

(19)



(11)

EP 4 474 509 A1

(12)

EUROPEAN PATENT APPLICATION

published in accordance with Art. 153(4) EPC

(43) Date of publication:

11.12.2024 Bulletin 2024/50(21) Application number: **23774239.0**(22) Date of filing: **30.01.2023**

(51) International Patent Classification (IPC):

C22C 38/00 ^(2006.01) **C21D 9/46** ^(2006.01)
C22C 38/06 ^(2006.01) **C22C 38/60** ^(2006.01)

(52) Cooperative Patent Classification (CPC):

C21D 9/46; C22C 38/00; C22C 38/06; C22C 38/60

(86) International application number:

PCT/JP2023/002913

(87) International publication number:

WO 2023/181640 (28.09.2023 Gazette 2023/39)

(84) Designated Contracting States:

**AL AT BE BG CH CY CZ DE DK EE ES FI FR GB
 GR HR HU IE IS IT LI LT LU LV MC ME MK MT NL
 NO PL PT RO RS SE SI SK SM TR**

Designated Extension States:

BA

Designated Validation States:

KH MA MD TN(30) Priority: **25.03.2022 JP 2022049756**

(71) Applicant: **JFE Steel Corporation**
Tokyo 100-0011 (JP)

(72) Inventors:

- **TOBATA, Junya**
Tokyo 100-0011 (JP)
- **TOJI, Yuki**
Tokyo 100-0011 (JP)
- **MINAMI, Hidekazu**
Tokyo 100-0011 (JP)

(74) Representative: **Hoffmann Eitle**

Patent- und Rechtsanwälte PartmbB
Arabellastraße 30
81925 München (DE)

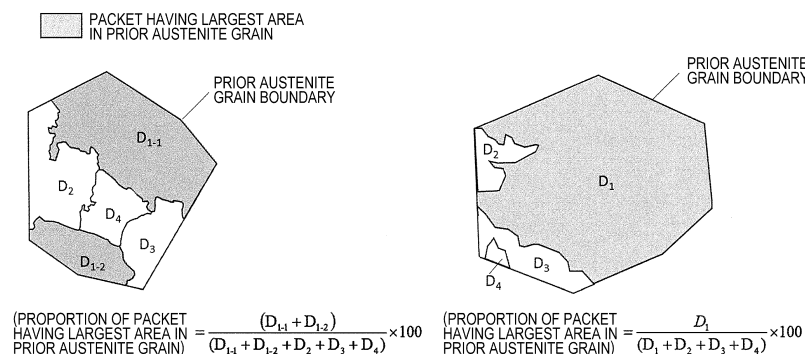
(54) HIGH STRENGTH STEEL SHEET AND MANUFACTURING METHOD THEREFOR

(57) Objects are to provide a high strength steel sheet having 1180 MPa or higher TS and 85% or more YR and being excellent in flatness in the width direction and working embrittlement resistance; and to provide a method for manufacturing the same.

The high strength steel sheet has a specific chemical composition and is such that in a region at 1/4 sheet thickness, the area fraction of tempered martensite is

90% or more, the volume fraction of retained austenite is less than 3%, the area fraction of the total of ferrite and bainitic ferrite is less than 10%, the average grain size of prior austenite is 20 μm or less, and the average of the proportions of packets having the largest area in prior austenite grains is 70% by area or less of the prior austenite grain.

FIG. 1



Description

Technical Field

[0001] The present invention relates to a high strength steel sheet excellent in tensile strength, flatness in the width direction, and working embrittlement resistance, and to a method for manufacturing the same. The high strength steel sheet of the present invention may be suitably used as structural members, such as automobile parts.

Background Art

[0002] Steel sheets for automobiles are being increased in strength in order to reduce CO₂ emissions by weight reduction of vehicles and to enhance crashworthiness by weight reduction of automobile bodies at the same time, with introduction of new laws and regulations one after another. To increase the strength of automobile bodies, high strength steel sheets having a tensile strength (TS) of 1180 MPa or higher grade are increasingly applied to principal structural parts of automobiles.

[0003] From the point of view of the performance of parts, high strength steel sheets used in automobiles require high working embrittlement resistance and excellent yield ratio. For example, high strength steel sheets applied to automobile frame parts, such as bumpers, are suitably those that excel in working embrittlement resistance and are not embrittled upon being press-formed, and have excellent collision impact absorption properties which are correlated with YR.

[0004] Furthermore, high strength steel sheets used in automobiles require high flatness. Patent Literature 1 describes that warpage of a steel sheet causes operational troubles in forming lines and adversely affects the dimensional accuracy of products. The present inventors carried out extensive studies and have found that the dimensional accuracy of products is affected not only by the warpage of steel sheets but also by the flatness in the width direction that is evaluated as steepness. For example, the steepness in the width direction is suitably 0.02 or less in order to achieve excellent dimensional accuracy.

[0005] To meet the above demands, for example, Patent Literature 2 provides a high strength steel sheet having a tensile strength of 1100 MPa or more and being excellent in YR, surface quality, and weldability, and a method for manufacturing the same. However, the technique described in Patent Literature 2 does not take into consideration flatness in the width direction and working embrittlement resistance.

[0006] Patent Literature 3 provides a hot-dip galvanized steel sheet with excellent press formability and low-temperature toughness that has a tensile strength of 980 MPa or more, and a method for manufacturing the same. While the steel sheet of Patent Literature 3 is improved in embrittlement at low temperatures, the technique does not take into consideration the working embrittlement of the steel sheet or the flatness in the width direction.

Citation List

Patent Literature

[0007]

PTL 1: Japanese Patent No. 4947176

PTL 2: Japanese Patent No. 6525114

PTL 3: Japanese Patent No. 6777272

Non Patent Literature

[0008] NPL 1: Journal of Smart Processing, 2013, Vol. 2, No. 3, pp. 110-118

Summary of Invention

Technical Problem

[0009] The present invention has been developed in view of the circumstances discussed above. Objects of the present invention are therefore to provide a high strength steel sheet having 1180 MPa or higher TS and 85% or more YR and being excellent in flatness in the width direction and working embrittlement resistance; and to provide a method for manufacturing the same.

Solution to Problem

[0010] The present inventors carried out extensive studies directed to solving the problems described above and have consequently found the following facts.

- (1) 1180 MPa or higher TS can be realized by limiting the amount of tempered martensite to 90% or more.
- (2) 85% or more YR can be achieved by limiting the amount of retained austenite to less than 3% and the amount of the total of ferrite and bainitic ferrite to less than 10%.
- (3) The flatness in the width direction can be enhanced by limiting the proportion of a packet having the largest area in tempered martensite to 70% or less of a prior austenite grain.
- (4) Excellent working embrittlement resistance can be achieved by limiting the proportion of a packet having the largest area in tempered martensite to 70% or less of a prior austenite grain and by limiting the average prior austenite grain size in tempered martensite to 20 μm or less.

[0011] The present invention has been made based on the above findings. Specifically, a summary of configurations of the present invention is as follows.

[1] A high strength steel sheet having a chemical composition including, 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, a balance being Fe and incidental impurities, the high strength steel sheet being such that in a region at 1/4 sheet thickness, an area fraction of tempered martensite is 90% or more, a volume fraction of retained austenite is less than 3%, an area fraction of the total of ferrite and bainitic ferrite is less than 10%, an average grain size of prior austenite is 20 μm or less, and an average of the proportions of packets having the largest area in prior austenite grains is 70% by area or less of the prior austenite grain.

[2] The high strength steel sheet according to [1], wherein the chemical composition further includes at least one element selected from, in mass%, 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, Co: 0.010% or less, Ni: 1.00% 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.

[3] The high strength steel sheet according to [1] or [2], which has a coated layer on a surface of the steel sheet.

[4] A method for manufacturing the high strength steel sheet described in [1] or [2], the method including providing a cold rolled steel sheet produced by subjecting a steel having the chemical composition described above to hot rolling, pickling, and cold rolling; heating the steel sheet at an annealing temperature T1 of 750°C or above and 950°C or below for a holding time t1 at the annealing temperature T1 of 10 seconds or more and 1000 seconds or less; cooling the steel sheet in such a manner that an average cooling rate from 750°C to 600°C is 20°C/s or more, an average cooling rate from (Ms + 50°C) to a quench start temperature T2 is 5°C/s or more and 30°C/s or less wherein the quench start temperature T2 is (Ms - 80°C) or above and is below Ms where Ms is martensite start temperature (°C) defined by formula (1), and an average cooling rate from the quench start temperature T2 to 80°C is 300°C/s or more; and heating the steel sheet at a tempering temperature T3 of 100°C or above and 400°C or below for a holding time t3 at the tempering temperature T3 of 10 seconds or more and 10000 seconds or less,

$$Ms = 519 - 474 \times [\%C] - 30.4 \times [\%Mn] - 12.1 \times [\%Cr] - 7.5 \times [\%Mo] - 17.7 \times [\%Ni] \quad (1)$$

wherein [% C], [% Mn], [% Cr], [% Mo], and [% Ni] indicate the contents (mass%) of C, Mn, Cr, Mo, and Ni, respectively, and are zero when the element is absent.

[5] The method for manufacturing the high strength steel sheet according to [4], further including performing a coating treatment.

Advantageous Effects of Invention

[0012] According to the present invention, a high strength steel sheet can be obtained that has 1180 MPa or higher TS and 85% or more YR and excels in flatness in the width direction and working embrittlement resistance. Furthermore, for example, the high strength steel sheet of the present invention may be applied to automobile structural members to reduce the weight of automobile bodies and thereby to enhance fuel efficiency. Thus, the present invention is highly valuable in industry.

Brief Description of Drawings

[0013]

5 [Fig. 1] Fig. 1 is a set of views illustrating a structure of a packet having the largest area in a prior austenite grain according to the present invention, and how the calculation is made.

[Fig. 2] Fig. 2 is a set of views illustrating the concept of the steepness θ of a steel sheet according to the present invention, and how the steepness is calculated. Description of Embodiments

10 [0014] Embodiments of the present invention will be described below.

[0015] First, appropriate ranges of the chemical composition of the high strength steel sheet and the reasons why the chemical composition is thus limited will be described. In the following description, "%" indicating the contents of constituent elements of steel means "mass%" unless otherwise specified.

15 [C: 0.030% or more and 0.500% or less]

[0016] Carbon is one of the important basic components of steel. Particularly in the present invention, carbon is an important element that affects the fraction of tempered martensite and the working embrittlement resistance. When the C content is less than 0.030%, the fraction of tempered martensite is so small that realizing 1180 MPa or higher TS is difficult. When, on the other hand, the C content is more than 0.500%, tempered martensite becomes brittle to cause deterioration in working embrittlement resistance. Thus, the C content is limited to 0.030% or more and 0.500% or less. The C content is preferably 0.050% or more. The C content is preferably 0.400% or less. The C content is more preferably 0.100% or more. The C content is more preferably 0.350% or less.

25 [Si: 0.01% or more and 2.50% or less]

[0017] Silicon is one of the important basic components of steel. Silicon suppresses the occurrence of carbides during continuous annealing and promotes the formation of retained austenite. Thus, particularly in the present invention, silicon is an important element that affects TS and the amount of retained austenite. When the Si content is less than 0.01%, realizing 1180 MPa or higher TS is difficult. When, on the other hand, the Si content is more than 2.50%, the amount of retained austenite is increased excessively to make it difficult to achieve 85% or more YR. Thus, the Si content is limited to 0.01% or more and 2.50% or less. The Si content is preferably 0.05% or more. The Si content is preferably 2.00% or less. The Si content is more preferably 0.10% or more. The Si content is more preferably 1.20% or less.

35 [Mn: 0.10% or more and 5.00% or less]

[0018] Manganese is one of the important basic components of steel. Particularly in the present invention, manganese is an important element that affects the fraction of tempered martensite and the working embrittlement resistance. When the Mn content is less than 0.10%, the fraction of tempered martensite is so small that realizing 1180 MPa or higher TS is difficult. When, on the other hand, the Mn content is more than 5.00%, tempered martensite becomes brittle to cause deterioration in working embrittlement resistance. Thus, the Mn content is limited to 0.10% or more and 5.00% or less. The Mn content is preferably 0.50% or more. The Mn content is preferably 4.50% or less. The Mn content is more preferably 0.80% or more. The Mn content is more preferably 4.00% or less.

45 [P: 0.100% or less]

[0019] Phosphorus is segregated at prior austenite grain boundaries and makes the grain boundaries brittle, thereby lowering the ultimate deformability of steel sheets and causing deterioration in working embrittlement resistance. Thus, the P content needs to be 0.100% or less. The lower limit of the P content is not particularly specified. In view of the fact that phosphorus is a solid solution strengthening element and can increase the strength of steel sheets, the lower limit is preferably 0.001% or more. For the reasons above, the P content is limited to 0.100% or less. The P content is preferably 0.001% or more. The P content is preferably 0.070% or less.

55 [S: 0.0200% or less]

[0020] Sulfur forms sulfides and lowers the ultimate deformability of steel sheets to cause deterioration in working embrittlement resistance. Thus, the S content needs to be 0.0200% or less. The lower limit of the S content is not particularly specified but is preferably 0.0001% or more due to production technique limitations. For the reasons above, the

S content is limited to 0.0200% or less. The S content is preferably 0.0001% or more. The S content is preferably 0.0050% or less.

[Al: 1.000% or less]

[0021] Aluminum raises the A_3 transformation temperature to allow more ferrite to be contained in the microstructure. The fraction of tempered martensite is correspondingly lowered to make it difficult to realize 1180 MPa or higher TS. Thus, the Al content needs to be 1.000% or less. The lower limit of the Al content is not particularly specified. In view of the fact that aluminum suppresses the occurrence of carbides during continuous annealing and promotes the formation of retained austenite, the Al content is preferably 0.001% or more. For the reasons above, the Al content is limited to 1.000% or less. The Al content is preferably 0.001% or more. The Al content is preferably 0.500% or less.

[N: 0.0100% or less]

[0022] Nitrogen forms nitrides and lowers the ultimate deformability of steel sheets to cause deterioration in working embrittlement resistance. Thus, the N content needs to be 0.0100% or less. The lower limit of the N content is not particularly specified but the N content is preferably 0.0001% or more due to production technique limitations. For the reasons above, the N content is limited to 0.0100% or less. The N content is preferably 0.0001% or more. The N content is preferably 0.0050% or less.

[O: 0.0100% or less]

[0023] Oxygen forms oxides and lowers the ultimate deformability of steel sheets to cause deterioration in working embrittlement resistance. Thus, the O content needs to be 0.0100% or less. The lower limit of the O content is not particularly specified but the O content is preferably 0.0001% or more due to production technique limitations. For the reasons above, the O content is limited to 0.0100% or less. The O content is preferably 0.0001% or more. The O content is preferably 0.0050% or less.

[0024] The chemical composition of the high strength steel sheet according to an embodiment of the present invention includes the components described above, and the balance is Fe and incidental impurities. Here, the incidental impurities include Zn, Pb, As, Ge, Sr, and Cs. A total of 0.100% or less of these impurities is acceptable.

[0025] In addition to the components in the proportions described above, the high strength steel sheet of the present invention may further include at least one element selected from, in mass%, 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. These elements may be contained singly or in combination.

[0026] When the contents of Ti, Nb, and V are each 0.200% or less, coarse precipitates and inclusions will not occur in large amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the contents of Ti, Nb, and V are each preferably 0.200% or less. The lower limits of the contents of Ti, Nb, and V are not particularly specified. These elements form fine carbides, nitrides, or carbonitrides during hot rolling or continuous annealing to increase the strength of steel sheets. In view of this fact, the contents of Ti, Nb, and V are each more preferably 0.001% or more. When titanium, niobium, and vanadium are added, the contents thereof are each limited to 0.200% or less for the reasons above. The contents are each more preferably 0.001% or more. The contents are each more preferably 0.100% or less.

[0027] When the contents of Ta and W are each 0.10% or less, coarse precipitates and inclusions will not occur in large amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the contents of Ta and W are each preferably 0.10% or less. The lower limits of the contents of Ta and W are not particularly specified. These elements form fine carbides, nitrides, or carbonitrides during hot rolling or continuous annealing to increase the strength of steel sheets. In view of this fact, the contents of Ta and W are each more preferably 0.01% or more. When tantalum and tungsten are added, the contents thereof are each limited to 0.10% or less for the reasons above. The contents are each more preferably 0.01% or more. The contents are each more preferably 0.08% or less.

[0028] When the B content is 0.0100% or less, inner cracks that lower the ultimate deformability of steel sheets will not form during casting or hot rolling and thus there will be no deterioration in working embrittlement resistance. Thus, the B content is preferably 0.0100% or less. The lower limit of the B content is not particularly specified. The B content is more preferably 0.0003% or more in view of the fact that this element is segregated at austenite grain boundaries during annealing and enhances hardenability. When boron is added, the content thereof is limited to 0.0100% or less for the reasons above. The content is more preferably 0.0003% or more. The content is more preferably 0.0080% or less.

[0029] When the contents of Cr, Mo, and Ni are each 1.00% or less, coarse precipitates and inclusions will not occur in increased amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the contents of Cr, Mo, and Ni are each preferably 1.00% or less. The lower limits of the contents of Cr, Mo, and Ni are not particularly specified. In view of the fact that these elements enhance hardenability, the contents of Cr, Mo, and Ni are each more preferably 0.01% or more. When chromium, molybdenum, and nickel are added, the contents thereof are each limited to 1.00% or less for the reasons above. The contents are each more preferably 0.01% or more. The contents are each more preferably 0.80% or less.

[0030] When the Co content is 0.010% or less, coarse precipitates and inclusions will not occur in increased amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the Co content is preferably 0.010% or less. The lower limit of the Co content is not particularly specified. In view of the fact that this element enhances hardenability, the Co content is more preferably 0.001% or more. When cobalt is added, the content thereof is limited to 0.010% or less for the reasons above. The content is more preferably 0.001% or more. The content is more preferably 0.008% or less.

[0031] When the Cu content is 1.00% or less, coarse precipitates and inclusions will not occur in increased amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the Cu content is preferably 1.00% or less. The lower limit of the Cu content is not particularly specified. In view of the fact that this element enhances hardenability, the Cu content is preferably 0.01% or more. When copper is added, the content thereof is limited to 1.00% or less for the reasons above. The content is more preferably 0.01% or more. The content is more preferably 0.80% or less.

[0032] When the Sn content is 0.200% or less, inner cracks that lower the ultimate deformability of steel sheets will not form during casting or hot rolling and thus there will be no deterioration in working embrittlement resistance. Thus, the Sn content is preferably 0.200% or less. The lower limit of the Sn content is not particularly specified. The Sn content is more preferably 0.001% or more in view of the fact that tin enhances hardenability (in general, is an element that enhances corrosion resistance). When tin is added, the content thereof is limited to 0.200% or less for the reasons above. The content is more preferably 0.001% or more. The content is more preferably 0.100% or less.

[0033] When the Sb content is 0.200% or less, coarse precipitates and inclusions will not occur in increased amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the Sb content is preferably 0.200% or less. The lower limit of the Sb content is not particularly specified. In view of the fact that this element enables control of the thickness of surface layer softening and the strength, the Sb content is more preferably 0.001% or more. When antimony is added, the content thereof is limited to 0.200% or less for the reasons above. The content is more preferably 0.001% or more. The content is more preferably 0.100% or less.

[0034] When the contents of Ca, Mg, and REM are each 0.0100% or less, coarse precipitates and inclusions will not occur in increased amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the contents of Ca, Mg, and REM are each preferably 0.0100% or less. The lower limits of the contents of Ca, Mg, and REM are not particularly specified. In view of the fact that these elements change the shapes of nitrides and sulfides into spheroidal and enhance the ultimate deformability of steel sheets, the contents of Ca, Mg, and REM are each more preferably 0.0005% or more. When calcium, magnesium, and rare earth metal(s) are added, the contents thereof are each limited to 0.0100% or less for the reasons above. The contents are each more preferably 0.0005% or more. The contents are each more preferably 0.0050% or less.

[0035] When the contents of Zr and Te are each 0.100% or less, coarse precipitates and inclusions will not occur in increased amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the contents of Zr and Te are each preferably 0.100% or less. The lower limits of the contents of Zr and Te are not particularly specified. In view of the fact that these elements change the shapes of nitrides and sulfides into spheroidal and enhance the ultimate deformability of steel sheets, the contents of Zr and Te are each more preferably 0.001% or more. When zirconium and tellurium are added, the contents thereof are each limited to 0.100% or less for the reasons above. The contents are each more preferably 0.001% or more. The contents are each more preferably 0.080% or less.

[0036] When the Hf content is 0.10% or less, coarse precipitates and inclusions will not occur in increased amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the Hf content is preferably 0.10% or less. The lower limit of the Hf content is not particularly specified. In view of the fact that this element changes the shapes of nitrides and sulfides into spheroidal and enhances the ultimate deformability of steel sheets, the Hf content is more preferably 0.01% or more. When hafnium is added, the content thereof is limited to 0.10% or less for the reasons above. The content is more preferably 0.01% or more. The content is more preferably 0.08% or less.

[0037] When the Bi content is 0.200% or less, coarse precipitates and inclusions will not occur in increased amounts and thus will not cause lowering of the ultimate deformability of steel sheets; hence there will be no deterioration in working embrittlement resistance. Thus, the Bi content is preferably 0.200% or less. The lower limit of the Bi content is not

particularly specified. In view of the fact that this element reduces the occurrence of segregation, the Bi content is more preferably 0.001% or more. When bismuth is added, the content thereof is limited to 0.200% or less for the reasons above. The content is more preferably 0.001% or more. The content is more preferably 0.100% or less.

[0038] When the content of any of Ti, Nb, V, Ta, W, B, Cr, Mo, Ni, Co, Cu, Sn, Sb, Ca, Mg, REM, Zr, Te, Hf, and Bi is below the preferred lower limit, the element does not impair the advantageous effects of the present invention and is regarded as an incidental impurity.

[0039] Next, the steel microstructure of the high strength steel sheet of the present invention will be described.

[Area fraction of tempered martensite: 90% or more]

[0040] This configuration is a very important requirement that constitutes the present invention. 1180 MPa or higher TS can be achieved when tempered martensite is the principal phase. In order to obtain this effect, the area fraction of tempered martensite needs to be 90% or more. Thus, the area fraction of tempered martensite is limited to 90% or more. The area fraction is preferably 94% or more, and more preferably 96% or more.

[0041] Here, tempered martensite is measured as follows. A longitudinal cross section of the steel sheet is polished and is etched with 3 vol% Nital. A portion at 1/4 sheet thickness (a location corresponding to 1/4 of the sheet thickness in the depth direction from the steel sheet surface) is observed using SEM in 10 fields of view at a magnification of $\times 2000$. In the microstructure images, tempered martensite is structures that have fine irregularities inside the structures and contain inner carbides. The values thus obtained are averaged to determine the tempered martensite.

[Amount of retained austenite: less than 3%]

[0042] This configuration is a very important requirement that constitutes the present invention. When the volume fraction of retained austenite is 3% or more, it is difficult to realize 85% or more YR. The reason for low YR is that retained austenite with a high fraction gives rise to a lowering in YS by undergoing strain-induced transformation. Thus, the retained austenite is limited to less than 3%. The amount of retained austenite is preferably 1% or less. The lower limit of retained austenite is not particularly limited and may be 0%.

[0043] Here, retained austenite is measured as follows. The steel sheet is polished to expose a face 0.1 mm below 1/4 sheet thickness and is thereafter further chemically polished to expose a face 0.1 mm below the face exposed above. The face is analyzed with an X-ray diffractometer using $\text{CoK}\alpha$ radiation to determine the integral intensity ratios of the diffraction peaks of {200}, {220}, and {311} planes of fcc iron and {200}, {211}, and {220} planes of bcc iron. Nine integral intensity ratios thus obtained are averaged to determine retained austenite.

[Area fraction of the total of ferrite and bainitic ferrite: less than 10%]

[0044] This configuration is a very important requirement that constitutes the present invention. When the total of ferrite and bainitic ferrite is 10% or more, it is difficult to realize 1180 MPa or higher TS and it is also difficult to achieve 85% or more YR. The reason for low YR is that ferrite and bainitic ferrite are soft microstructures and hasten the occurrence of yielding. Thus, the total of ferrite and bainitic ferrite is limited to less than 10%. The total amount is preferably 8% or less, and more preferably 5% or less. The lower limit of the total of ferrite and bainitic ferrite is not particularly limited and may be 0%.

[0045] Here, the total of ferrite and bainitic ferrite is measured as follows. A longitudinal cross section of the steel sheet is polished and is etched with 3 vol% Nital. A portion at 1/4 sheet thickness (a location corresponding to 1/4 of the sheet thickness in the depth direction from the steel sheet surface) is observed using SEM in 10 fields of view at a magnification of $\times 2000$. In the microstructure images, ferrite and bainitic ferrite are recessed structures having a flat interior and containing no inner carbides. The values thus obtained are averaged to determine the total of ferrite and bainitic ferrite.

[0046] Possible microstructures other than those described above include pearlite, fresh martensite, and acicular ferrite. These microstructures do not affect characteristics as long as their fractions are 5% or less, and thus may be present within that range.

[Average grain size of prior austenite: 20 μm or less]

[0047] This configuration is a very important requirement that constitutes the present invention. Reducing the average grain size of prior austenite can suppress crack propagation and thereby enhances the working embrittlement resistance of steel sheets. In order to obtain these effects, the average grain size of prior austenite needs to be 20 μm or less. The lower limit of the average grain size of prior austenite is not particularly specified. When, however, the average grain size of prior austenite is less than 2 μm , more retained austenite may form. Thus, the average grain size is preferably 2 μm or more. For the reasons above, the average grain size of prior austenite is limited to 20 μm or less. The average grain size is preferably 2 μm or more. The average grain size is preferably 15 μm or less. The average grain size is more preferably 3 μm

or more. The average grain size is more preferably 10 μm or less.

[0048] Here, the average grain size of prior austenite is measured as follows. A longitudinal cross section of the steel sheet is polished and is etched with, for example, a mixed solution of picric acid and ferric chloride to expose prior austenite grain boundaries. Portions at 1/4 sheet thickness (locations corresponding to 1/4 of the sheet thickness in the depth direction from the steel sheet surface) are photographed with an optical microscope each in 3 to 10 fields of view at a magnification of $\times 400$. Twenty straight lines including 10 vertical lines and 10 horizontal lines are drawn at regular intervals on the image data obtained, and the grain size is determined by a linear intercept method.

[Average of the proportions of packets having the largest area in prior austenite grains: 70% by area or less]

[0049] This configuration is a very important requirement that constitutes the present invention. The proportion of a packet having the largest area in a prior austenite grain affects the flatness in the width direction and the working embrittlement resistance. As illustrated in Fig. 1, a prior austenite grain contains up to four kinds of packets distinguished by crystal habit plane formed by transformation. The packet having the largest area in a prior austenite grain is the packet that occupies the largest area among such packets. The proportion of one packet in a prior austenite grain is determined by dividing the area of the packet of interest by the area of the whole prior austenite grain. As a result of extensive studies, the present inventors have found that strain among the packets is reduced and the flatness in the width direction is improved by lowering the proportion of a packet having the largest area in a prior austenite grain. The present inventors have also found that lowering the proportion of a packet having the largest area in a prior austenite grain leads to a fine microstructure and suppresses crack propagation, thereby enhancing the working embrittlement resistance of the steel sheet. Thus, the average of the proportions of packets having the largest area in prior austenite grains is limited to 70% or less. The average proportion is preferably 60% or less. The lower limit of the average proportion of packets having the largest area in prior austenite grains is not particularly limited. The grains contain up to four kinds of packets. When four packets are evenly distributed, the proportion of a packet having the largest area in the prior austenite grain is 25%. Thus, the lower limit of the average proportion of packets having the largest area in prior austenite grains may be 25% or more but is not necessarily limited thereto.

[0050] Here, the average proportion of packets having the largest area in prior austenite grains is measured as follows. First, a test specimen for microstructure observation is sampled from the cold rolled steel sheet. Next, the sampled test specimen is polished by vibration polishing with colloidal silica to expose a cross section in the rolling direction (a longitudinal cross section) for use as observation surface. The observation surface is specular. Next, electron backscatter diffraction (EBSD) measurement is performed with respect to a portion at 1/4 sheet thickness (a location corresponding to 1/4 of the sheet thickness in the depth direction from the steel sheet surface) to obtain local crystal orientation data. Here, the SEM magnification is $\times 1000$, the step size is 0.2 μm , the measured region is 80 μm square, and the WD is 15 mm. The local orientation data obtained is analyzed with OIM Analysis 7 (OIM), and a map (a CP map) that shows close-packed plane groups (CP groups) with different colors is created using the method described in Non Patent Literature 1. In the present invention, a packet is defined as a region or regions belonging to the same CP group. From the CP map obtained, the area of the packet having the largest area is determined and is divided by the area of the whole prior austenite grain to give the proportion of the packet having the largest area in the prior austenite grain. This analysis is performed with respect to 10 or more adjacent prior austenite grains, and the results are averaged to give the average proportion of packets having the largest area in prior austenite grains.

[0051] Next, a manufacturing method of the present invention will be described.

[0052] In the present invention, a steel material (a steel slab) may be obtained by any known steelmaking method without limitation, such as a converter or an electric arc furnace. To prevent macro-segregation, the steel slab (the slab) is preferably produced by a continuous casting method.

[0053] In the present invention, the slab heating temperature, the slab soaking holding time, and the coiling temperature in hot rolling are not particularly limited. For example, the steel slab may be hot rolled in such a manner that the slab is heated and is then rolled, that the slab is subjected to hot direct rolling after continuous casting without being heated, or that the slab is subjected to a short heat treatment after continuous casting and is then rolled. The slab heating temperature, the slab soaking holding time, the finish rolling temperature, and the coiling temperature in hot rolling are not particularly limited. The lower limit of the slab heating temperature is preferably 1100°C or above. The upper limit of the slab heating temperature is preferably 1300°C or below. The lower limit of the slab soaking holding time is preferably 30 minutes or more. The upper limit of the slab soaking holding time is preferably 250 minutes or less. The lower limit of the finish rolling temperature is preferably A_{r3} transformation temperature or above. Furthermore, the lower limit of the coiling temperature is preferably 350°C or above. The upper limit of the coiling temperature is preferably 650°C or below.

[0054] The hot rolled steel sheet thus produced is pickled. Pickling can remove oxides on the steel sheet surface and is thus important to ensure good chemical convertibility and a high quality of coating in the final high strength steel sheet. Pickling may be performed at a time or several. The hot rolled sheet that has been pickled may be cold rolled directly or may be subjected to heat treatment before cold rolling.

[0055] The rolling reduction in cold rolling and the sheet thickness after rolling are not particularly limited. The lower limit of the rolling reduction is preferably 30% or more. The upper limit of the rolling reduction is preferably 80% or less. The advantageous effects of the present invention may be obtained without any limitations on the number of rolling passes and the rolling reduction in each pass.

[0056] The cold rolled steel sheet obtained as described above is annealed. Annealing conditions are as follows.

[Annealing temperature T1: 750°C or above and 950°C or below]

[0057] When the annealing temperature T1 is below 750°C, the area fraction of the total of ferrite and bainitic ferrite is 10% or more to make it difficult to realize 1180 MPa or higher TS and 85% or more YR. When, on the other hand, the annealing temperature T1 is above 950°C, prior austenite grains are excessively increased in size and the prior austenite grain size exceeds 20 μ m to give rise to a decrease in working embrittlement resistance. Thus, the annealing temperature T1 is limited to 750°C or above and 950°C or below. The annealing temperature T1 is preferably 800°C or above. The annealing temperature T1 is preferably 900°C or below.

[Holding time t1 at the annealing temperature T1: 10 seconds or more and 1000 seconds or less]

[0058] When the holding time t1 at the annealing temperature T1 is less than 10 seconds, austenitization is insufficient and the area fraction of the total of ferrite and bainitic ferrite is 10% or more. As a result, it is difficult to achieve 1180 MPa or higher TS and it is also difficult to realize 85% or more YR. When, on the other hand, the holding time at the annealing temperature T1 is more than 1000 seconds, the prior austenite grain size is excessively increased, and the working embrittlement resistance is lowered. For the reasons above, the holding time t1 at the annealing temperature T1 is limited to 10 seconds or more and 1000 seconds or less. The holding time t1 is preferably 50 seconds or more. The holding time t1 is preferably 500 seconds or less.

[Average cooling rate from 750°C to 600°C: 20°C/s or more]

[0059] When the average cooling rate from 750°C to 600°C is less than 20°C/s, the area fraction of the total of ferrite and bainitic ferrite is 10% or more to make it difficult to achieve 1180 MPa or higher TS and to realize 85% or more YR. For the reasons above, the average cooling rate from 750°C to 600°C is limited to 20°C/s or more. The average cooling rate is preferably 30°C/s or more. The upper limit is not necessarily specified but is preferably 2000°C/s or less.

[Average cooling rate from (Ms + 50°C) to a quench start temperature T2: 5°C/s or more and 30°C/s or less]

[0060] This configuration is a very important requirement that constitutes the present invention. The average cooling rate from (Ms + 50°C) to a quench start temperature T2 affects the area fraction of the total of ferrite and bainitic ferrite and the average proportion of packets having the largest area in prior austenite grains. When the average cooling rate from (Ms + 50°C) to a quench start temperature T2 is less than 5°C/s, the area fraction of the total of ferrite and bainitic ferrite is 10% or more to make it difficult to achieve 1180 MPa or higher TS and to realize 85% or more YR. When, on the other hand, the average cooling rate from (Ms + 50°C) to a quench start temperature T2 is more than 30°C/s, the average proportion of packets having the largest area in prior austenite grains exceeds 70% to cause deterioration in flatness in the width direction and working embrittlement resistance. For the reasons above, the average cooling rate from (Ms + 50°C) to a quench start temperature T2 is limited to 5°C/s or more and 30°C/s or less. The average cooling rate is preferably 10°C/s or more. The average cooling rate is preferably 20°C/s or less.

[Quench start temperature T2: (Ms - 80°C) or above and below Ms]

[0061] This configuration is a very important requirement that constitutes the present invention. The quench start temperature T2 is controlled to (Ms - 80°C) or above and below Ms to ensure that the martensite transformation rate before the start of quenching is 1% or more and 80% or less. In this manner, quenching can give microstructures in which the average proportion of packets having the largest area in prior austenite grains is 70% or less and the volume fraction of retained austenite is less than 3%. When the quench start temperature T2 is below (Ms - 80°C), the martensite transformation rate before the start of quenching exceeds 80% and consequently the volume fraction of retained austenite is 3% or more to make it difficult to achieve 85% or more YR. When, on the other hand, the quench start temperature T2 is above Ms, the martensite transformation rate before the start of quenching is less than 1% and the average proportion of packets having the largest area in prior austenite grains exceeds 70% to cause deterioration in flatness in the width direction and working embrittlement resistance. Thus, the quench start temperature T2 is limited to (Ms - 80°C) or above and below Ms. The quench start temperature T2 is preferably (Ms - 50°C) or above.

[0062] The quench start temperature T2 is preferably (Ms - 5°C) or below. The martensite start temperature Ms (°C) is defined by the following formula (1):

$$Ms = 519 - 474 \times [\%C] - 30.4 \times [\%Mn] - 12.1 \times [\%Cr] - 7.5 \times [\%Mo] - 17.7 \times [\%Ni] \quad (1)$$

wherein [% C], [% Mn], [% Cr], [% Mo], and [% Ni] indicate the contents (mass%) of C, Mn, Cr, Mo, and Ni, respectively, and are zero when the element is absent.

[Average cooling rate from the quench start temperature T2 to 80°C: 300°C/s or more]

[0063] When the average cooling rate from the quench start temperature T2 to 80°C is less than 300°C/s, the volume fraction of retained austenite is 3% or more to make it difficult to achieve 85% or more YR. Thus, the average cooling rate from the quench start temperature T2 to 80°C is limited to 300°C/s or more. The average cooling rate is preferably 800°C/s or more. The upper limit is not necessarily specified but is preferably 2000°C/s or less.

[Tempering temperature T3: 100°C or above and 400°C or below]

[0064] In the present invention, tempered martensite is a microstructure that is formed when martensite at 80°C or below is heat-treated at a tempering temperature of 100°C or above for a holding time of 10 seconds or more. Thus, martensite is not sufficiently tempered when the tempering temperature T3 is below 100°C. The resultant microstructures will be based on as-quenched martensite, which deteriorates the working embrittlement resistance. When, on the other hand, the tempering temperature T3 is above 400°C, tempered martensite is decomposed into ferrite and the area fraction of tempered martensite is less than 90% to make it difficult to achieve 1180 MPa or higher TS. For the reasons above, the tempering temperature T3 is limited to 100°C or above and 400°C or below. The tempering temperature T3 is preferably 150°C or above. The tempering temperature T3 is preferably 350°C or below.

[Holding time t3 at the tempering temperature T3: 10 seconds or more and 10000 seconds or less]

[0065] In the present invention, tempered martensite is a microstructure that is formed when martensite at 80°C or below is heat-treated at a tempering temperature of 100°C or above for a holding time of 10 seconds or more. Thus, martensite is not sufficiently tempered when the holding time t3 at the tempering temperature T3 is less than 10 seconds. The resultant microstructures will be based on as-quenched martensite, which deteriorates the working embrittlement resistance. When, on the other hand, the tempering temperature T3 is more than 10000 seconds, tempered martensite is decomposed into ferrite and the area fraction of tempered martensite is less than 90% to make it difficult to achieve 1180 MPa or higher TS. For the reasons above, the holding time t3 at the tempering temperature T3 is limited to 10 seconds or more and 10000 seconds or less. The holding time t3 is preferably 50 seconds or more. The holding time t3 is preferably 5000 seconds or less.

[0066] Post-temper cooling is not particularly limited and the steel sheet may be cooled to a desired temperature in an appropriate manner. Incidentally, the desired temperature is preferably about room temperature.

[0067] Furthermore, the high strength steel sheet described above may be worked under conditions where the amount of equivalent plastic strain is 0.10% or more and 5.00% or less. The working may be followed by reheating at 100°C or above and 400°C or below.

[0068] When the high strength steel sheet is a product that is traded, the steel sheet is usually traded after being cooled to room temperature.

[0069] The high strength steel sheet may be subjected to coating treatment during annealing or after annealing.

[0070] For example, the coating treatment during annealing may be hot-dip galvanizing treatment performed when the steel sheet is being cooled or has been cooled from 750°C to 600°C at an average cooling rate of 20°C/s or more. The hot-dip galvanizing treatment may be followed by alloying. For example, the coating treatment after annealing may be Zn-Ni electrical alloy coating treatment or pure Zn electroplated coating treatment performed after tempering. A coated layer may be formed by electroplated coating, or hot-dip zinc-aluminum-magnesium alloy coating may be applied. While the coating treatment has been described above focusing on zinc coating, the types of coating metals, such as Zn coating and Al coating, are not particularly limited. Other conditions in the manufacturing method are not particularly limited. From the point of view of productivity, the series of treatments including annealing, hot-dip galvanizing, and alloying treatment of the coated zinc layer is preferably performed on hot-dip galvanizing line CGL (continuous galvanizing line). To control the coating weight of the coated layer, the hot-dip galvanizing treatment may be followed by wiping. Conditions for operations, such as coating, other than those conditions described above may be determined in accordance with the usual hot-dip galvanizing technique.

[0071] After the coating treatment after annealing, the steel sheet may be worked again under conditions where the

EP 4 474 509 A1

amount of equivalent plastic strain is 0.10% or more and 5.00 or less. The working may be followed by reheating at 100°C or above and 400°C or below.

EXAMPLES

5

[0072] Steels having a chemical composition described in Table 1 and 2, with the balance being Fe and incidental impurities, were smelted in a converter and were continuously cast into slabs. Next, the slabs obtained were heated, hot rolled, pickled, cold rolled, and subjected to annealing treatment and tempering treatment described in Tables 3 to 6. High strength cold rolled steel sheets having a sheet thickness of 0.6 to 2.2 mm were thus obtained. Incidentally, some of the steel sheets were subjected to coating treatment during or after annealing.

10

15

20

25

30

35

40

45

50

55

[Table 1]

Steels	Chemical composition (mass%)													
	C	Si	Mn	P	S	N	O	Al	Ti	B	Nb	Cu	Others	
A	0.205	0.193	2.69	0.005	0.0006	0.0067	0.007	0.053						INV. EX.
B	0.211	0.246	2.40	0.011	0.0012	0.0032	0.004	0.060						INV. EX.
C	0.200	0.331	2.57	0.007	0.0014	0.0049	0.006	0.047						INV. EX.
D	0.297	0.185	2.43	0.012	0.0008	0.0029	0.007	0.058						INV. EX.
E	0.292	0.173	2.39	0.010	0.0009	0.0043	0.005	0.016						INV. EX.
F	0.049	0.341	2.47	0.008	0.0010	0.0052	0.001	0.039						INV. EX.
G	<u>0.021</u>	0.165	2.61	0.012	0.0005	0.0068	0.004	0.057						COMP. EX.
H	0.460	0.325	2.58	0.010	0.0007	0.0028	0.003	0.043						INV. EX.
I	<u>0.522</u>	0.338	2.61	0.014	0.0013	0.0045	0.002	0.057						COMP. EX.
J	0.214	0.079	2.42	0.012	0.0010	0.0050	0.005	0.038						INV. EX.
K	0.216	<u>0.003</u>	2.57	0.015	0.0011	0.0052	0.003	0.052						COMP. EX.
L	0.187	2.403	2.59	0.012	0.0008	0.0061	0.006	0.021						INV. EX.
M	0.216	<u>2.532</u>	2.33	0.009	0.0013	0.0014	0.004	0.053						COMP. EX.
N	0.189	<u>0.178</u>	0.88	0.013	0.0008	0.0065	0.001	0.041						INV. EX.
O	0.209	0.211	<u>0.08</u>	0.008	0.0015	0.0020	0.001	0.041						COMP. EX.
P	0.195	0.164	4.76	0.007	0.0014	0.0066	0.005	0.039						INV. EX.
Q	0.196	0.324	<u>5.12</u>	0.013	0.0009	0.0060	0.003	0.016						COMP. EX.
R	0.208	0.326	2.68	0.099	0.0012	0.0052	0.003	0.049						INV. EX.
S	0.187	0.279	2.60	<u>0.121</u>	0.0012	0.0039	0.004	0.052						COMP. EX.
T	0.195	0.288	2.49	0.010	0.0182	0.0065	0.007	0.031						INV. EX.
U	0.180	0.326	2.51	0.005	<u>0.0222</u>	0.0025	0.005	0.041						COMP. EX.
V	0.200	0.181	2.58	0.005	0.0005	0.0062	0.005	0.976						INV. EX.
W	0.182	0.226	2.32	0.010	0.0010	0.0024	0.003	<u>1.135</u>						COMP. EX.
X	0.183	0.281	2.34	0.006	0.0006	0.0089	0.002	<u>0.015</u>						INV. EX.
Y	0.191	0.212	2.56	0.008	0.0010	<u>0.0112</u>	0.004	0.035						COMP. EX.

(continued)

Steels	Chemical composition (mass%)													
	C	Si	Mn	P	S	N	O	Al	Ti	B	Nb	Cu	Others	
Z	0.190	0.293	2.64	0.010	0.0012	0.0037	0.0090	0.026						INV. EX.
AA	0.209	0.251	2.69	0.007	0.0012	0.0057	<u>0.0110</u>	0.024						COMP. EX.
AB	0.188	0.284	2.31	0.012	0.0011	0.0011	0.006	0.013	0.002					INV. EX.
AC	0.188	0.317	2.48	0.006	0.0007	0.0032	0.001	0.031	0.198					INV. EX.
AD	0.212	0.199	2.39	0.014	0.0010	0.0032	0.004	0.055	<u>0.213</u>					COMP. EX.
AE	0.186	0.263	2.39	0.005	0.0008	0.0039	0.002	0.038		0.0003				INV. EX.
AF	0.197	0.198	2.58	0.009	0.0012	0.0043	0.002	0.059		0.0093				INV. EX.
AG	0.216	0.324	2.66	0.013	0.0015	0.0016	0.003	0.029		<u>0.0123</u>				COMP. EX.
AH	0.218	0.263	2.45	0.006	0.0014	0.0057	0.003	0.011			0.001			INV. EX.
AI	0.188	0.277	2.46	0.006	0.0007	0.0033	0.002	0.047			0.195			INV. EX.
Underlines indicate being outside the range of the present invention.														

[Table 2]

Steels	Chemical composition (mass%)												Others	
	C	Si	Mn	P	S	N	O	Al	Ti	B	Nb	Cu		
AJ	0.202	0.317	2.43	0.011	0.0013	0.0040	0.003	0.034			0.204			COMP. EX.
AK	0.187	0.236	2.64	0.007	0.0009	0.0026	0.007	0.016				0.03		INV. EX.
AL	0.202	0.214	2.40	0.012	0.0012	0.0043	0.002	0.039				1.00		INV. EX.
AM	0.219	0.208	2.66	0.005	0.0013	0.0049	0.006	0.024				1.12		COMP. EX.
AN	0.187	0.191	2.33	0.007	0.0009	0.0037	0.004	0.015					V:0.146	INV. EX.
AO	0.202	0.218	2.43	0.005	0.0012	0.0011	0.004	0.039					Ta0.02	INV. EX.
AP	0.186	0.187	2.49	0.011	0.0006	0.0044	0.006	0.059					W:0.02	INV. EX.
AQ	0.211	0.296	2.67	0.013	0.0009	0.0053	0.002	0.049					Cr:0.82	INV. EX.
AR	0.216	0.267	2.48	0.014	0.0013	0.0024	0.006	0.048					Mo:0.74	INV. EX.
AS	0.208	0.223	2.30	0.013	0.0006	0.0010	0.005	0.041					Co:0.008	INV. EX.
AT	0.196	0.163	2.66	0.012	0.0013	0.0046	0.001	0.050					Ni:0.79	INV. EX.
AU	0.190	0.184	2.31	0.015	0.0012	0.0049	0.002	0.058					Sn:0.155	INV. EX.
AV	0.202	0.163	2.65	0.008	0.0007	0.0011	0.004	0.012					Sb:0.025	INV. EX.
AW	0.220	0.262	2.42	0.011	0.0007	0.0063	0.006	0.054					Ca:0.0075	INV. EX.
AX	0.196	0.245	2.34	0.005	0.0011	0.0050	0.002	0.052					Mg:0.0025	INV. EX.
AY	0.203	0.256	2.64	0.005	0.0014	0.0014	0.005	0.040					Zr:0.034	INV. EX.
AZ	0.212	0.202	2.35	0.014	0.0010	0.0042	0.003	0.057					Te:0.065	INV. EX.
BA	0.310	0.300	2.63	0.012	0.0014	0.0033	0.003	0.040					Hf:0.05	INV. EX.
BB	0.290	0.178	2.54	0.005	0.0007	0.0050	0.003	0.031					REM:0.0008	INV. EX.
BC	0.318	0.159	2.48	0.007	0.0005	0.0010	0.005	0.032					Bi:0.010	INV. EX.
BD	0.318	0.249	2.44	0.012	0.0014	0.0024	0.001	0.026					Zn:0.052	INV. EX.
BE	0.281	0.168	2.30	0.012	0.0008	0.0019	0.005	0.025					Pb:0.048	INV. EX.
BF	0.312	0.239	2.34	0.009	0.0010	0.0060	0.006	0.014					As:0.050	INV. EX.
BG	0.306	0.280	2.53	0.014	0.0005	0.0064	0.002	0.032					Ge:0.019	INV. EX.
BH	0.287	0.286	2.68	0.010	0.0006	0.0032	0.004	0.036					Sr:0.055	INV. EX.

5
10
15
20
25
30
35
40
45
50
55

(continued)

Steels	Chemical composition (mass%)													
	C	Si	Mn	P	S	N	O	Al	Ti	B	Nb	Cu	Others	
BI	0.296	0.183	2.51	0.013	0.0006	0.0023	0.007	0.017					Cs:0.065	INV. EX.
BJ	0.198	0.870	2.70	0.010	0.0003	0.0040	0.001	0.045	0.007	0.0017	0.014	0.18	Ni:0.05	INV. EX.
BK	0.180	0.286	2.45	0.014	0.0014	0.0063	0.006	0.011						INV. EX.
BL	0.288	0.266	2.61	0.008	0.0013	0.0062	0.003	0.029						INV. EX.
BM	0.310	0.205	2.58	0.012	0.0015	0.0039	0.002	0.049						INV. EX.
BN	0.107	0.367	2.18	0.008	0.0011	0.0065	0.004	0.020						INV. EX.
BO	0.102	0.306	1.92	0.009	0.0012	0.0025	0.005	0.016						INV. EX.
Underlines indicate being outside the range of the present invention.														

[Table 3]

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50°C)-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type"	
1	A	866	208	76	12	340	260	331	818	164	674	CR	INV. EX.
2	B	862	339	60	18	346	266	339	890	166	608	CR	INV. EX.
3	B	788	400	54	19	346	266	327	864	184	766	CR	INV. EX.
4	B	<u>743</u>	210	54	15	346	266	328	967	173	506	CR	COMP. EX.
5	B	938	221	54	16	346	266	327	997	191	853	CR	INV. EX.
6	B	<u>965</u>	295	59	19	346	266	335	835	214	713	CR	COMP. EX.
7	B	856	62	63	10	346	266	340	952	196	918	CR	INV. EX.
8	B	864	<u>8</u>	78	13	346	266	341	953	210	643	CR	COMP. EX.
9	B	872	999	59	11	346	266	331	849	162	614	CR	INV. EX.
10	B	880	<u>1015</u>	78	13	346	266	332	818	161	918	CR	COMP. EX.
11	B	890	214	21	12	346	266	331	975	170	659	CR	INV. EX.
12	B	857	327	<u>15</u>	11	346	266	340	993	217	650	CR	COMP. EX.
13	B	860	468	73	15	346	266	338	861	207	806	CR	INV. EX.
14	B	890	274	50	14	346	266	336	814	169	613	CR	INV. EX.
15	B	875	467	65	7	346	266	335	995	175	732	CR	INV. EX.
16	B	832	247	64	<u>4</u>	346	266	337	973	212	760	CR	COMP. EX.
17	B	861	457	67	27	346	266	341	974	158	874	CR	INV. EX.
18	B	853	441	56	<u>38</u>	346	266	329	993	203	696	CR	COMP. EX.

(continued)

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50°C)-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type"	
19	B	883	203	65	12	346	266	268	898	189	782	CR	INV. EX.
20	B	859	463	60	18	346	266	<u>37</u>	903	167	831	CR	COMP. EX.
21	B	888	399	67	14	346	266	<u>349</u>	974	165	556	CR	COMP. EX.
22	B	899	257	52	18	346	266	<u>496</u>	803	185	669	CR	COMP. EX.
23	B	835	445	76	19	346	266	329	312	209	616	CR	INV. EX.
24	B	871	330	66	14	346	266	328	<u>284</u>	211	876	CR	COMP. EX.
25	B	892	430	55	16	346	266	336	<u>34</u>	152	909	CR	COMP. EX.
26	B	856	406	56	12	346	266	330	833	166	695	CR	INV. EX.
27	B	842	393	77	10	346	266	332	822	111	764	CR	INV. EX.
28	B	874	477	77	15	346	266	329	900	121	638	CR	INV. EX.
29	B	872	321	59	10	346	266	335	867	389	922	CR	INV. EX.
30	B	856	237	78	16	346	266	334	957	381	654	CR	INV. EX.
31	B	871	237	70	18	346	266	327	961	159	23	CR	INV. EX.
32	B	854	345	58	11	346	266	329	868	208	12	CR	INV. EX.
33	B	846	277	73	11	346	266	338	928	195	9860	CR	INV. EX.
34	B	856	232	66	11	346	266	329	954	211	9982	CR	INV. EX.
35	B	850	367	803	20	346	266	326	926	212	983	CR	INV. EX.
36	B	834	347	977	18	346	266	330	923	154	514	CR	INV. EX.

5

10

15

20

25

30

35

40

45

50

55

(continued)

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50°C)-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type"	
37	C	846	433	67	15	346	266	64	956	182	775	CR	COMP. EX.
38	D	860	466	78	12	304	224	554	869	167	659	CR	COMP. EX.
39	D	763	206	54	17	304	224	287	895	162	978	CR	INV. EX.

Underlines indicate being outside the range of the present invention.
 (*)CR: cold rolled steel sheet (no coating), GI: hot-dip galvanized steel sheet (no alloying of zinc coating), GA: galvanized steel sheet, EG: electrogalvanized steel sheet

[Table 4]

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50°C)-quench start temp. T2 (°C/s)	Ms (°C)	(M-s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type*	
40	D	935	228	62	19	304	224	295	925	204	792	CR	INV. EX.
41	D	857	83	80	19	304	224	298	808	206	617	CR	INV. EX.
42	D	831	986	73	11	304	224	293	861	183	515	CR	INV. EX.
43	D	864	475	24	14	304	224	285	909	184	585	CR	INV. EX.
44	D	844	460	855	12	304	224	289	807	196	772	CR	INV. EX.
45	D	869	390	51	8	304	224	290	814	158	962	CR	INV. EX.
46	D	847	480	59	29	304	224	286	996	197	754	CR	INV. EX.
47	D	860	271	65	15	304	224	232	976	175	949	CR	INV. EX.
48	D	830	372	63	13	304	224	303	994	186	677	CR	INV. EX.
49	D	882	367	53	17	304	224	286	324	203	911	CR	INV. EX.
50	D	886	205	75	11	304	224	299	879	168	999	CR	INV. EX.
51	D	865	221	68	13	304	224	292	983	114	953	CR	INV. EX.
52	D	866	422	55	18	304	224	288	839	391	967	CR	INV. EX.
53	D	894	407	66	19	304	224	286	913	210	11	CR	INV. EX.
54	D	842	306	77	18	304	224	294	875	205	9910	CR	INV. EX.
55	D	895	231	821	15	304	224	297	963	196	814	CR	INV. EX.
56	D	864	377	884	13	304	224	290	913	190	603	CR	INV. EX.
57	D	888	325	74	11	304	224	299	920	189	545	CR	INV. EX.
58	D	839	359	56	10	304	224	30	945	196	685	CR	COMP. EX.
59	D	873	456	66	12	304	224	529	928	212	959	GA	COMP. EX.
60	D	882	470	76	18	304	224	289	894	190	730	GA	INV. EX.

(continued)

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50°C)-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type*	
61	D	862	346	62	20	304	224	298	895	209	746	GA	INV. EX.
62	D	884	494	66	10	304	224	287	874	186	903	GA	INV. EX.
63	D	864	376	70	15	304	224	286	998	181	766	EG	INV. EX.
64	D	851	262	68	11	304	224	287	838	207	774	CR	INV. EX.
65	E	851	485	76	12	308	228	302	849	185	973	CR	INV. EX.
66	F	836	382	62	15	421	341	410	938	205	944	GA	INV. EX.
67	G	836	481	64	16	430	350	418	884	165	522	GA	COMP. EX.
68	H	834	439	71	12	223	143	205	849	151	814	GI	INV. EX.
69	I	862	324	72	10	192	112	180	841	199	764	GA	COMP. EX.
70	J	857	236	54	15	344	264	337	992	158	984	GA	INV. EX.
71	K	874	381	61	20	338	258	327	963	357	601	GA	COMP. EX.
72	L	836	465	64	17	352	272	339	919	218	652	GA	INV. EX.
73	M	855	387	75	19	346	266	333	922	214	889	GI	COMP. EX.
74	N	892	340	52	19	403	323	388	858	169	964	GA	INV. EX.
75	O	856	318	62	15	418	338	411	932	192	732	GA	COMP. EX.
76	P	835	255	75	10	282	202	277	929	219	913	GA	INV. EX.
77	Q	848	466	54	10	270	190	261	867	191	656	GA	COMP. EX.

5
10
15
20
25
30
35
40
45
50
55

(continued)

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50°C)-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type*	INV. EX.
78	R	856	291	73	18	339	259	326	871	173	662	GA	INV. EX.
Underlines indicate being outside the range of the present invention. (*)CR: cold rolled steel sheet (no coating), GI: hot-dip galvanized steel sheet (no alloying of zinc coating), GA: galvanized steel sheet, EG: electrogalvanized steel sheet													

[Table 5]

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50)°C-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type*	
79	S	846	498	61	16	351	271	346	804	157	773	GI	COMP. EX.
80	T	839	377	76	11	351	271	346	895	195	686	GA	INV.EX.
81	U	881	390	79	19	357	277	345	961	188	655	GA	COMP. EX.
82	V	867	483	67	20	346	266	339	905	196	934	GA	INV.EX.
83	W	878	413	52	14	362	282	352	961	158	551	GA	COMP. EX.
84	X	862	469	71	14	361	281	353	814	165	888	CR	INV.EX.
85	Y	853	340	78	13	351	271	337	853	151	732	CR	COMP. EX.
86	Z	836	202	61	10	349	269	334	876	151	876	GA	INV.EX.
87	AA	848	423	65	10	338	258	319	955	206	924	GA	COMP. EX.
88	AB	876	268	80	16	360	280	352	940	183	516	GA	INV.EX.
89	AC	876	306	61	13	354	274	343	807	152	718	GA	INV.EX.
90	AD	883	326	66	15	346	266	339	868	158	681	GA	COMP. EX.
91	AE	833	424	55	11	358	278	343	839	173	714	GA	INV.EX.
92	AF	843	202	72	17	347	267	339	854	183	577	GA	INV.EX.
93	AG	830	463	77	14	336	256	321	948	155	731	CR	COMP. EX.
94	AH	844	445	72	17	341	261	335	890	179	615	CR	INV.EX.
95	AI	845	216	51	13	355	275	343	959	196	893	CR	INV.EX.

(continued)

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50)°C-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type*	
96	AJ	887	482	69	19	349	269	337	867	164	767	CR	COMP. EX.
97	AK	839	374	76	18	350	270	344	974	168	812	CR	INV.EX.
98	AL	874	360	62	17	350	270	334	944	218	877	CR	INV.EX.
99	AM	881	250	58	11	334	254	326	893	182	546	CR	COMP. EX.
100	AN	785	250	72	16	360	280	348	899	182	888	CR	INV.EX.
101	AO	943	475	54	15	349	269	329	991	180	638	CR	INV.EX.
102	AP	846	23	72	19	355	275	346	921	166	938	CR	INV.EX.
103	AQ	880	851	52	13	328	248	322	959	200	521	CR	INV.EX.
104	AR	855	229	21	12	336	256	330	878	184	660	CR	INV.EX.
105	AS	864	233	915	11	350	270	338	809	176	853	CR	INV.EX.
106	AT	834	430	75	5	331	251	312	890	211	617	CR	INV.EX.
107	AU	874	437	72	27	359	279	341	964	156	567	CR	INV.EX.
108	AV	848	220	77	11	343	263	270	942	152	627	CR	INV.EX.
109	AW	844	376	69	18	341	261	339	807	177	662	CR	INV.EX.
110	AX	859	338	63	19	355	275	349	325	196	726	CR	INV.EX.
111	AY	844	441	75	19	343	263	331	925	194	706	CR	INV.EX.
112	AZ	834	329	78	13	347	267	329	850	108	628	CR	INV.EX.
113	BA	839	263	70	14	292	212	286	931	394	551	CR	INV.EX.
114	BB	846	487	75	13	304	224	294	940	175	23	CR	INV.EX.
115	BC	847	245	58	15	293	213	281	910	198	9851	CR	INV.EX.
116	BD	876	366	60	15	294	214	284	802	212	978	CR	INV.EX.

5
10
15
20
25
30
35
40
45
50
55

(continued)

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50)°C-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type*	INV. EX.
117	BE	851	324	67	11	316	236	310	973	210	656	CR	INV. EX.
Underlines indicate being outside the range of the present invention. (*)CR: cold rolled steel sheet (no coating), GI: hot-dip galvanized steel sheet (no alloying of zinc coating), GA: galvanized steel sheet, EG: electrogalvanized steel sheet													

[Table 6]

Nos.	Steels	Annealing temp. T1 (°C)	Holding time t1 (s)	Average cooling rate in temperature range of 750-600°C (°C/s)	Average cooling rate in temperature range of (Ms+50°C)-quench start temp. T2 (°C/s)	Ms (°C)	(M- s-80) (°C)	Quench start temp. T2 (°C)	Cooling rate from T2 to 80°C (°C/s)	Tempering temp. T3 (°C)	Holding time t3 (°C)	Type*
118	BF	858	315	76	18	300	220	289	917	197	890	CR
119	BG	876	295	77	11	297	217	291	925	165	881	CR
120	BH	850	301	64	19	301	221	283	962	209	932	CR
121	BI	849	363	64	11	302	222	288	911	172	931	CR
122	BJ	880	310	20	30	331	251	420	1000	180	800	COMP. EX.
123	BK	841	381	987	20	359	279	351	929	172	945	CR
124	BL	876	461	72	20	303	223	289	977	155	531	CR
125	BM	868	317	815	18	294	214	283	970	191	894	CR
126	BN	850	264	78	15	402	322	391	869	168	759	CR
127	BO	896	468	872	11	412	332	396	980	182	587	CR
Underlines indicate being outside the range of the present invention.												
(*)CR: cold rolled steel sheet (no coating), GI: hot-dip galvanized steel sheet (no alloying of zinc coating), GA: galvanized steel sheet, EG: electrogalvanized steel sheet												

[0073] The high strength cold rolled steel sheets obtained as described above were used as test steels. Tensile characteristics, flatness in the width direction, and working embrittlement resistance were evaluated in accordance with the following test methods.

5 (Microstructure observation)

[0074] The amount of tempered martensite, the amount of retained austenite, the total amount of ferrite and bainitic ferrite, and the average grain size of prior austenite were determined by the methods described hereinabove.

10 (Proportion of packets having the largest area in prior austenite grains)

[0075] The average proportion of packets having the largest area in prior austenite grains was determined by the method described hereinabove.

15 (Tensile test)

[0076] A JIS No. 5 test specimen (gauge length: 50 mm, parallel section width: 25 mm) was sampled so that the longitudinal direction of the test specimen would be perpendicular to the rolling direction. A tensile test was performed in accordance with JIS Z 2241 under conditions where the crosshead speed was 1.67×10^{-1} mm/sec. YS and TS were thus measured. In the present invention, 1180 MPa or higher TS was determined to be acceptable, and 85% or more yield ratio (YR) was determined to be acceptable. YR is determined from the formula (2) below:

$$YR = 100 \times YS/TS \cdots (2)$$

25 (Flatness in the width direction)

[0077] The cold rolled steel sheets obtained as described above were analyzed to measure the flatness in the width direction. The measurement is illustrated in Fig. 2. Specifically, a sheet with a length of 500 mm in the rolling direction (coil width $\times$ 500 mm L $\times$ sheet thickness) was cut out from the coil and was placed on a surface plate in such a manner that the warp at the ends would face upward. The height on the steel sheet was measured with a contact displacement meter by continuously moving the stylus over the width. Based on the results, the steepness θ as an index of the flatness of the steel sheet shape was measured as illustrated in Fig. 2. The flatness was rated as "×" when the steepness was more than 0.02, as "o" when the steepness was more than 0.01 and 0.02 or less, and as "◎" when the steepness was 0.01 or less. The steel sheet was evaluated as "excellent in the flatness in the width direction" when the steepness was 0.02 or less.

(Working embrittlement resistance)

[0078] The working embrittlement resistance was evaluated by Charpy test. A Charpy test specimen was a 2 mm deep V-notched test piece that was a stack of steel sheets fastened together with bolts to eliminate any gaps between the steel sheets. The number of steel sheets that were stacked was controlled so that the thickness of the stack as the test piece would be closer to 10 mm. When, for example, the sheet thickness was 1.2 mm, eight sheets were stacked to give a 9.6 mm thick test piece. The sheets for stacking into the Charpy test specimen were sampled so that the width direction would be the longitudinal direction. As an index of the working embrittlement resistance, the ratio $vE_{0\%}/vE_{10\%}$ of the absorbed impact energy at room temperature of the as-produced (unworked) steel sheet to that of the steel sheet after 10% rolling was measured. The working embrittlement resistance was rated as "×" when $vE_{0\%}/vE_{10\%}$ was less than 0.6, as "o" when $vE_{0\%}/vE_{10\%}$ was 0.6 or more and less than 0.7, and as "◎" when $vE_{0\%}/vE_{10\%}$ was 0.7 or more. The Charpy test specimen was evaluated as "excellent in working embrittlement resistance" when $vE_{0\%}/vE_{10\%}$ was 0.6 or more. Conditions other than those described above conformed to JIS Z 2242: 2018.

[0079] The results are described in Tables 7 to 10. As shown in the tables, INVENTIVE EXAMPLES achieved 1180 MPa or higher TS, 85% or more YR, excellent flatness in the width direction, and excellent working embrittlement resistance. In contrast, COMPARATIVE EXAMPLES were unsatisfactory in one or more of TS, YR, flatness in the width direction, and working embrittlement resistance.

55

[Table 7]

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior austenite grain size (μm)	YS (MPa)	TS (MPa)	YR (%)	Flatness in width direction	Working embrittlement resistance	
1	A	100	0	0	53	11	1433	1592	90	⊙	⊙	INV. EX.
2	B	99	1	0	57	8	1444	1570	92	⊙	⊙	INV. EX.
3	B	91	1	8	49	11	1189	1383	86	⊙	⊙	INV. EX.
4	B	<u>72</u>	0	<u>28</u>	52	13	745	<u>1020</u>	<u>73</u>	⊙	⊙	COMP. EX.
5	B	96	0	4	57	16	1340	1473	91	⊙	○	INV. EX.
6	B	98	0	2	51	<u>24</u>	1330	1478	90	⊙	×	COMP. EX.
7	B	92	0	7	59	8	1191	1385	86	⊙	⊙	INV. EX.
8	B	<u>72</u>	0	<u>17</u>	52	13	790	<u>964</u>	<u>82</u>	⊙	⊙	COMP. EX.
9	B	98	1	2	52	18	1400	1556	90	⊙	○	INV. EX.
10	B	98	1	1	45	<u>24</u>	1402	1558	90	⊙	×	COMP. EX.
11	B	91	1	9	51	11	1207	1404	86	⊙	⊙	INV. EX.
12	B	<u>72</u>	1	<u>14</u>	51	15	782	<u>954</u>	<u>82</u>	⊙	⊙	COMP. EX.
13	B	98	0	2	53	15	1340	1489	90	⊙	⊙	INV. EX.
14	B	98	1	1	59	15	1345	1546	87	⊙	⊙	INV. EX.
15	B	92	0	8	53	14	1247	1417	88	⊙	⊙	INV. EX.
16	B	<u>72</u>	1	<u>12</u>	57	10	778	<u>961</u>	<u>81</u>	⊙	⊙	COMP. EX.
17	B	96	0	4	67	11	1415	1522	93	○	○	INV. EX.
18	B	95	1	4	<u>89</u>	8	1292	1435	90	×	×	COMP. EX.
19	B	98	2	0	49	9	1334	1516	88	⊙	⊙	INV. EX.

(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior austenite grain size (μm)	YS (MPa)	TS (MPa)	YR (%)	Flatness in width direction	Working embrittlement resistance	
20	B	91	<u>7</u>	2	56	10	1099	1409	<u>78</u>	⊙	⊙	COMP. EX.
21	B	97	0	3	<u>95</u>	14	1379	1532	90	×	×	COMP. EX.
22	B	98	0	2	<u>79</u>	8	1400	1522	92	×	×	COMP. EX.
23	B	97	2	1	52	11	1305	1466	89	⊙	⊙	INV. EX.
24	B	94	<u>6</u>	0	47	13	1150	1403	<u>82</u>	⊙	⊙	COMP. EX.
25	B	95	<u>5</u>	0	59	8	1254	1511	<u>83</u>	⊙	⊙	COMP. EX.
26	B	96	1	3	45	8	1329	1510	88	⊙	⊙	INV. EX.
27	B	98	1	2	45	11	1437	1633	88	⊙	○	INV. EX.
28	B	99	0	0	50	14	1507	1638	92	⊙	○	INV. EX.
29	B	98	1	1	51	11	1107	1216	91	⊙	⊙	INV. EX.
30	B	99	0	1	58	11	1136	1248	91	⊙	⊙	INV. EX.
31	B	97	1	2	47	13	1341	1541	87	⊙	○	INV. EX.
32	B	97	0	3	54	13	1364	1467	93	⊙	○	INV. EX.
33	B	98	0	2	56	9	1417	1507	94	⊙	⊙	INV. EX.
Underlines indicate being outside the range of the present invention.												

[Table 8]

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior austenite grain size (μm)	YS (MPa)	TS (MPa)	YR (%)	Flatness in width direction	Working embrittlement resistance	
34	B	98	1	2	46	11	1305	1483	88	⊙	⊙	INV. EX.
35	B	99	1	0	47	9	1336	1501	89	⊙	⊙	INV. EX.
36	B	98	1	1	60	14	1427	1568	91	⊙	⊙	INV. EX.
37	C	94	<u>6</u>	0	52	13	1188	1431	<u>83</u>	⊙	⊙	COMP. EX.
38	D	99	1	0	<u>79</u>	12	1606	1825	88	×	×	COMP. EX.
39	D	92	1	8	50	11	1455	1692	86	⊙	⊙	INV. EX.
40	D	96	1	3	46	17	1487	1709	87	⊙	○	INV. EX.
41	D	91	1	8	53	13	1365	1606	85	⊙	⊙	INV. EX.
42	D	99	0	0	54	18	1675	1801	93	⊙	⊙	INV. EX.
43	D	93	1	6	54	8	1478	1679	88	⊙	○	INV. EX.
44	D	100	0	0	50	13	1693	1801	94	⊙	⊙	INV. EX.
45	D	91	1	8	47	8	1460	1678	87	⊙	⊙	INV. EX.
46	D	99	1	0	69	13	1655	1780	93	○	○	INV. EX.
47	D	97	2	1	47	14	1596	1773	90	⊙	⊙	INV. EX.
48	D	95	1	4	55	9	1493	1716	87	⊙	⊙	INV. EX.
49	D	97	2	1	58	9	1558	1731	90	⊙	⊙	INV. EX.
50	D	99	0	1	51	8	1677	1823	92	⊙	⊙	INV. EX.
51	D	98	0	2	60	13	1752	1884	93	⊙	○	INV. EX.
52	D	98	1	1	56	10	1351	1469	92	⊙	⊙	INV. EX.
53	D	97	1	3	50	14	1565	1720	91	⊙	○	INV. EX.
54	D	98	1	2	55	13	1608	1748	92	⊙	⊙	INV. EX.
55	D	97	0	3	50	9	1602	1741	92	⊙	⊙	INV. EX.
56	D	99	0	1	55	15	1665	1790	93	⊙	⊙	INV. EX.

(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior austenite grain size (μm)	YS (MPa)	TS (MPa)	YR (%)	Flatness in width direction	Working embrittlement resistance	
57	D	96	1	3	54	10	1559	1732	90	⊙	⊙	INV. EX.
58	D	90	<u>7</u>	3	55	12	1185	1601	<u>74</u>	⊙	⊙	COMP. EX.
59	D	98	0	1	<u>93</u>	12	1563	1737	90	×	×	COMP. EX.
60	D	98	1	1	59	13	1558	1770	88	⊙	⊙	INV. EX.
61	D	99	0	0	54	11	1621	1762	92	⊙	⊙	INV. EX.
62	D	99	1	1	49	15	1616	1796	90	⊙	⊙	INV. EX.
63	D	96	1	4	53	8	1570	1744	90	⊙	⊙	INV. EX.
64	D	98	0	2	46	11	1553	1745	89	⊙	⊙	INV. EX.
65	E	100	0	0	57	13	1689	1797	94	⊙	⊙	INV. EX.
66	F	91	0	9	50	15	1067	1226	87	⊙	⊙	INV. EX.
67	G	<u>72</u>	1	<u>20</u>	49	14	358	<u>471</u>	<u>76</u>	⊙	⊙	COMP. EX.
Underlines indicate being outside the range of the present invention.												

[Table 9]

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior austenite grain size (μm)	YS (MPa)	TS (MPa)	YR (%)	Flatness in width direction	Working embrittlement resistance	
68	H	98	0	2	47	12	1915	2037	94	⊙	○	INV. EX.
69	I	98	1	1	46	14	1808	2055	88	⊙	×	COMP. EX.
70	J	98	0	2	59	10	1451	1560	93	⊙	⊙	INV. EX.
71	K	96	1	3	47	11	1004	<u>1154</u>	87	⊙	⊙	COMP. EX.
72	L	97	2	1	52	9	1327	1543	86	⊙	⊙	INV. EX.
73	M	92	<u>8</u>	1	53	8	1192	1528	<u>78</u>	⊙	⊙	COMP. EX.
74	N	91	0	9	55	12	1093	1228	89	⊙	⊙	INV. EX.
75	O	<u>72</u>	0	<u>19</u>	55	9	664	<u>820</u>	<u>81</u>	⊙	⊙	COMP. EX.
76	P	99	1	1	52	12	1409	1601	88	⊙	○	INV. EX.
77	Q	96	0	4	49	14	1492	1622	92	⊙	×	COMP. EX.
78	R	97	0	3	53	14	1413	1536	92	⊙	○	INV. EX.
79	S	99	0	0	55	12	1390	1528	91	⊙	×	COMP. EX.
80	T	99	0	1	58	13	1398	1487	94	⊙	○	INV. EX.
81	U	98	1	1	52	11	1322	1437	92	⊙	×	COMP. EX.
82	V	93	1	6	59	14	1049	1220	86	⊙	⊙	INV. EX.
83	W	<u>72</u>	0	<u>22</u>	47	8	782	<u>1002</u>	<u>78</u>	⊙	⊙	COMP. EX.
84	X	96	1	3	56	8	1282	1424	90	⊙	○	INV. EX.
85	Y	98	1	1	47	13	1383	1520	91	⊙	×	COMP. EX.

(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior austenite grain size (μm)	YS (MPa)	TS (MPa)	YR (%)	Flatness in width direction	Working embrittlement resistance	
86	Z	96	0	3	55	12	1309	1488	88	⊙	o	INV. EX.
87	AA	98	0	2	59	12	1383	1503	92	⊙	×	COMP. EX.
88	AB	100	0	0	50	9	1356	1490	91	⊙	⊙	INV. EX.
89	AC	97	1	2	48	8	1424	1582	90	⊙	○	INV. EX.
90	AD	98	1	1	56	9	1448	1646	88	⊙	×	COMP. EX.
91	AE	98	0	1	46	11	1376	1464	94	⊙	⊙	INV. EX.
92	AF	99	1	1	54	13	1384	1555	89	⊙	○	INV. EX.
93	AG	98	1	2	51	11	1431	1645	87	⊙	×	COMP. EX.
94	AH	99	0	1	58	10	1448	1574	92	⊙	⊙	INV. EX.
95	AI	97	1	2	47	14	1480	1626	91	⊙	○	INV. EX.
96	AJ	97	1	2	56	15	1528	1717	89	⊙	×	COMP. EX.
97	AK	95	1	4	59	8	1243	1429	87	⊙	⊙	INV. EX.
98	AL	97	1	3	55	13	1363	1498	91	⊙	○	INV. EX.
99	AM	98	1	1	53	11	1430	1589	90	⊙	×	COMP. EX.
Underlines indicate being outside the range of the present invention.												

[Table 10]

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior austenite grain size (μm)	YS (MPa)	TS (MPa)	YR (%)	Flatness in width direction	Working embrittlement resistance	
100	AN	93	0	7	48	9	1155	1343	86	⊙	⊙	INV. EX.
101	AO	97	1	3	51	19	1318	1481	89	⊙	○	INV. EX.
102	AP	92	0	7	47	13	1221	1357	90	⊙	⊙	INV. EX.
103	AQ	96	0	4	55	17	1303	1481	88	⊙	○	INV. EX.
104	AR	91	1	8	52	11	1208	1405	86	⊙	⊙	INV. EX.
105	AS	97	1	2	51	15	1362	1497	91	⊙	⊙	INV. EX.
106	AT	92	0	7	46	14	1144	1330	86	⊙	⊙	INV. EX.
107	AU	97	1	2	68	11	1340	1472	91	○	○	INV. EX.
108	AV	95	2	3	55	10	1270	1494	85	⊙	⊙	INV. EX.
109	AW	98	0	2	45	14	1391	1563	89	⊙	⊙	INV. EX.
110	AX	96	2	2	50	14	1232	1416	87	⊙	⊙	INV. EX.
111	AY	96	0	3	51	8	1344	1461	92	⊙	⊙	INV. EX.
112	AZ	98	1	1	49	12	1487	1634	91	⊙	○	INV. EX.
113	BA	97	1	2	47	14	1369	1504	91	⊙	⊙	INV. EX.
114	BB	98	0	2	55	12	1599	1777	90	⊙	○	INV. EX.
115	BC	97	0	3	60	8	1657	1801	92	⊙	⊙	INV. EX.
116	BD	95	1	4	49	13	1517	1744	87	⊙	⊙	INV. EX.
117	BE	97	1	2	49	12	1444	1660	87	⊙	⊙	INV. EX.
118	BF	99	0	1	52	15	1638	1820	90	⊙	⊙	INV. EX.
119	BG	98	0	2	46	9	1736	1847	94	⊙	⊙	INV. EX.
120	BH	99	1	1	50	9	1615	1755	92	⊙	⊙	INV. EX.
121	BI	98	1	1	58	10	1636	1798	91	⊙	⊙	INV. EX.
122	BJ	89	0	1	88	9	1360	1600	85	×	×	COMP. EX.

5

10

15

20

25

30

35

40

45

50

55

(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior austenite grain size (μm)	YS (MPa)	TS (MPa)	YR (%)	Flatness in width direction	Working embrittlement resistance	
123	BK	97	1	2	56	12	1303	1432	91	⊙	⊙	INV. EX.
124	BL	98	1	1	52	13	1595	1812	88	⊙	⊙	INV. EX.
125	BM	97	0	3	51	12	1619	1799	90	⊙	⊙	INV. EX.
126	BN	98	0	1	59	10	1128	1226	92	⊙	⊙	INV. EX.
127	BO	97	0	3	45	12	1101	1184	93	⊙	⊙	INV. EX.
Underlines indicate being outside the range of the present invention.												

Claims

1. A high strength steel sheet having a chemical composition comprising, 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,

a balance being Fe and incidental impurities,

the high strength steel sheet being such that in a region at 1/4 sheet thickness,

an area fraction of tempered martensite is 90% or more,

a volume fraction of retained austenite is less than 3%,

an area fraction of a total of ferrite and bainitic ferrite is less than 10%,

an average grain size of prior austenite is 20 μm or less, and

an average of proportions of packets having a largest area in prior austenite grains is 70% by area or less of the prior austenite grain.

2. The high strength steel sheet according to claim 1, wherein the chemical composition further comprises at least one element selected from, in mass%,

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,

Co: 0.010% or less, Ni: 1.00% 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.

3. The high strength steel sheet according to claim 1 or 2, which has a coated layer on a surface of the steel sheet.

4. A method for manufacturing the high strength steel sheet described in claim 1 or 2, the method comprising:

providing a cold rolled steel sheet produced by subjecting a steel having the chemical composition to hot rolling, pickling, and cold rolling;

heating the steel sheet at an annealing temperature T1 of 750°C or above and 950°C or below for a holding time t1 at the annealing temperature T1 of 10 seconds or more and 1000 seconds or less;

cooling the steel sheet in such a manner that:

an average cooling rate from 750°C to 600°C is 20°C/s or more,

an average cooling rate from (Ms + 50°C) to a quench start temperature T2 is 5°C/s or more and 30°C/s or less

wherein the quench start temperature T2 is (Ms - 80°C) or above and is below Ms where Ms is martensite start temperature (°C) defined by formula (1), and

an average cooling rate from the quench start temperature T2 to 80°C is 300°C/s or more; and

heating the steel sheet at a tempering temperature T3 of 100°C or above and 400°C or below for a holding time t3 at the tempering temperature T3 of 10 seconds or more and 10000 seconds or less,

$$Ms = 519 - 474 \times [\%C] - 30.4 \times [\%Mn] - 12.1 \times [\%Cr] - 7.5 \times [\%Mo] - 17.7 \times [\%Ni] \quad (1)$$

wherein [% C], [% Mn], [% Cr], [% Mo], and [% Ni] indicate contents (mass%) of C, Mn, Cr, Mo, and Ni, respectively, and

are zero when the element is absent.

5. The method for manufacturing the high strength steel sheet according to claim 4, further comprising performing a coating treatment.

5

10

15

20

25

30

35

40

45

50

55

FIG. 1

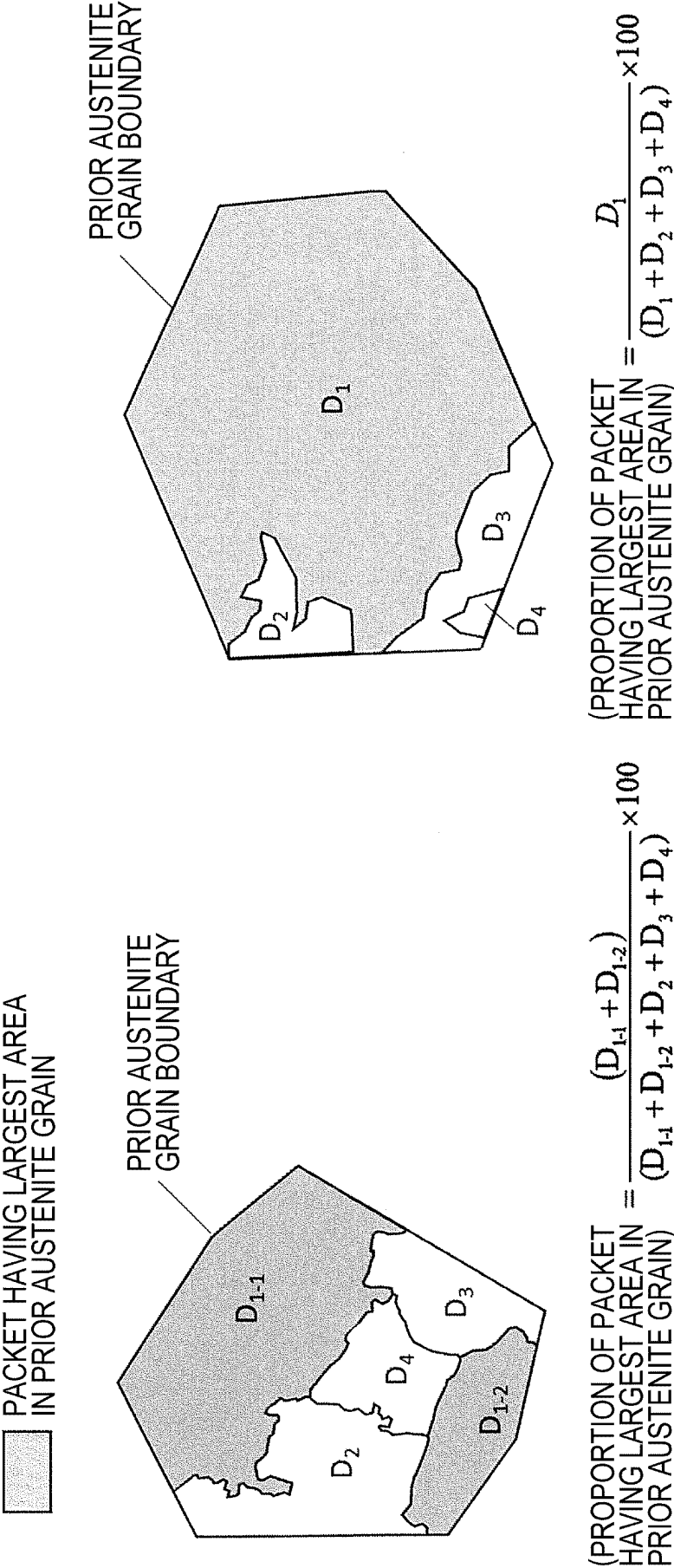
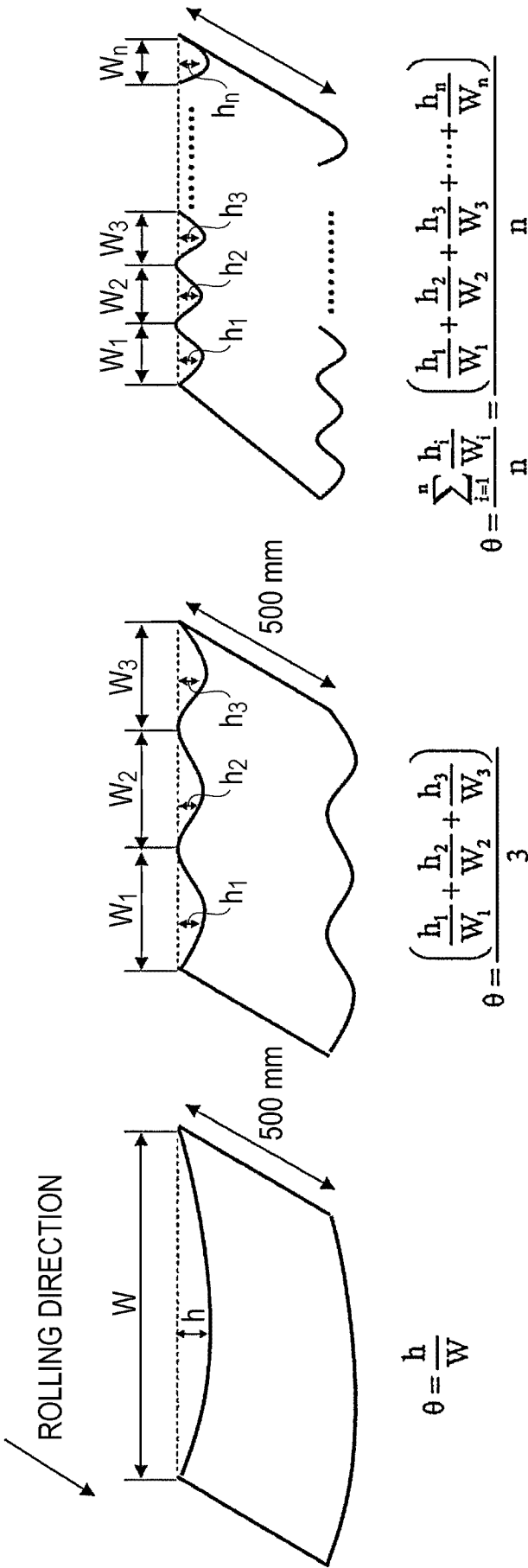


FIG. 2



INTERNATIONAL SEARCH REPORT

International application No.

PCT/JP2023/002913

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/00(2006.01)i; **C21D 9/46**(2006.01)i; **C22C 38/06**(2006.01)i; **C22C 38/60**(2006.01)i
 FI: C22C38/00 301U; C22C38/00 301T; C22C38/06; C22C38/60; C21D9/46 F; C21D9/46 J

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)

C22C38/00-38/60; C21D9/46

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 2014/129379 A1 (KOBE STEEL LTD) 28 August 2014 (2014-08-28) claims	1-5
A	JP 2017-503072 A (ARCELORMITTAL) 26 January 2017 (2017-01-26) claims	1-5
A	JP 2021-105197 A (NIPPON STEEL CORP) 26 July 2021 (2021-07-26) claims, paragraphs [0066]-[0076]	1-5
A	JP 2013-104081 A (KOBE STEEL LTD) 30 May 2013 (2013-05-30) claims, tables 5, 6	1-5
P, A	WO 2022/209519 A1 (JFE STEEL CORP) 06 October 2022 (2022-10-06) claims, paragraphs [0054]-[0058], [0068]-[0074]	1-5

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

* Special categories of cited documents:

“A” document defining the general state of the art which is not considered to be of particular relevance

“E” earlier application or patent but published on or after the international filing date

“L” document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified)

“O” document referring to an oral disclosure, use, exhibition or other means

“P” document published prior to the international filing date but later than the priority date claimed

“T” later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention

“X” document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone

“Y” document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art

“&” document member of the same patent family

Date of the actual completion of the international search

06 April 2023

Date of mailing of the international search report

18 April 2023

Name and mailing address of the ISA/JP

Japan Patent Office (ISA/JP)
 3-4-3 Kasumigaseki, Chiyoda-ku, Tokyo 100-8915
 Japan

Authorized officer

Telephone No.

INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/JP2023/002913

Patent document cited in search report			Publication date (day/month/year)	Patent family member(s)		Publication date (day/month/year)
WO	2014/129379	A1	28 August 2014	EP	2960353	A1
				claims		
				US	2015/0368742	A1
				CN	105074040	A
				JP	2014-159610	A
JP	2017-503072	A	26 January 2017	US	2016/0304981	A1
				claims		
				WO	2015/088514	A1
				CN	106164319	A
JP	2021-105197	A	26 July 2021	(Family: none)		
JP	2013-104081	A	30 May 2013	(Family: none)		
WO	2022/209519	A1	06 October 2022	(Family: none)		

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

REFERENCES CITED IN THE DESCRIPTION

This list of references cited by the applicant is for the reader's convenience only. It does not form part of the European patent document. Even though great care has been taken in compiling the references, errors or omissions cannot be excluded and the EPO disclaims all liability in this regard.

Patent documents cited in the description

- JP 4947176 B [0007]
- JP 6525114 B [0007]
- JP 6777272 B [0007]

Non-patent literature cited in the description

- *Journal of Smart Processing*, 2013, vol. 2 (3), 110-118
[0008]