 % or more. The Sb content is more preferably 0.100 % or less.

[0045] When Ca content, Mg content, and REM content are respectively 0.0100 % or less, the amount of coarse precipitates and inclusions does not increase, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. Therefore, Ca content is preferably 0.0100 % or

less. Mg content is preferably 0.0100 % or less. REM content is preferably 0.0100 % or less. Lower limits of Ca, Mg, and REM content are not particularly specified. Ca, Mg, and REM spheroidize the shape of nitrides and sulfides and improve steel sheet ultimate deformability, and therefore the content of each of Ca, Mg, and REM is respectively preferably 0.0005 % or more. The Ca content is more preferably 0.0050 % or less. The Mg content is more preferably 0.0050 % or less. The REM content is more preferably 0.0050 % or less.

[0046] When Zr content and Te content are respectively 0.100 % or less, the amount of coarse precipitates and inclusions does not increase, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. Therefore, Zr content is preferably 0.100 % or less. Te content is preferably 0.100 % or less. Lower limits of Zr and Te content are not particularly specified. Zr and Te are elements that spheroidize the shape of nitrides and sulfides and improve steel sheet ultimate deformability, and therefore the content of each of Zr and Te is respectively more preferably 0.001 % or more. The Zr content is more preferably 0.080 % or less. The Te content is more preferably 0.080 % or less.

[0047] When Hf content is 0.10 % or less, the amount of coarse precipitates and inclusions does not decrease, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. The Hf content is therefore preferably 0.10 % or less. A lower limit of the Hf content is not particularly specified. Hf is an element that spheroidizes the shape of nitrides and sulfides and improves steel sheet ultimate deformability, and therefore the Hf content is preferably 0.01 % or more. The Hf content is more preferably 0.08 % or less.

[0048] When Bi content is 0.200 % or less, the amount of coarse precipitates and inclusions does not increase, the ultimate deformability of the steel sheet does not decrease, and therefore λ does not decrease. Further, bendability does not decrease. The Bi content is therefore preferably 0.200 % or less. A lower limit of the Bi content is not particularly specified. Bi is an element that reduces segregation, and therefore the Bi content is preferably 0.001 % or more. The Bi content is more preferably 0.100 % or less.

[0049] When the content of each of Ti, Nb, V, Ta, W, B, Cr, Mo, Ni, Co, Cu, Sn, Sb, Ca, Mg, REM, Zr, Te, Hf, and Bi described above is below the respective preferred lower limit, the effect of the present disclosure is not impaired, and therefore such elements are included as inevitable impurity.

[0050] Next, the microstructure of the high strength steel sheet is described.

((Area fraction of martensite: 60 % or more))

[0051] By containing martensite, TS of 980 MPa or more and excellent warm workability become possible to realize. By increasing the area fraction of martensite, as described below, a ductility-improving mechanism caused by a microscopic interaction between a large amount of dislocations in martensite and appropriately controlled solute C can be expressed as macroscopic elongation in warm working. The area fraction of martensite is therefore 60 % or more. The area fraction of martensite is preferably 70 % or more. The area fraction of martensite is more preferably 90 % or more.

[0052] An upper limit of the area fraction of martensite is not particularly limited, and the effects described above are obtainable even at 100 %.

[0053] Martensite is a transformation phase that forms at the Ms temperature or below, with or without tempering. Further, martensite includes lower bainite that forms at the Ms temperature or below.

[0054] The observed position of martensite is at the 1/4 sheet thickness position, as described below.

<<Area fraction of ferrite: 40 % or less>>

[0055] The desired strength and stretch flangeability are obtainable when the area fraction of ferrite is 40 % or less. The effects are obtainable even when the area fraction of ferrite is 0 %. When there is too much ferrite, sufficient martensite cannot be secured, and the desired TS cannot be obtained. Further, the hardness difference between microstructures increases, promoting void generation and consolidation, which reduces the hole expansion ratio and degrades stretch flangeability. The area fraction of ferrite is therefore 40 % or less. The area fraction of ferrite is preferably 30 % or less. The area fraction of ferrite is more preferably 20 % or less.

[0056] Ferrite is a soft bcc iron that forms at temperatures higher than the Ms temperature and includes allotriomorphic ferrite, idiomorphic ferrite, and upper bainite.

[0057] The observation position of ferrite is the 1/4 sheet thickness position (a position corresponding to 1/4 of the thickness in the depth direction from the surface of the steel sheet), as described below.

[0058] The area fraction measurements for martensite and ferrite are as follows.

[0059] First, a sample is cut from the steel sheet so that the thickness cross-section parallel to the rolling direction (L-section at the 1/4 sheet thickness position) becomes the observation plane. The sample observation plane is mirror polished with diamond paste, followed by finish polishing with colloidal silica, and then etched with 1 vol% nital to reveal the microstructure.

[0060] The sample observation plane is then observed using a scanning electron microscope (SEM) at a magnification

of 3000× under the condition of an accelerating voltage of 10 kV to obtain SEM images for three fields of view.

[0061] From the obtained SEM images, the area fraction of each microstructure is calculated using Adobe Photoshop (produced by Adobe Systems). Specifically, the value obtained by dividing the area of each microstructure by the measured area is the area fraction of each microstructure. The area fraction of each microstructure is calculated for three fields of view, and the average of these is the area fraction of each microstructure.

[0062] In the SEM image, ferrite is gray in color and is a flat microstructure region that does not encapsulate carbides having a white contrast. In contrast, martensite has a hierarchical microstructure with fine internal irregularities. Thus, martensite and ferrite can be distinguished from each other.

((Retained austenite area fraction: 20 % or less))

[0063] Good component strength and stretch flangeability are obtainable when the area fraction of retained austenite is 20 % or less. The area fraction of retained austenite is therefore 20 % or less. The area fraction of retained austenite is preferably 15 % or less.

[0064] A lower limit of the area fraction of retained austenite is not particularly limited, and the effects described above are obtainable even at 0 %.

[0065] The area fraction measurement method for retained austenite is as follows.

[0066] First, the steel sheet is ground so that the 1/4 sheet thickness position becomes the measurement surface, and then the sample is further ground by 0.1 mm by chemical polishing to obtain a sample. For the sample measurement surface, the X-ray diffraction device measures the integrated reflection intensity of the (200), (220), and (311) planes of fcc iron (austenite) and the (200), (211), and (220) planes of bcc iron, using a Co K α radiation source. The intensity ratio of the integrated reflection intensity of each plane of fcc iron to the integrated reflection intensity of each plane of bcc iron is determined. The average value of nine intensity ratios is the volume fraction of retained austenite. The volume fraction of retained austenite is considered to be the area fraction of retained austenite.

((Lattice constant of martensite))

[0067] The lattice constant of martensite is a very important microstructure factor according to the present disclosure. When the lattice constant a_M (nm) of martensite at room temperature satisfies Expression 1, the high strength steel sheet having excellent ductility and bendability at room temperature, and excellent warm workability, is obtainable.

$$\text{Expression 1: } 1.00005 \leq a_M/a_R \leq 1.00500$$

[0068] In Expression 1, a_R (nm) is the lattice constant of martensite at room temperature after heat treatment of the high strength steel sheet at 500 °C for 30 min.

[0069] The inventors found that by setting the ratio a_M/a_R between the lattice constant a_M of martensite at room temperature and the lattice constant a_R at room temperature after heat treatment of the high strength steel sheet containing the martensite at 500 °C for 30 min to an appropriate value, the steel sheet having excellent ductility at room temperature and excellent warm workability is obtainable. It is assumed that the presence of an appropriate amount of solute C in martensite that contains a large amount of dislocations responsible for plastic deformation improves both uniform elongation and local elongation during warm working due to the interaction between dislocations and solute C. The lattice constant a_M of martensite is affected not only by C, which is an interstitial solute element, but also by Si and Mn, which are substitutional solute elements. By heat-treating the high strength steel sheet at 500 °C for 30 min, the solute C in martensite precipitates as cementite, and the solute C concentration becomes very low. Therefore, the ratio a_M/a_R between the lattice constant a_M of martensite at room temperature and the lattice constant a_R at room temperature after heat treatment of the high strength steel sheet including martensite at 500 °C for 30 min is a parameter that precisely represents the concentration of solute C in martensite. The concentration of solute C in martensite affects ductility at room temperature and warm workability. Therefore, by controlling a_M/a_R to satisfy Expression 1, the high strength steel sheet having excellent ductility and bendability at room temperature, and excellent warm workability, is obtainable.

[0070] When a_M/a_R is less than 1.00005, the concentration of solute C in martensite is low and the interaction between dislocations in martensite and solute C during warm working becomes small, resulting in poor warm workability. On the other hand, when a_M/a_R exceeds 1.00500, dislocation migration at room temperature is significantly suppressed, resulting in reduced ductility and bendability at room temperature. Therefore, the lattice constant a_M of martensite needs to satisfy $1.00005 \leq a_M/a_R \leq 1.00500$. Preferably, a_M/a_R is 1.00010 or more. Preferably, a_M/a_R is 1.00200 or less.

[0071] The lattice constant of martensite is determined by an X-ray diffraction method. The high strength steel sheet is ground so that the 1/4 sheet thickness position becomes the measurement surface, and then further ground by 0.1 mm by chemical polishing to obtain a sample.

[0072] For the sample measurement surface, an X-ray diffraction device is used with a Cu K α radiation source to calculate the peak positions by fitting the peaks of the (110), (200), (211), (220), (310), and (222) planes of bcc iron using a pseudo-Voigt approximation. From each peak position obtained, assuming that c/a is 1, the lattice constant is calculated from Bragg's law, and the average value is taken as the lattice constant a_M of martensite.

[0073] The lattice constant a_R of martensite is determined for the steel sheet after heat treatment at 500 °C for 30 min, in the same way as for the high strength steel sheet described above.

((Ratio of number of martensitic blocks containing carbide having major axis length of 200 nm or more to number of martensitic blocks containing carbide: 50 % or less))

[0074] Warm workability can be further improved by reducing the ratio of the number of martensitic blocks containing carbide having a major axis length of 200 nm or more to the number of martensitic blocks containing carbide. Carbides that are 200 nm or more in martensitic blocks are thermally stable in terms of interfacial energy and are more likely to deplete surrounding solute C during warm working, resulting in a localized decrease in solute C. Uniform elongation and local elongation during warm working can be further improved by decreasing the number of martensitic blocks containing carbides that are 200 nm or more. The ratio of the number of martensitic blocks containing carbide having a major axis length of 200 nm or more to the number of martensitic blocks containing carbide is therefore preferably 50 % or less. The ratio is more preferably 30 % or less. The ratio is even more preferably 10 % or less. The lower limit of the ratio of the number of martensitic blocks containing carbide having a major axis length of 200 nm or more to the number of martensitic blocks containing carbide is not particularly limited and may be 0 %.

[0075] Carbide contained in a martensitic block is at least one of the following: cementite, epsilon carbide, eta carbide, or chi carbide.

[0076] The method of measuring the ratio of the number of martensitic blocks containing carbide having a major axis length of 200 nm or more to the number of martensitic blocks containing carbide is as follows.

[0077] First, the steel sheet is ground so that the 1/4 sheet thickness position becomes the observation plane, and then the sample is electropolished. The observation plane of the prepared sample is observed using a transmission electron microscope (TEM) at an accelerating voltage of 200 kV. The dislocation density of martensite is significantly higher than that of ferrite or retained austenite, and therefore they can be distinguished from each other by observing the strain contrast in a TEM bright-field image. A martensite block is a hierarchical unit comprising martensite and is a group of laths having the same crystallographic planes and crystal orientation. Therefore, in the TEM bright-field image, the diffraction contrast differs from block to block within martensite, making it possible to distinguish other hierarchical structures such as packets and laths. When an electron beam is incident on a martensitic block in the [100] orientation, an electron diffraction pattern of the martensite matrix phase is obtainable. Adjacent martensitic blocks are distinguished from each other by different contrast in a bright-field image because of different crystal orientation across block boundaries.

[0078] FIG. 1 illustrates an example of an electron diffraction pattern of martensite with carbide present.

[0079] When carbide is present in a single observed martensitic block, an electron diffraction pattern of the carbide is obtainable in addition to the electron diffraction pattern of the matrix phase martensite (α), as illustrated in FIG. 1. In FIG. 1, the black circles indicate electron diffraction spots in the matrix phase martensite when an electron beam is incident from the [100] orientation. Further, white circles indicate electron diffraction spots of carbide. When such a diffraction pattern of matrix phase martensite and carbide is obtained, the martensitic block is considered to be a martensitic block containing carbide. Electron diffraction spots obtained from carbide are used to obtain a dark-field image. In a dark-field image, a metastable carbide exhibits white contrast.

[0080] The major axis is determined from the dark-field image of the carbide, and when the major axis length of the largest carbide present in the martensitic block is 200 nm or more, the martensitic block is considered to be a martensitic block containing a carbide having a major axis length of 200 nm or more. 50 martensitic blocks are observed. The number of martensitic blocks containing carbide having a major axis length of 200 nm or more is divided by the number of martensitic blocks containing carbide, and then multiplied by 100 ((number of martensitic blocks containing carbide having major axis length of 200 nm or more) / (number of martensitic blocks containing carbide) $\times$ 100). The value determined is the ratio [%] of the number of martensitic blocks containing carbide having a major axis length of 200 nm or more to the number of martensitic blocks containing carbide.

⟨⟨Residual microstructure⟩⟩

[0081] The microstructure of the steel sheet may include a microstructure other than the martensite, ferrite, and retained austenite described above (residual microstructure). However, in order to avoid impairing the effects of the present disclosure, the area fraction of the residual microstructure is preferably 3 % or less. Examples of residual microstructure include pearlite, alloy carbonitride precipitated in ferrite, and other microstructure known to be in steel sheets. Note that iron-based carbide present in martensite is not included in the residual microstructure.

[0082] For the residual microstructure, the area fraction of the residual microstructure is calculated using Adobe Photoshop (produced by Adobe Systems) from the SEM images taken when measuring area fractions of martensite and ferrite. Specifically, the value obtained by dividing the area of the residual microstructure by the measured area is the area fraction of the residual microstructure. The area fraction of the residual microstructure is calculated for three fields of view, and the average of these is the area fraction of the residual microstructure.

[0083] In SEM images, pearlite exhibits a layered microstructure consisting of white contrast cementite and gray contrast ferrite, while alloy carbonitride exhibits an angular, dark contrast microstructure.

[0084] The thickness of the high strength steel sheet is not particularly limited and is typically 0.3 mm or more and 2.8 mm or less.

(Coated or plated layer)

[0085] The high strength steel sheet may have a coated or plated layer on a surface thereof. The coated or plated layer is formed by a coating or plating treatment described below. The coated or plated layer is not particularly limited, and examples include a hot-dip coated layer and an electroplated layer. The coated or plated layer may be an alloyed coated or plated layer (alloy-coated or alloy-plated layer).

[0086] Examples of the coated or plated layer include a zinc galvanized layer (Zn coated or plated layer) and an Al coated or plated layer. A galvanized layer is preferred as the coated or plated layer. The galvanized layer may contain elements such as Al and Mg.

[0087] The composition of the coated or plated layer is not particularly limited and may be a typical composition.

[0088] For example, when the coated or plated layer is a hot-dip galvanized layer or a galvanized layer, an example of a typical composition is Fe: 20 mass% or less, Al: 0.001 mass% to 1.0 mass%, at least one selected from the group consisting of Pb, Sb, Si, Sn, Mg, Mn, Ni, Cr, Co, Ca, Cu, Li, Ti, Be, Bi, and REM having a total content from 0 mass% to 3.5 mass%, with the balance being Zn and inevitable impurity.

[0089] When the coated or plated layer is a hot-dip galvanized layer, the coating weight of the coated or plated layer per side is preferably 20 g/m² or more. The coating weight per side is preferably 80 g/m² or less. Further, a galvanized layer obtained by alloying a hot-dip galvanized layer having such a coating weight is also preferred.

[0090] When the coated or plated layer is a hot-dip galvanized layer, the Fe content in the coated or plated layer is preferably less than 7 mass%. When the coated or plated layer is a galvanized layer, the Fe content in the coated or plated layer is preferably 7 mass% or more. Further, when the coated or plated layer is a galvanized layer, the Fe content in the coated or plated layer is preferably 20 mass% or less. The Fe content in the coated or plated layer is more preferably 15 mass% or less.

[Member]

[0091] A member according to an embodiment of the present disclosure is described below. The member is formed using the high strength steel sheet according to an embodiment of the present disclosure described above. The member is, for example, the high strength steel sheet according to an embodiment of the present disclosure, made into a desired shape by a forming process, a joining process, or the like. The member according to an embodiment of the present disclosure is suitable for use as an automobile frame structural component or for use as an automobile reinforcing component. Herein, the high strength steel sheet according to an embodiment of the present disclosure as described above is a high strength steel sheet having a tensile strength of 980 MPa or more and excellent component strength, ductility, stretch flangeability, bendability, and warm workability. Therefore, the member according to an embodiment of the present disclosure is particularly suitable for general use as an automobile frame structural component or an automobile reinforcing component.

[Method of producing high strength steel sheet]

[0092] The following is description of a method of producing a high strength steel sheet.

[0093] First, a steel slab is produced by melting steel material having the chemical composition described above. The method for producing molten steel to be used as a steel slab (steel material) is not particularly limited, and any known method using a converter, electric furnace, or the like may be used. Steel slabs are preferably produced by a continuous casting method to prevent macro-segregation, but may be produced by other methods such as ingot casting or thin slab casting. The high strength steel sheet according to the present embodiment includes a cold-rolled steel sheet (cold-rolled sheet), the production of which is completed by hot rolling, pickling, cold rolling, and annealing, and a high strength steel sheet obtained by coating or plating a cold-rolled steel sheet.

[0094] The steel slab is then hot rolled to obtain a hot-rolled sheet. In one example, the steel slab is once cooled to room temperature and then heated again to apply hot rolling (rough rolling and finish rolling). The steel slab may be charged to a

heating furnace as a warm slab without cooling to room temperature, or may be briefly held at a temperature and then immediately rough-rolled.

(Rough rolling)

[0095] A rough-rolled sheet is obtainable by rough rolling the steel slab under the following conditions.

[0096] The temperature at which the steel slab is heated (slab heating temperature) is preferably 1100 °C or more, from the viewpoint of carbide dissolution and reduction of rolling load. On the other hand, to prevent an increase in scale loss, the slab heating temperature is preferably 1300 °C or less. The slab heating temperature is based on the surface temperature of the steel slab. Next, the steel slab heated to the slab heating temperature is rough rolled under the following conditions.

((Total rolling reduction of 50 % or more at average strain rate of 1×10^{-4} /s or more and 1×10^{-1} /s or less))

[0097] In steel, C, which is an interstitial solute element, interacts with Si and Mn, which are substitutional solute elements. That is, solute atoms such as Si and Mn affect the solute C of martensite. The lattice constant a_M of martensite in the final microstructure can be optimized by optimizing the average strain rate and total rolling reduction during rough rolling. The inventors consider the reason for this discovery to be as follows. During the plastic deformation and dynamic recrystallization process of austenite grains during rough rolling, solute atoms such as Si and Mn are considered to be appropriately distributed by high-speed diffusion through dislocations and crystal grain boundaries. The inventors believe that the appropriate distribution of solute atoms such as Si and Mn results in a uniform amount of solute C in martensite during the partial tempering process described below, and therefore optimizes the lattice constant a_M of martensite in the final microstructure. By optimizing the lattice constant a_M of martensite, the high strength steel sheet having excellent warm workability is obtainable, as described above.

[0098] The average strain rate during rough rolling is defined as the rolling ratio ε (-) from the first mill to the last mill of rough rolling divided by the time t_R (s) taken from the start of rolling at the first mill to the completion of rolling at the last mill of rough rolling (ε/t_R).

[0099] When the average strain rate during rough rolling exceeds 1×10^{-1} /s, plastic deformation of austenite grains and diffusion of solute atoms such as Si and Mn during dynamic recrystallization become insufficient, resulting in a low lattice constant ratio a_M/a_R and poor warm workability. On the other hand, when the average strain rate during rough rolling is less than 1×10^{-4} /s, the recovery of dislocations in austenite grains is promoted and the driving force for recrystallization is decreased, suppressing dynamic recrystallization. This results in insufficient diffusion of solute atoms such as Si and Mn, which lowers the ratio a_M/a_R of the Ms lattice constant and decreased warm workability. Accordingly, the average strain rate during rough rolling is 1×10^{-4} /s or more and 1×10^{-1} /s or less. The average strain rate during rough rolling is preferably 1×10^{-3} /s or more. The average strain rate during rough rolling is preferably 1×10^{-2} /s or less.

[0100] When the total rolling reduction of rough rolling is less than 50 %, plastic deformation of austenite grains and diffusion of solute atoms such as Si and Mn during dynamic recrystallization are insufficient, resulting in a low lattice constant ratio a_M/a_R and poor warm workability. Therefore, the total rolling reduction of rough rolling is 50 % or more. The total rolling reduction of rough rolling is preferably 60 % or more. On the other hand, the total rolling reduction of rough rolling is preferably 90 % or less.

[0101] The rolling finish temperature of rough rolling is not particularly limited. From the viewpoint of completing the recrystallization of austenite grains, the rolling finish temperature of rough rolling is preferably 950 °C or more.

[0102] Next, finish rolling is applied to the rough-rolled sheet to obtain a hot-rolled sheet. The hot-rolled sheet is coiled as appropriate. When the slab heating temperature is low, from the viewpoint of preventing problems during hot rolling, the rough-rolled sheet is preferably heated using a bar heater or the like prior to finish rolling. The temperature at which finish rolling is carried out (rolling finish temperature) is preferably 700 °C or more. This decreases the rolling load. Further, the rolling reduction in the un-recrystallized state of austenite is decreased, and development of an abnormal microstructure elongated in the rolling direction is suppressed, resulting in excellent workability.

[0103] Finish rolling may be carried out continuously by joining rough-rolled sheets together. The rough-rolled sheet may be rolled once prior to carrying out finish rolling.

[0104] Some or all of finish rolling may be conducted as lubrication rolling to decrease rolling load. Lubrication rolling is also preferred from the viewpoint of uniformity of steel sheet shape and material property. The frictional coefficient for lubrication rolling is preferably 0.10 or more. The frictional coefficient for lubrication rolling is preferably 0.25 or less.

[0105] To secure good sheet passing during cold rolling and annealing, which are described later, the coiling temperature after hot rolling is preferably 300 °C or more. The coiling temperature after hot rolling is preferably 700 °C or less.

[0106] The hot-rolled sheet obtained by hot rolling is then pickled as appropriate. Pickling removes oxides from the surface of the hot-rolled sheet, resulting in excellent chemical convertibility and coated or plated layer quality in the final product, the high strength steel sheet. Pickling may be carried out in one or more batches.

[0107] After pickling, the hot-rolled sheet is optionally subjected to softening heat treatment and then cold rolled. A cold-rolled sheet is thereby obtained. Conditions for cold rolling are not limited, but the cumulative rolling reduction ratio is preferably 20 % to 75 %. The number of rolling passes and the rolling reduction in each pass are not particularly limited.

[0108] The cold-rolled sheet obtained as described above is then annealed as described below.

(Annealing)

((Heating temperature: 800 °C or more))

[0109] When the heating temperature is too low, reverse transformation to austenite does not progress sufficiently, resulting in a low area fraction of martensite, and the desired TS cannot be obtained. Further, the amount of martensite, which contributes to warm workability, is small, and therefore excellent warm workability cannot be obtained. For these reasons, the heating temperature is 800 °C or more. Further, the heating temperature is preferably 830 °C or more. An upper limit of the heating temperature is not particularly limited. From the viewpoint of operability and the like, the heating temperature is preferably 1000 °C or less. The heating temperature is based on the temperature of the steel sheet surface.

[0110] The time for which the cold-rolled sheet is heated at the heating temperature (heating time) is not particularly limited, but when too short, there is a risk that the reverse transformation to austenite may not progress sufficiently. Therefore, the heating time is preferably 30 s or longer. The heating time is more preferably 60 s or longer. An upper limit of the heating time is not particularly limited. For example, the heating time is preferably 6000 s or less. The heating time is more preferably 3000 s or less. Note that "s" means seconds.

((Cooling under the condition that residence time t_1 in temperature range T_1 from M_s temperature to 700 °C is 1000 s or less))

[0111] After heating, the cold-rolled sheet is cooled through a temperature range T_1 from 700 °C to the M_s temperature. When the time that the cold-rolled sheet remains in the temperature range T_1 (residence time t_1) is too long, excessive ferrite transformation occurs and the area fraction of ferrite becomes high, and the desired TS and good stretch flangeability cannot be obtained. The residence time t_1 is therefore 1000 s or less. The residence time t_1 is preferably 800 s or less. The residence time t_1 is more preferably 600 s or less. A lower limit of the residence time t_1 is not particularly limited. From the viewpoint of reducing a capital investment burden, the residence time t_1 is preferably 1 s or longer. The residence time is more preferably 5 s or longer. The residence time is even more preferably 10 s or longer. The temperature range T_1 is based on the temperature of the steel sheet surface.

[0112] The M_s temperature is determined by the following Expression (1).

$$M_s = 499 - 317 [\%C] - 11 [\%Si] - 33 [\%Mn] - 17 [\%Ni] - 28 [\%Cr] - 11 [\%Mo] \quad (1)$$

[0113] Here, [%M] indicates M content in the steel (mass%).

[0114] After annealing, a partial tempering process is carried out, cooling the cold-rolled sheet from the M_s temperature to 50 °C or less.

((Partial tempering process in which partial tempering parameter S (μm^2) satisfies Expression 2))

[0115]

Expression 2:

$$0.01 \leq S \leq 30$$

Expression 3:

$$S = \sum_{t=1}^{t_E} f(t) D(T) \times 10^{12}$$

Expression 4:

$$f(t) = 1 - \exp(-1.1 \times 10^{-2} (M_s - T_{min}))$$

Expression 5:

$$D(T) = 0.394 \times 10^{-6} \exp \left(-\frac{80220}{8.314 \times (T+273)} \right)$$

[0116] In Expressions 2 to 5,

a time when the temperature of the cold-rolled sheet first reaches the Ms temperature after the annealing is defined as $t = 0$ (s) and $t = t_E$ (s) is the time when the partial tempering process is completed and the temperature of the cold-rolled sheet is 50 °C.

T (°C) is an average temperature of the cold-rolled sheet at time $t-1$ to t (s).

$T_{\min}$ (°C) is the lowest temperature of T at time 0 to t (s).

M_s (°C) indicates the Ms temperature of the high strength steel sheet.

[0117] The various temperatures in partial tempering are based on temperature of the steel sheet surface.

[0118] After passing through the temperature range T_1 , the steel sheet reaches the Ms temperature and the martensite transformation proceeds. After the martensite transformation, a tempering phenomenon occurs in which solute C precipitates as carbide, and solute C decreases. Accordingly, in the temperature range after reaching the Ms temperature for the first time, martensite tempering always progresses and solute C decreases. Therefore, in the entire range from the first time the Ms temperature is reached to 50 °C, where tempering is negligible, the inventors found it necessary to consider the temperature, time, and the volume fraction of martensite already formed at that temperature. Therefore, the process of carrying out tempering in consideration of the volume fraction of martensite already formed at that temperature is defined as a partial tempering process.

[0119] The inventors have found that by defining a partial tempering parameter S and controlling the value of S appropriately, the ratio a_M/a_R of the lattice constant can be kept within a defined range of values. The parameter S is the sum of the amount of tempering of already formed martensite per unit time, which is a combination of the volume fraction $f(t)$ of martensite at the temperature T (°C) at time t (s) and the diffusion coefficient $D(T)$ of C, over the thermal hysteresis from when the Ms temperature is first reached to 50 °C, at which C diffusion can be ignored. The time when the cold-rolled sheet first reaches the Ms temperature after annealing is defined as $t = 0$ (s), and $t = t_E$ (s) is the time when partial tempering is completed and the temperature reaches 50 °C. T (°C) is an average temperature of the cold-rolled sheet at time $t-1$ to t (s). $T_{\min}$ (°C) is the lowest temperature of T at time 0 to t (s). M_s (°C) indicates the Ms temperature of the high strength steel sheet.

[0120] When S (μm^2) is too small, the partial tempering process is insufficient and the ratio a_M/a_R of the lattice constant becomes high, resulting in decreased ductility and bendability at room temperature. Therefore, S (μm^2) is 0.01 or more. S (μm^2) is preferably 0.03 or more. S (μm^2) is more preferably 0.10 or more. When S (μm^2) is too large, the partial tempering process is excessive, resulting in a lower ratio a_M/a_R of the lattice constant and decreased warm workability. S (μm^2) is therefore 30 or less. S (μm^2) is preferably 15 or less. S (μm^2) is more preferably 5 or less.

[0121] The heating pattern in the partial tempering process is not particularly limited, as long as the partial tempering parameter S is in a range described above. As examples, the partial tempering process may be the heating patterns illustrated in FIG. 2 and FIG. 3.

[0122] FIG. 2 illustrates a schematic diagram of thermal hysteresis of the partial tempering process of continuously cooling from the Ms temperature to 50 °C or less, and a schematic diagram of $f(t)$. In this example, the heating pattern from the Ms temperature is one of continuous cooling from the Ms temperature to 50 °C or less. Cooling methods are not particularly limited, and a known cooling method such as gas cooling may be used. The cooling rate for continuous cooling is not particularly limited, and may be in a range from 1 °C/s to 100 °C/s, for example.

[0123] As illustrated in FIG. 2, the time when the Ms temperature is reached for the first time after annealing is defined as $t = 0$ (s). In the case of continuous cooling as illustrated in FIG. 2, $t = t_E$ (s) is the time when the temperature reaches 50 °C for the first time after the annealing. When $t = t_a$, $T_{\min}$ (°C) is the lowest temperature T at time 0 to t (s), that is, the temperature at $t = t_a$.

[0124] (In partial tempering process, cooling is stopped at cooling stop temperature from room temperature to Ms temperature, followed by reheating to reheating temperature and then cooling to 50 °C or less (suitable conditions))

[0125] In the partial tempering process, cooling may be stopped at any temperature from the Ms temperature to room temperature, followed by reheating to a reheating temperature and then cooling to 50 °C or less. This decreases a hardness difference between microstructures and further improves bendability and stretch flangeability. FIG. 3 illustrates a schematic diagram of the thermal hysteresis of the partial tempering process in which cooling is stopped at a temperature that is the Ms temperature or less and room temperature or more, followed by heating and then cooling to 50 °C or less. When $t = t_b$, $T_{\min}$ (°C) is the cooling stop temperature in the example in FIG. 3. When $T_{\min}$ (°C) is less than 50 °C, the time after reheating when the steel sheet is again at 50 °C is defined as $t = t_E$ (s).

[0126] After annealing, the cold-rolled sheet is cooled to the cooling stop temperature that is room temperature or more and the Ms temperature or less. The cooling stop temperature is not particularly limited. To allow sufficient martensite formation, promote partial tempering, and achieve suitable warm workability, the cooling stop temperature is preferably 250 °C or less. The cooling stop temperature may be, for example, room temperature. The cooling method is also not particularly limited, and gas cooling, water cooling, or the like may be used. The cooling rate is not particularly limited, and may be in the range from 1 °C/s to 100° C/s, for example.

[0127] Next, the cold-rolled sheet is reheated to the reheating temperature. The heating method for reheating to the reheating temperature is not particularly restricted. Induction heating (IH) may be used, for example.

[0128] The reheating temperature is preferably 500 °C or less because this suppresses the recovery phenomenon of dislocations in martensite and results in suitable strength. The reheating temperature is preferably 130 °C or more to promote C diffusion and partial tempering in a short time to achieve suitable ductility and bendability. After reheating, the temperature may be held constant until cooling.

[0129] The cold-rolled sheet is then cooled to 50 °C or less. Cooling methods are not particularly limited. A known method such as water cooling or gas cooling, for example, may be used. The cooling rate is not particularly limited, and may be in the range from 1 °C/s to 100° C/s, for example. After reaching 50 °C or less, the cold-rolled sheet is cooled to room temperature by any cooling method. Cooling methods are not particularly limited. A known method such as gas cooling, air cooling, water cooling, or the like may be used.

[0130] The partially tempered cold-rolled sheet is then cooled to room temperature. Thus, the high strength steel sheet (cold-rolled steel sheet) described above is obtained.

[0131] When the coating or plating treatment described below is applied in the present production method, the resulting high strength steel sheet is a coated or plated steel sheet including a coated or plated layer.

[0132] For the series of heat treatment in the present production method, as long as the thermal hysteresis described above is satisfied, other conditions are not particularly limited, nor is the line, apparatus, or the like where the heat treatment is carried out.

(Coating or plating treatment)

[0133] In the present production method, a coating or plating treatment may be applied to the cold-rolled sheet.

[0134] Examples of the coating or plating treatment include hot-dip galvanizing treatment (a treatment to form a hot-dip galvanized layer) and galvannealing treatment (a treatment to form a galvannealed layer by performing an alloying treatment after the hot-dip galvanizing treatment). An electroplated layer may be formed by an electroplating treatment.

[0135] When applying hot-dip galvanizing treatment, it is preferable to dip the cold-rolled sheet in a galvanizing bath, followed by gas wiping or the like to adjust the coating weight of the coated layer. The bath temperature of the galvanizing bath is not particularly limited. The bath temperature is preferably 440 °C or more. The bath temperature is preferably 500 °C or less.

[0136] The amount of Al in the galvanizing bath is preferably 0.10 mass% or more. The amount of Al in the galvanizing bath is preferably 0.23 mass% or less.

[0137] The galvanizing treatment is preferably carried out during holding in the temperature range T1 from the Ms temperature to 700 °C during annealing, as described above.

[0138] The temperature of the alloying treatment is preferably 470 °C or more for a more suitable Zn-Fe alloying rate and more suitable productivity. Further, in order to suitably avoid untransformed austenite transforming to pearlite and to make TS more suitable, the temperature of the alloying treatment is preferably 600 °C or less. The temperature of the alloying treatment is more preferably 560 °C or less. The standard temperature of the alloying treatment is 530 °C.

[0139] After the coating or plating treatment, the coated or plated steel sheet may be subjected to skin pass rolling. From the viewpoint of increasing yield strength, the rolling reduction of the skin pass rolling is preferably 0.05 % or more. An upper limit of the rolling reduction is not particularly limited. From the viewpoint of productivity, the rolling reduction is preferably 1.50 % or less. The skin pass rolling may be carried out on-line or off-line. The skin pass to the desired rolling reduction may be carried out in one pass, or may be carried out in a plurality of passes.

[0140] From the viewpoint of productivity, it is preferable to perform a series of processes such as the annealing and the coating or plating treatment described above in a continuous annealing line (CAL) or continuous galvanizing line (CGL).

[0141] Production conditions other than those described above may be determined according to conventional methods.

[Method of producing member]

[0142] The high strength steel sheet described above may be subjected to at least one of forming or joining to produce a member. Forming and joining may be carried out using a conventional method.

EXAMPLES

[0143] The present disclosure is described in detail below by reference to examples. However, the present disclosure is not limited to the examples described below.

(Steel sheet production)

[0144] Molten steel having the chemical compositions listed in Table 1 below (the balance being Fe and inevitable impurity) was melted in a converter and steel slabs were obtained by a continuous casting method.

[0145] The steel slabs were subjected to hot rolling to obtain hot-rolled sheets. Specifically, each steel slab was heated to 1250 °C, rough rolled at an average strain rate for a total rolling reduction as listed in Table 2, followed by finish rolling at a rolling finish temperature of 900 °C, then coiled at 500 °C, and then cooled to room temperature to obtain a hot-rolled sheet. After pickling, the obtained hot-rolled sheet was subjected to softening heat treatment at 500 °C, followed by cold rolling at a rolling ratio of 50 %. Thus, a cold-rolled sheet having a thickness of 1.6 mm was obtained.

[0146] The cold-rolled steel sheet was subjected to annealing and partial tempering treatment under the conditions listed in Table 2 below to obtain the high strength steel sheet (cold-rolled steel sheet) according to the present disclosure. In the partial tempering process, in the case of continuous cooling from the Ms temperature to 50 °C or less, "Continuous" is noted in the "Thermal hysteresis during partial tempering" column in Table 2. In the partial tempering process, in the case of cooling being stopped at a temperature that is the Ms temperature or less and room temperature or more, followed by reheating and then cooling to 50 °C or less, "Reheating" is noted in the "Thermal hysteresis during partial tempering" column in Table 2. The heating time at the heating temperature for the cold-rolled sheet was 200 s.

((Coating or plating treatment))

[0147] For some cold-rolled sheets, hot-dip galvanizing treatment was carried out during the residence in the temperature range T1 (Ms °C to 700 °C) to form a coated or plated layer (hot-dip galvanized layer) on both sides. In other words, a hot-dip galvanized steel sheet (GI) was obtained. For the hot-dip galvanizing treatment, a hot-dip galvanizing bath (bath temperature: 470 °C) was used that contained Al: 0.20 mass% with the balance being Zn and inevitable impurity. The coating weight per side of the hot-dip galvanized layer was about 45 g/m² to 72 g/m². The composition of the hot-dip galvanized layer formed was Fe: 0.1 mass% to 1.0 mass% and Al: 0.2 mass% to 1.0 mass%, with the balance being Zn and inevitable impurity.

[0148] For another portion of the cold-rolled sheets, galvannealing treatment was carried out during the residence in the temperature range T1 (Ms °C to 700 °C) during annealing to form a coated layer (galvannealed layer) on both sides. That is, a galvannealed steel sheet (GA) was obtained. For the hot-dip galvanizing treatment, a hot-dip galvanizing bath (bath temperature: 470 °C) was used that contained Al: 0.14 mass% with the balance being Zn and inevitable impurity. The alloying treatment was carried out at 550 °C. The coating weight per side of the galvannealed layer was about 45 g/m². The composition of the galvannealed layer formed was Fe: 7 mass% to 15 mass% and Al: 0.1 mass% to 1.0 mass%, with the balance being Zn and inevitable impurity.

[0149] In the "Coating or plating type" column of Table 2 below, the case where a hot-dip galvanized layer was formed is indicated as "GI", the case where a galvannealed layer was formed is indicated as "GA", and the case where no coated or plated layer was formed is indicated as "CR".

((Observation of steel microstructure))

[0150] For the obtained steel sheets, the area fractions of martensite, ferrite, and retained austenite, the lattice constant of martensite, the ratio of the number of martensite blocks containing carbide having a major axis length of 200 nm or more to the number of martensitic blocks containing carbide, and the residual microstructure were measured according to the methods described above. The results are listed in Table 3 below.

((Evaluation))

[0151] The obtained steel sheets were subjected to the tests described below to evaluate various properties. The results are listed in Table 3 below.

((Tensile test))

[0152] Tensile tests were conducted in accordance with JIS Z 2241:2021.

[0153] Specifically, a JIS No. 5 test piece was taken from the obtained steel sheet so that the longitudinal direction was

perpendicular to the rolling direction of the steel sheet. Yield strength (YS) [MPa], tensile strength (TS) [MPa], and total elongation (El) [%] were measured by a tensile test at a crosshead speed of 1.67×10^{-1} mm/s using the collected test piece. Further, the yield ratio (YR) ($= 100 \times \text{YS/TS}$) [%] was calculated. When the tensile strength (TS) was 980 MPa or more, this was considered to be high strength. When the yield ratio (YR) was 60 % or more, this was considered to be excellent component strength. When the total elongation (El) was 6.0 % or more, this was considered to be excellent ductility.

((Hole expanding test))

[0154] The hole expanding test was conducted in accordance with JIS Z 2256:2020.

[0155] Specifically, each obtained steel sheet was sheared to obtain a 100 mm $\times$ 100 mm test piece. A 10 mm diameter hole was punched through the collected test piece with a clearance of 12.5 %. Subsequently, using a die having an inner diameter of 75 mm, a conical punch having an apex angle of 60° was pressed into the hole while applying a blank holding force of 9 tonnes (88.26 kN), and the hole diameter D_f [mm] at the crack initiation limit was measured. The initial hole diameter was D_0 [mm], and the hole expansion ratio λ [%] was obtained from the following Expression (6).

$$\text{Expression 6: } \lambda = \{(D_f - D_0)/D_0\} \times 100$$

[0156] When the hole expansion ratio (λ) is 30 % or more, stretch flangeability is considered excellent.

((Bend test))

[0157] Bend tests were conducted in accordance with JIS Z 2248:2022. Specifically, test pieces of strips 30 mm wide and 100 mm long were taken from the obtained steel sheets so that a direction parallel to the rolling direction of the steel sheet was the axial direction of the bend test. The rolling direction end face of each test piece was the ground end face. Each test piece was subjected to a 90° V-bend test under a set of conditions including a pressing load of 100 kN and a hold time of 5 s.

[0158] Bend tests were carried out on five test pieces at appropriate values of bend radius R. Next, the presence or absence of crack initiation at the tip of the bend test piece was checked.

[0159] The presence or absence of crack initiation was checked by observing the ridge at the tip of each bend test piece using a digital microscope (RH-2000, produced by Hirox Co. Ltd.) at 40 $\times$ magnification.

[0160] The minimum bend radius R at which no cracks occurred in any of the five test pieces was determined and divided by the sheet thickness t (R/t) to obtain the limit bending radius. When the limit bending radius (R/t) was 5.0 or less, this was considered to be excellent bendability.

((Warm workability))

[0161] Tensile tests were conducted in accordance with JIS Z 2241:2021.

[0162] Specifically, a JIS No. 5 test piece was taken from the obtained steel sheet so that the longitudinal direction was perpendicular to the rolling direction of the steel sheet. Tensile tests were carried out on collected test pieces at room temperature and 200 °C with a crosshead speed of 1.67×10^{-1} mm/s. Yield strength (YS) [MPa], tensile strength (TS) [MPa], uniform elongation (U-El) [%], and local elongation (L-El) [%] were measured. In the tensile test at 200 °C, the test piece was held at 200 °C for 10 min in a furnace attached to the tensile testing machine, and then the tensile test was started. The uniform elongation and local elongation determined by tensile testing at room temperature were U-El_{RT} and L-El_{RT}, respectively. The uniform elongation and local elongation determined by tensile testing at 200 °C were U-El₂₀₀ and L-El₂₀₀, respectively. When U-El₂₀₀/U-El_{RT} was 1.10 or more and L-El₂₀₀/L-El_{RT} was 1.30 or more, this was considered to be excellent warm workability. Underlining in Tables 1 to 3 below indicates a value outside the appropriate range of the present disclosure.

[Table 1]

[0163]

Steel sample ID	Chemical composition [mass%]																										Remarks			
	C	Si	Mn	P	S	Al	N	O	Ti	Nb	V	Ta	W	B	Cr	Mo	Ni	Co	Cu	Sn	Sb	Ca	Mg	REM	Zr	Te		Hf	Bi	
A	0.191	1.01	3.05	0.017	0.0023	0.040	0.0032	0.0008	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
B	0.195	0.45	2.89	0.015	0.0027	0.036	0.0045	0.0012	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
C	0.196	0.21	1.24	0.019	0.0028	0.031	0.0046	0.0008	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
D	0.078	0.05	1.92	0.019	0.0010	0.024	0.0046	0.0007	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
E	0.310	0.51	2.68	0.019	0.0028	0.043	0.0044	0.0012	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
F	0.162	1.74	2.64	0.016	0.0028	0.023	0.0033	0.0008	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
G	0.200	0.31	0.75	0.013	0.0015	0.045	0.0042	0.0011	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
H	0.196	0.61	3.87	0.009	0.0016	0.021	0.0044	0.0012	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
I	0.025	1.02	2.76	0.017	0.0010	0.037	0.0046	0.0013	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Comparative steel	
J	0.230	2.57	3.21	0.014	0.0030	0.027	0.0047	0.0009	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Comparative steel	
K	0.134	0.98	0.08	0.010	0.0020	0.029	0.0045	0.0009	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Comparative steel	
L	0.178	0.68	5.12	0.020	0.0012	0.030	0.0045	0.0012	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Comparative steel	
M	0.201	0.65	2.98	0.010	0.0014	0.046	0.0049	0.0006	0.032	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel	
N	0.198	0.25	2.79	0.013	0.0025	0.046	0.0044	0.0010	-	0.019	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
O	0.345	0.47	2.67	0.019	0.0012	0.040	0.0046	0.0006	-	-	0.12	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
P	0.201	0.75	2.76	0.014	0.0018	0.036	0.0044	0.0009	0.020	-	-	-	-	0.002	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
Q	0.187	0.54	2.54	0.019	0.0013	0.038	0.0035	0.0013	-	-	-	-	-	-	-	0.21	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
R	0.114	0.60	2.45	0.019	0.0029	0.029	0.0042	0.0010	-	-	-	-	-	-	0.59	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
S	0.141	0.64	2.89	0.014	0.0024	0.028	0.0038	0.0009	-	-	-	-	-	0.05	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
T	0.185	0.53	2.78	0.018	0.0017	0.035	0.0036	0.0009	-	-	-	-	-	-	-	-	0.1	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
U	0.191	0.61	3.02	0.019	0.0020	0.032	0.0044	0.0015	-	-	-	-	-	-	-	-	-	-	-	-	0.010	-	-	-	-	-	-	-	-	Disclosed steel
V	0.321	0.54	3.04	0.018	0.0012	0.035	0.0045	0.0011	-	-	-	-	-	-	-	-	-	-	-	0.01	-	-	-	-	-	-	-	-	-	Disclosed steel
W	0.145	0.97	3.14	0.013	0.0022	0.035	0.0033	0.0009	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
X	0.198	0.62	2.79	0.012	0.0024	0.031	0.0042	0.0005	0.020	0.018	-	-	-	0.002	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
Y	0.310	0.21	1.24	0.020	0.0025	0.033	0.0042	0.0011	-	-	-	-	-	-	-	-	-	-	-	-	-	-	0.0010	-	-	-	-	-	-	Disclosed steel
Z	0.131	0.64	3.10	0.013	0.0023	0.027	0.0037	0.0006	-	-	-	-	-	-	-	-	-	0.0050	-	-	-	-	-	-	0.0210	-	-	-	-	Disclosed steel
AA	0.187	0.65	2.87	0.017	0.0021	0.021	0.0048	0.0011	-	-	-	-	-	-	-	-	-	-	-	-	-	-	0.001	-	-	-	-	-	-	Disclosed steel
AB	0.187	1.01	3.04	0.019	0.0027	0.032	0.0047	0.0008	-	-	-	-	-	-	-	-	-	0.13	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
AC	0.210	0.14	2.78	0.014	0.0017	0.023	0.0047	0.0015	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	0.020	-	-	-	Disclosed steel
AD	0.185	0.26	2.98	0.013	0.0014	0.033	0.0041	0.0009	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	Disclosed steel
AE	0.203	0.71	2.94	0.015	0.0016	0.019	0.0042	0.0013	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	-	0.030	-	Disclosed steel
AF	0.200	0.51	2.73	0.006	0.0009	0.034	0.0032	0.0004	-	-	-	-	-	-	-	-	-	-	-	-	-	0.0015	-	-	-	-	-	-	-	Disclosed steel

Underlined: indicates value outside scope of present disclosure.

[Table 2]

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Table 2

No.	Steel sample ID	Average strain rate of rough rolling [s ⁻¹]	Total rolling reduction of rough rolling [%]	Heating temp. [°C]	Residence time t ₁ (temp. range T ₁) [s]	Thermal hysteresis during partial tempering	Cooling stop temp. [°C]	Reheating temp. [°C]	Partial tempering parameter S [μm ²]	Coating or plating type	Remarks
1	A	6E-03	72	900	115	Continuous	-	-	0.1500	GA	Example
2	A	6E-02	76	900	115	Continuous	-	-	0.1500	GA	Example
3	A	6E-01	76	900	115	Continuous	-	-	0.1500	GA	Comparative Example
4	A	6E-04	80	900	115	Continuous	-	-	0.1500	GA	Example
5	A	6E-05	70	900	115	Continuous	-	-	0.1500	GA	Comparative Example
6	A	6E-03	54	900	115	Continuous	-	-	0.1500	GA	Example
7	A	8E-03	40	900	115	Continuous	-	-	0.1500	GA	Comparative Example
8	A	6E-03	72	780	115	Continuous	-	-	0.1500	GA	Comparative Example
9	A	8E-03	72	900	750	Continuous	-	-	0.1500	GA	Example
10	A	6E-03	72	900	1500	Continuous	-	-	0.1500	GA	Comparative Example
11	A	6E-03	72	900	115	Continuous	-	-	0.0150	GA	Example
12	A	8E-03	76	900	115	Continuous	-	-	0.0006	GA	Comparative Example
13	A	6E-03	76	900	115	Reheating	50	380	8.0000	GA	Example
14	A	6E-03	76	900	115	Reheating	50	420	18.0000	GA	Example
15	A	6E-03	80	900	115	Reheating	50	450	35.0000	GA	Comparative Example
16	B	6E-03	72	900	100	Continuous	-	-	0.1500	GA	Example
17	C	6E-03	72	875	4	Reheating	25	180	0.2000	CR	Example
18	D	8E-03	72	900	240	Reheating	25	300	1.0000	GA	Example
19	E	6E-03	72	900	115	Continuous	-	-	0.3000	GA	Example
20	F	6E-03	72	950	90	Continuous	-	-	0.3000	GA	Example
21	G	6E-03	72	875	4	Reheating	40	180	0.2000	CR	Example
22	H	6E-03	72	850	550	Continuous	-	-	0.1900	GA	Example

[0164]

(continued)

No.	Steel sample ID	Average strain rate of rough rolling [s ⁻¹]	Total rolling reduction of rough rolling [%]	Heating temp. [°C]	Residence time t ₁ (temp. range T ₁) [s]	Thermal hysteresis during partial tempering	Cooling stop temp. [°C]	Reheating temp. [°C]	Partial tempering parameter S [μm ²]	Coating or plating type	Remarks
23	<u>I</u>	6E-03	72	900	90	Continuous	-	-	0.3000	GA	Comparative Example
24	<u>J</u>	8E-03	76	900	80	Continuous	-	-	0.3000	GA	Comparative Example
25	<u>K</u>	6E-03	76	875	4	Reheating	40	250	0.2000	CR	Comparative Example
26	<u>L</u>	6E-03	80	900	400	Reheating	50	350	4.5000	GA	Comparative Example
27	M	8E-03	76	860	115	Continuous	-	-	0.2100	GA	Example
28	N	6E-03	76	875	115	Reheating	40	300	1.4000	GA	Example
29	O	8E-03	80	900	115	Reheating	40	320	2.0000	GA	Example
30	P	6E-03	72	900	115	Reheating	200	310	1.0000	GA	Example
31	Q	6E-03	72	900	115	Continuous	-	-	0.1500	GA	Example
32	R	6E-03	72	900	120	Reheating	40	300	1.0000	GA	Example
33	S	6E-03	72	900	130	Continuous	-	-	0.0800	GA	Example
34	T	6E-03	72	875	4	Reheating	40	280	0.2000	CR	Example
35	U	8E-03	72	900	115	Continuous	-	-	0.0900	GA	Example
36	v	6E-03	80	900	115	Continuous	-	-	0.1500	GA	Example
37	w	8E-03	68	900	115	Reheating	50	300	1.0000	GI	Example
38	X	6E-03	72	890	115	Reheating	40	340	1.0000	GA	Example
39	Y	6E-03	68	900	115	Reheating	100	340	1.0000	GA	Example
40	Z	8E-03	72	900	70	Continuous	-	-	1.0000	GA	Example
41	AA	6E-03	72	900	60	Continuous	-	-	0.0400	CR	Example
42	AB	6E-03	72	900	30	Reheating	100	300	0.5000	GA	Example
43	AC	6E-03	72	950	115	Continuous	-	-	0.2000	GA	Example
44	AD	6E-03	70	950	115	Reheating	40	290	1.0000	GA	Example
45	AE	8E-03	70	900	115	Reheating	200	320	1.1000	GA	Example

(continued)

No.	Steel sample ID	Average strain rate of rough rolling [s ⁻¹]	Total rolling reduction of rough rolling [%]	Heating temp. [°C]	Residence time t ₁ (temp. range T ₁) [s]	Thermal hysteresis during partial tempering	Cooling stop temp. [°C]	Reheating temp. [°C]	Partial tempering parameter S [μm ²]	Coating or plating type	Remarks
46	AF	8E-03	72	900	80	Continuous	-	-	0.1500	GA	Example
Underlined: indicates value outside scope of present disclosure.											

[Table 3]

[0165]

No.	Steel sample ID	Area fraction [%]				a _M /a _r	Ratio of martensitic blocks containing carbides having major axis length 200 nm or more [%]	Properties at room temperature								Warm workability					Remarks				
		Martensite	Ferrite	Retained austenite	Residual micro-structure			YS [MPa]	TS [MPa]	YR [%]	T-EI [%]	U-E _{lkr} [%]	L-E _{lkr} [%]	λ [%]	Bendability R/t	U-E _{l000} [%]	L-E _{l000} [%]	U-E _{l000} /U-E _{lkr}	L-E _{l000} /L-E _{lkr}						
1	A	97	0	1	2	1.00065	0	1134	1532	74	9.9	5.5	4.3	35	3.9	6.5	9.1	1.17	2.11						Example
2	A	97	0	2	1	1.00008	5	1169	1568	75	9.9	5.6	4.2	37	3.9	6.3	5.8	1.12	1.38						Example
3	A	95	1	2	2	<u>1.00004</u>	7	1112	1525	73	9.8	5.6	4.2	39	3.6	5.6	5.0	<u>1.04</u>	<u>1.21</u>						Comparative Example
4	A	95	0	3	2	1.00007	4	1125	1491	75	10.0	5.6	4.4	35	3.6	6.2	5.9	1.11	1.33						Example
5	A	96	0	3	1	<u>1.00004</u>	4	1170	1535	76	9.7	5.4	4.3	34	3.6	5.5	5.2	<u>1.03</u>	<u>1.21</u>						Comparative Example
6	A	95	0	3	2	1.00006	2	1197	1511	79	9.8	5.4	4.5	34	3.9	6.0	5.9	1.12	1.32						Example
7	A	97	0	2	1	<u>1.00001</u>	2	1130	1600	71	9.7	5.4	4.3	39	3.6	5.4	4.9	<u>1.03</u>	<u>1.21</u>						Comparative Example
8	A	<u>41</u>	<u>47</u>	8	4	1.00035	0	578	<u>789</u>	73	15.3	9.2	6.1	32	3.6	9.2	7.5	<u>1.03</u>	<u>1.23</u>						Comparative Example
9	A	63	34	2	1	1.00037	2	695	1002	69	12.5	6.9	5.6	31	3.6	8.0	8.0	1.16	1.43						Example
10	A	<u>44</u>	<u>46</u>	8	2	1.00065	2	620	<u>954</u>	65	13.2	6.9	5.2	<u>25</u>	3.6	7.8	7.8	1.16	1.50						Comparative Example
11	A	95	2	1	2	1.00450	2	1067	1623	66	6.4	3.4	3.0	33	4.6	4.6	6.1	1.35	2.03						Example
12	A	95	0	3	2	<u>1.00541</u>	0	1207	1501	80	<u>4.7</u>	3.0	1.7	35	<u>5.4</u>	3.8	4.1	1.27	2.41						Comparative Example
13	A	95	0	3	2	1.00009	24	1102	1603	69	10.1	5.7	4.4	43	3.2	6.5	6.0	1.13	1.36						Example
14	A	95	0	3	2	1.00007	52	1193	1513	79	9.9	5.5	4.4	45	3.2	6.1	5.8	1.11	1.32						Example
15	A	98	0	1	1	<u>1.00003</u>	55	1048	1387	76	9.1	4.9	4.2	46	3.2	4.5	4.7	<u>0.92</u>	<u>1.11</u>						Comparative Example
16	B	95	0	3	2	1.00051	0	1208	1512	80	10.1	5.5	4.6	34	3.6	6.7	9.2	1.22	2.00						Example
17	C	100	0	0	0	1.00037	2	1321	1515	87	8.5	3.9	4.6	45	3.2	5.9	9.2	1.51	2.00						Example
18	D	68	32	0	0	1.00008	4	801	982	82	6.2	3.1	3.1	48	2.9	3.5	4.9	1.13	1.58						Example
19	E	89	0	10	1	1.00058	2	1114	1802	62	10.1	5.8	4.3	31	4.6	6.9	7.6	1.19	1.78						Example
20	F	84	3	11	2	1.00081	2	1041	1657	63	10.3	5.9	4.4	31	3.6	7.1	6.9	1.20	1.56						Example
21	G	71	25	2	2	1.00045	14	674	1023	66	10.1	5.6	4.5	41	3.2	6.3	5.9	1.13	1.31						Example
22	H	86	0	12	2	1.00122	2	998	1621	62	9.5	5.4	4.2	31	3.6	6.5	8.6	1.21	2.06						Example
23	I	<u>29</u>	<u>67</u>	2	2	<u>1.00001</u>	0	475	<u>594</u>	80	28.8	13.4	15.4	<u>28</u>	3.2	10.5	15.2	<u>0.78</u>	<u>0.99</u>						Comparative Example

Underlined: indicates value outside scope of present disclosure.

No.		Steel sample ID	Area fraction [%]				a _m /a _r	Ratio of martensitic blocks containing carbides having major axis length 200 nm or more [%]	Properties at room temperature								Warm workability					Remarks
			Martensite	Ferrite	Retained austenite	Residual micro-structure			Y _S [MPa]	TS [MPa]	YR [%]	T-EI [%]	U-E _{kT} [%]	L-E _{kT} [%]	λ [%]	Bendability R/t	U-E _{b00} [%]	L-E _{b00} [%]	U-E _{b00} /U-E _{kT}	L-E _{b00} /L-E _{kT}		
24	J	65	12	21	2	1.00046	0	965	1681	57	10.7	6.2	4.5	21	3.6	7.2	9.3	1.16	2.08	Comparative Example		
25	K	45	46	3	6	1.00053	0	700	870	80	10.6	6.3	4.3	47	3.2	6.1	8.5	0.97	1.95	Comparative Example		
26	L	51	24	23	2	1.00028	0	874	1533	57	11.3	7.1	4.2	24	3.2	8.0	8.4	1.13	1.99	Comparative Example		
27	M	95	1	2	2	1.00019	4	1139	1580	72	9.7	5.5	4.2	36	3.6	6.6	8.9	1.21	2.11	Example		
28	N	95	1	2	2	1.00023	0	1230	1571	78	9.8	5.4	4.4	41	3.2	6.4	9.4	1.18	2.15	Example		
29	O	95	0	4	1	1.00015	2	1537	1821	84	9.8	5.3	4.5	34	3.9	6.2	7.8	1.17	1.75	Example		
30	P	96	1	2	1	1.00041	2	1217	1521	80	9.0	4.9	4.1	42	3.2	5.9	8.1	1.20	1.96	Example		
31	Q	95	0	3	2	1.00052	2	1195	1569	76	9.2	5.0	4.2	37	3.6	6.3	8.2	1.26	1.95	Example		
32	R	78	18	3	1	1.00031	8	940	1236	76	12.0	6.0	6.0	54	2.9	6.9	9.2	1.15	1.53	Example		
33	S	95	0	3	2	1.00045	2	1024	1329	77	9.5	5.2	4.3	34	2.9	6.2	9.3	1.19	2.17	Example		
34	T	100	0	0	0	1.00030	2	1311	1478	89	7.6	3.7	3.9	48	3.2	4.4	7.7	1.19	1.98	Example		
35	U	96	1	2	1	1.00062	0	1192	1513	79	9.8	5.4	4.4	35	3.6	6.3	7.8	1.17	1.76	Example		
36	V	95	1	2	2	1.00041	0	1258	1820	69	9.6	5.2	4.4	33	4.3	6.1	8.4	1.17	1.92	Example		
37	W	96	1	2	1	1.00021	0	1135	1311	87	9.6	5.4	4.2	47	2.9	6.6	8.0	1.22	1.89	Example		
38	X	96	0	2	2	1.00031	0	1214	1513	80	9.4	5.2	4.2	43	3.2	6.1	8.6	1.17	2.06	Example		
39	Y	95	1	2	2	1.00024	2	1520	1810	84	9.1	5.0	4.1	41	3.9	5.9	8.5	1.18	2.07	Example		
40	Z	97	0	2	1	1.00057	0	810	1198	68	11.9	6.9	5.0	36	3.2	8.2	9.0	1.19	1.80	Example		
41	AA	98	0	1	1	1.00019	0	1160	1501	77	7.1	3.5	3.6	34	3.6	6.4	9.0	1.84	2.50	Example		
42	AB	97	1	1	1	1.00064	0	1206	1504	63	8.1	4.5	3.6	36	3.2	6.1	7.7	1.36	2.13	Example		
43	AC	95	2	2	1	1.00116	2	1175	1566	67	9.3	5.0	4.3	37	3.6	6.1	8.0	1.22	1.87	Example		
44	AD	95	3	1	1	1.00042	0	1219	1516	62	9.3	5.2	4.1	45	3.2	6.2	8.7	1.20	2.12	Example		
45	AE	95	2	2	1	1.00052	0	1227	1504	61	9.6	5.3	4.3	42	3.2	6.2	8.8	1.17	2.06	Example		
46	AF	96	0	4	0	1.00062	0	1098	1504	73	9.2	4.9	4.3	39	3.6	6.3	8.7	1.29	2.03	Example		

Undefined: indicates value outside scope of present disclosure.

Underlined: indicates value outside scope of present disclosure.

[0166] As indicated in Table 3, the Examples each had a TS of 980 MPa or more, and excellent component strength, ductility, stretch flangeability, bendability, and warm workability. In contrast, among the Comparative Examples, one or more of component strength, ductility, stretch flangeability, bendability, or warm workability were inferior.

[0167] Although embodiments of the present disclosure have been described above, the present disclosure is not limited to the description of the embodiments that forms part of the present disclosure. That is, a person skilled in the art

may make various modifications to the embodiments, Examples, operational techniques, and the like disclosed herein, and all such modifications still fall within the scope of the present disclosure. For example, in the above-described series of heat treatments in the production method disclosed herein, any apparatus or the like may be used to perform the heat treatments on the steel sheet as long as the thermal hysteresis conditions are met.

Claims

1. A high strength steel sheet comprising a chemical composition containing, in mass%,

C: 0.030 % or more and 0.500 % or less,

Si: 0.01 % or more and 2.50 % or less,

Mn: 0.10 % or more and 5.00 % or less,

P: 0.100 % or less,

S: 0.0200 % or less,

Al: 1.000 % or less,

N: 0.0100 % or less, and

O: 0.0100 % or less,

with the balance being Fe and inevitable impurity, wherein

at a 1/4 sheet thickness position, the steel microstructure is such that an area fraction of martensite is 60 % or more,

an area fraction of ferrite is 40 % or less,

an area fraction of retained austenite is 20 % or less, and

a lattice constant a_M of the martensite, in nm, satisfies the following Expression 1,

$$\text{Expression 1: } 1.00005 \leq a_M/a_R \leq 1.00500$$

where, a_R is the lattice constant of the martensite, in nm, at room temperature after heat treatment of the high strength steel sheet at 500 °C for 30 min.

2. The high strength steel sheet according to claim 1, wherein the number of martensitic blocks containing carbide having a major axis length of 200 nm or more is 50 % or less of the number of martensitic blocks containing carbide.

3. The high strength steel sheet according to claim 1 or 2, the chemical composition further containing, in mass%, at least one element selected from the group consisting of:

Ti: 0.200 % or less,

Nb: 0.200 % or less,

V: 0.200 % or less,

Ta: 0.10 % or less,

W: 0.10 % or less,

B: 0.0100 % or less,

Cr: 1.00 % or less,

Mo: 1.00 % or less,

Ni: 1.00 % or less,

Co: 0.010 % or less,

Cu: 1.00 % or less,

Sn: 0.200 % or less,

Sb: 0.200 % or less,

Ca: 0.0100 % or less,

Mg: 0.0100 % or less,

REM: 0.0100 % or less,

Zr: 0.100 % or less,

Te: 0.100 % or less,

Hf: 0.10 % or less, and

Bi: 0.200 % or less.

4. The high strength steel sheet according to any one of claims 1 to 3, further comprising a coated or plated layer on at least one side.
5. A member formed using the high strength steel sheet according to any one of claims 1 to 4.
6. The member according to claim 5, wherein the member is for an automobile frame structural component or reinforcing component.

7. A method of producing a high strength steel sheet, the method comprising:

rough rolling a steel slab having the chemical composition according to claim 1 or 3 at an average strain rate of 1×10^{-4} /s or more and 1×10^{-1} /s or less with a total rolling reduction of 50 % or more to obtain a rough rolled sheet, then finish rolling the rough rolled sheet to obtain a hot-rolled sheet;
cold rolling the hot-rolled sheet to obtain a cold-rolled sheet;
annealing the cold-rolled sheet by heating to a heating temperature of 800 °C or more, and cooling under a condition that a residence time t_1 in a temperature range T_1 from Ms temperature to 700 °C is 1000 s or less; and partially tempering the cold-rolled sheet by cooling from the Ms temperature to 50 °C or less under a condition that a partial tempering parameter S, in μm^2 , satisfies the following Expression 2,

Expression 2:

$$0.01 \leq S \leq 30$$

Expression 3:

$$S = \sum_{t=1}^{t_E} f(t) D(T) \times 10^{12}$$

Expression 4:

$$f(t) = 1 - \exp(-1.1 \times 10^{-2}(M_s - T_{\min}))$$

Expression 5:

$$D(T) = 0.394 \times 10^{-6} \exp\left(-\frac{80220}{8.314 \times (T+273)}\right)$$

where, in Expressions 2 to 5,

a time when the temperature of the cold-rolled sheet first reaches the Ms temperature after the annealing is defined as $t = 0$, in s, and $t = t_E$, in s, is the time when the partial tempering is completed and the temperature of the cold-rolled sheet is 50 °C,

T , in °C, is an average temperature of the cold-rolled sheet at time $t-1$ to t , in s,

$T_{\min}$, in °C, is the lowest temperature of T at time 0 to t , in s, and

M_s , in °C, indicates the Ms temperature of the high strength steel sheet.

8. The method of producing a high strength steel sheet according to claim 7, where, in the partial tempering, cooling is stopped at a cooling stop temperature that is room temperature or more and the Ms temperature or less, followed by reheating to a reheating temperature and then cooling to 50 °C or less.
9. The method of producing a high strength steel sheet according to claim 7 or 8, where, in the annealing, coating or plating treatment is further applied to at least one side of the cold-rolled sheet.
10. A method of producing a member, the method comprising at least one of a forming process or a joining process applied to the high strength steel sheet according to any one of claims 1 to 4 to obtain the member.

FIG. 1

- : Electron diffraction spots of matrix phase martensite when electron beam is incident from $[100]$ orientation
- : Electron diffraction spots of carbide

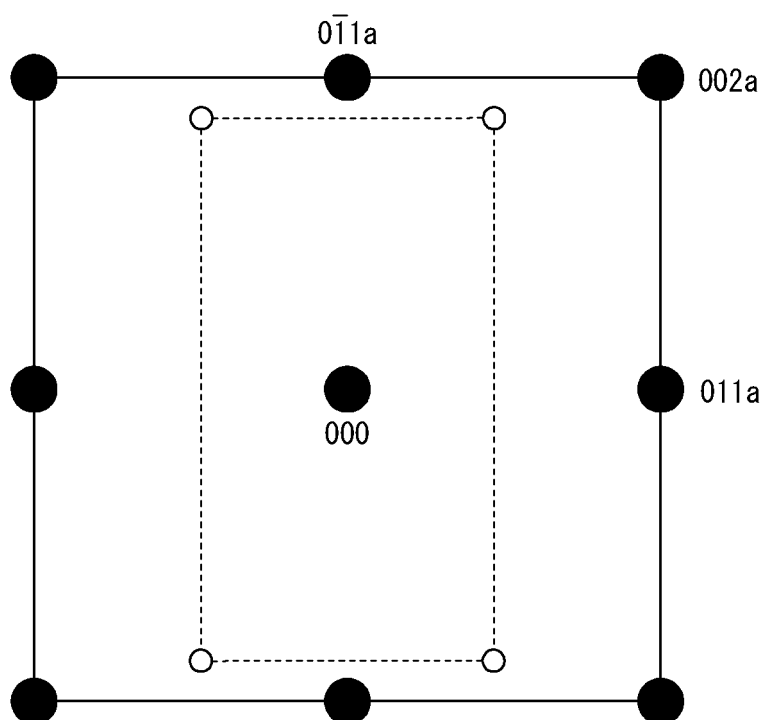


FIG. 2

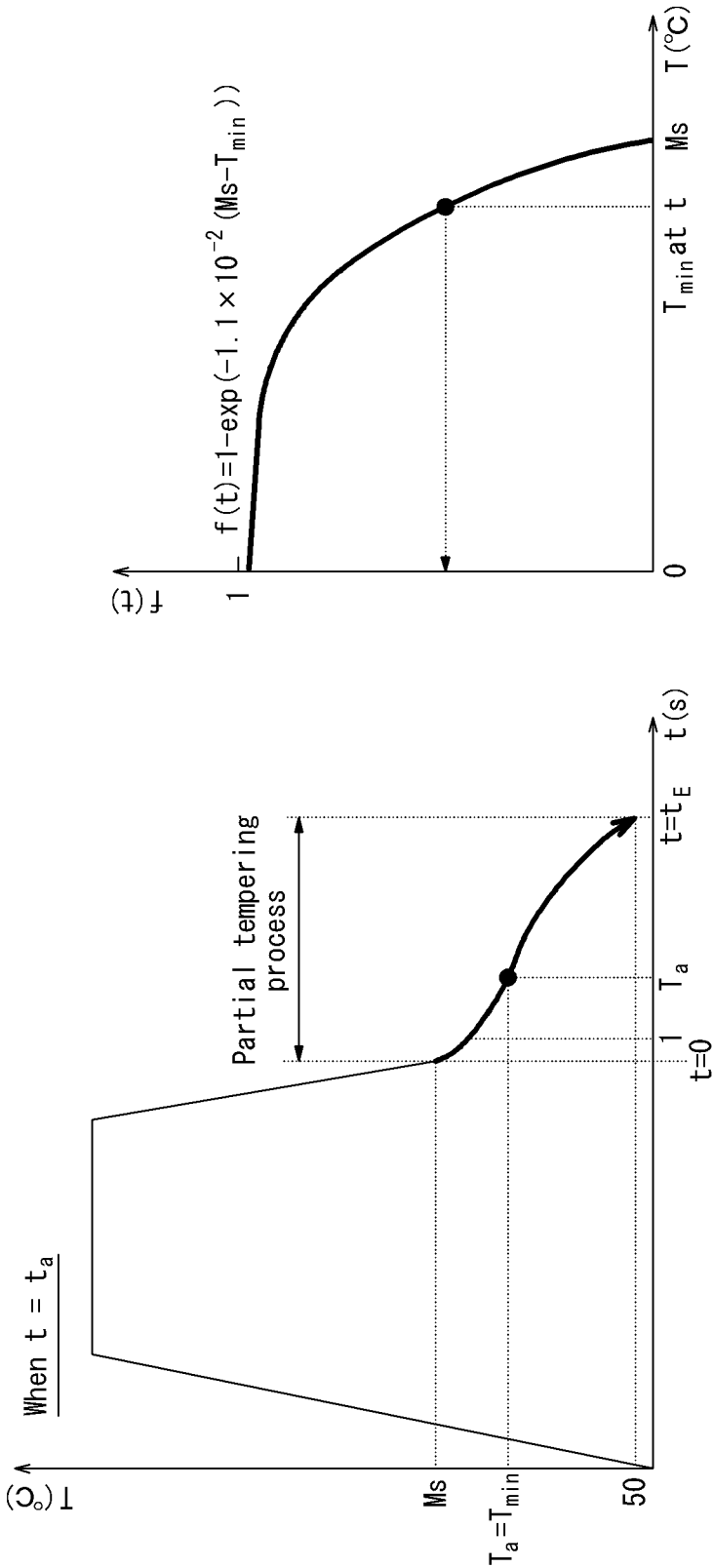
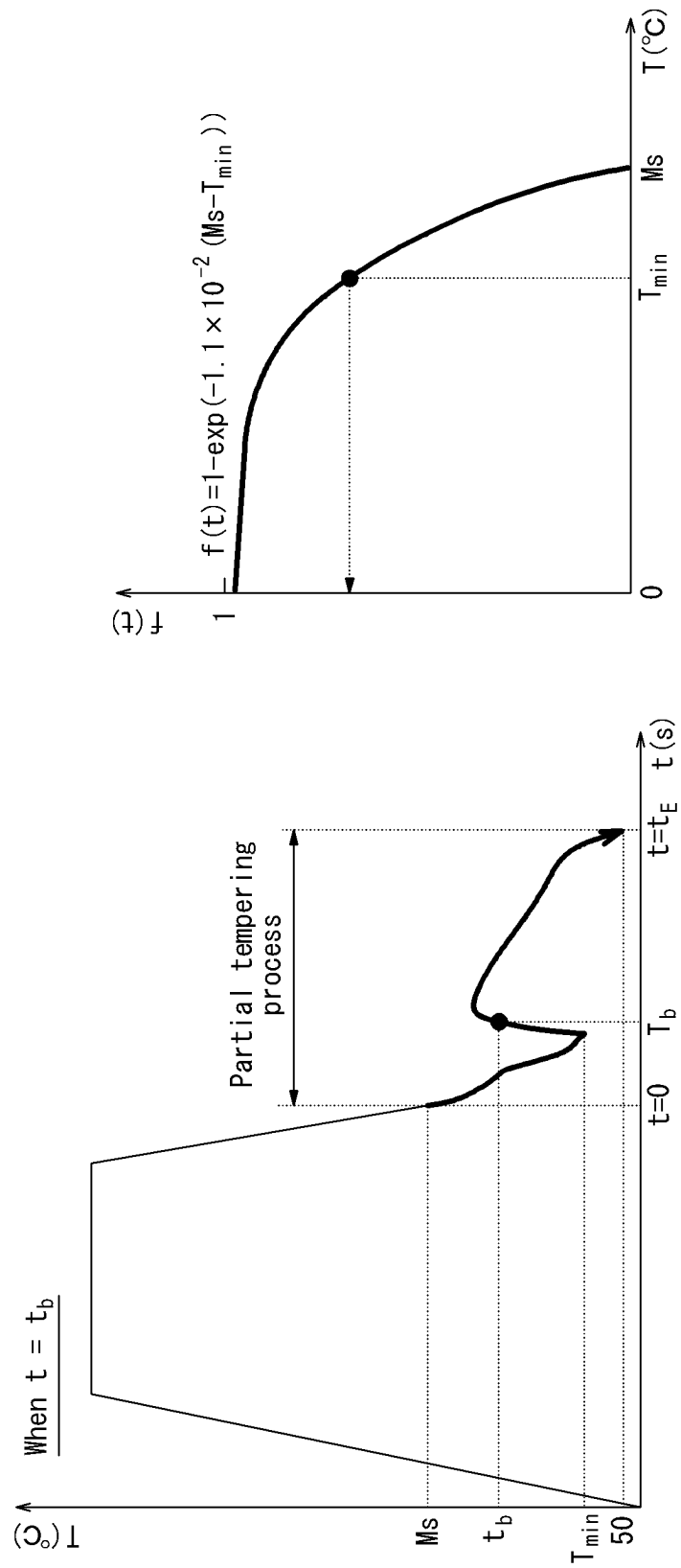


FIG. 3



INTERNATIONAL SEARCH REPORT

International application No.

PCT/JP2023/031053

A. CLASSIFICATION OF SUBJECT MATTER

C21D 9/46(2006.01)i; *C22C 18/00*(2006.01)i; *C22C 18/04*(2006.01)i; *C22C 38/00*(2006.01)i; *C22C 38/06*(2006.01)i;
C22C 38/60(2006.01)i

FI: C22C38/00 301S; C22C38/00 301T; C22C38/06; C22C38/60; C21D9/46 G; C21D9/46 J; C22C18/00; C22C18/04

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)

C21D9/46; C22C38/00-38/60

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched

Published examined utility model applications of Japan 1922-1996
 Published unexamined utility model applications of Japan 1971-2023
 Registered utility model specifications of Japan 1996-2023
 Published registered utility model applications of Japan 1994-2023

Electronic data base consulted during the international search (name of data base and, where practicable, search terms used)

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	WO 2021/193310 A1 (JFE STEEL CORP) 30 September 2021 (2021-09-30) entire text	1-10
A	WO 2017/150117 A1 (KOBE STEEL LTD) 08 September 2017 (2017-09-08) entire text	1-10
A	JP 2004-308002 A (KOBE STEEL LTD) 04 November 2004 (2004-11-04) entire text	1-10
A	JP 2012-188738 A (KOBE STEEL LTD) 04 October 2012 (2012-10-04) entire text	1-10
A	JP 2009-24226 A (NIPPON STEEL CORP) 05 February 2009 (2009-02-05) entire text	7-10

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

* Special categories of cited documents:	"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
"A" document defining the general state of the art which is not considered to be of particular relevance	"X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone
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"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)	"&" document member of the same patent family
"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	

Date of the actual completion of the international search

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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.

PCT/JP2023/031053

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WO 2017/150117 A1	08 September 2017	US 2019/0085426 A1 entire text EP 3412786 A1 CN 108699653 A KR 10-2018-0120712 A	
JP 2004-308002 A	04 November 2004	(Family: none)	
JP 2012-188738 A	04 October 2012	(Family: none)	
JP 2009-24226 A	05 February 2009	(Family: none)	

Form PCT/ISA/210 (patent family annex) (January 2015)

REFERENCES CITED IN THE DESCRIPTION

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- WO 2019181950 A1 [0004]
- WO 2017131053 A1 [0004]