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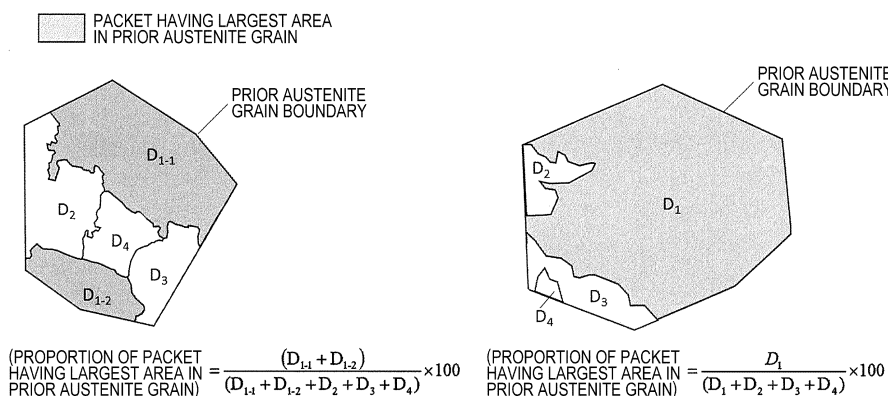
(54) HIGH-STRENGTH STEEL SHEET AND METHOD FOR PRODUCING SAME

(57) Objects are to provide a high strength steel sheet having 980 MPa or higher TS and being excellent in press formability, 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

38% or more and less than 90%, the volume fraction of retained austenite is less than 3%, the area fraction of the total of ferrite and bainitic ferrite is 100 or more and 60% or less, 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



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Description

Technical Field

[0001] The present invention relates to a high strength steel sheet excellent in tensile strength, press formability, 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] High strength steel sheets used in automobiles require excellent press formability. For example, high strength steel sheets with high EI and excellent hole expansion ratio λ are suitably applied to automobile frame parts, such as bumpers. From the point of view of crash safety, excellent working embrittlement resistance is required.

[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 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 2 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

[0006]

PTL 1: Japanese Patent No. 4947176

PTL 2: Japanese Patent No. 6777272

Non Patent Literature

[0007] NPL 1: Journal of Smart Processing, 2013, Vol. 2, No. 3, pp. 110-118

Summary of Invention

Technical Problem

[0008] 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 980 MPa or higher TS and being excellent in press formability, flatness in the width direction, and working embrittlement resistance; and to provide a method for manufacturing the same.

Solution to Problem

[0009] The present inventors carried out extensive studies directed to solving the problems described above and have consequently found the following facts.

(1) 980 MPa or higher TS and excellent press formability can be realized by limiting the amount of tempered martensite

to 38% or more and less than 90%, the amount of the total of ferrite and bainitic ferrite to 10% or more and 60% or less, and the amount of retained austenite to less than 3%.

(2) 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.

(3) 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.

[0010] 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 38% or more and less than 90%, a volume fraction of retained austenite is less than 3%, an area fraction of the total of ferrite and bainitic ferrite is 10% or more and 60% or less, 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 according to [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 700°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 the average cooling rate from 750°C to 600°C is less than 20°C/s, the average cooling rate from (Ms + 50°C) to a quench start temperature T2 is less than 5°C/s 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 the 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,

$$Ms = 519 - 474 \times [\% \text{ C}] - 30.4 \times [\% \text{ Mn}] - 12.1 \times [\% \text{ Cr}] - 7.5 \times [\% \text{ Mo}] - 17.7 \times [\% \text{ Ni}] - T1/80 \quad (1)$$

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

[0011] According to the present invention, a high strength steel sheet can be obtained that has 980 MPa or higher TS and excels in press formability, 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

[0012]

[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

[0013] Embodiments of the present invention will be described below.

[0014] 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.

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

[0015] 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 martensite and the working embrittlement resistance. When the C content is less than 0.030%, the fraction of martensite is so small that realizing 980 MPa or higher TS is difficult. When, on the other hand, the C content is more than 0.500%, 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.

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

[0016] 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 980 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.

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

[0017] 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 martensite and the working embrittlement resistance. When the Mn content is less than 0.10%, the fraction of martensite is so small that realizing 980 MPa or higher TS is difficult. When, on the other hand, the Mn content is more than 5.00%, 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.

[P: 0.100% or less]

[0018] 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.

[S: 0.0200% or less]

[0019] 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]

[0020] Aluminum raises the A_3 transformation temperature to allow more ferrite to be contained in the microstructure. The fraction of martensite is correspondingly lowered to make it difficult to realize 980 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]

[0021] 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]

[0022] 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.

[0023] 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.

[0024] 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.

[0025] 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.

[0026] 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 some cases. 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.

[0027] 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.

[0028] 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.

[0029] 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.

[0030] 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.

[0031] 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.

[0032] 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.

[0033] 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.

[0034] 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.

[0035] 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.

[0036] 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.

[0037] 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.

[0038] Next, the steel microstructure of the high strength steel sheet of the present invention will be described.

[Area fraction of tempered martensite: 38% or more and less than 90%]

[0039] When the amount of tempered martensite is less than 38%, realizing 980 MPa or higher TS is difficult. When, on the other hand, the amount of tempered martensite is 90% or more, the amount of ferrite is lowered to cause a decrease in El and consequently press formability is lowered. Thus, the amount of tempered martensite is limited to 38% or more and less than 90%. The amount is preferably 40% or more. The amount is preferably 60% or less.

[0040] 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%]

[0041] This configuration is a very important requirement that constitutes the present invention. When the volume fraction of retained austenite is 3% or more, press formability is lowered. The reason for low press formability is that retained austenite with a high fraction gives rise to a lowering in λ 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%.

[0042] 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: 10% or more and 60% or less]

[0043] This configuration is a very important requirement that constitutes the present invention. When the total of ferrite and bainitic ferrite is less than 10%, El is lowered and consequently press formability is deteriorated. When, on the other hand, the total of ferrite and bainitic ferrite is more than 60%, realizing 980 MPa or higher TS is difficult. Thus, the total of ferrite and bainitic ferrite is limited to 10% or more and 60% or less. The total amount is preferably 35% or more. The total amount is preferably 55% or less.

[0044] 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 is recessed structures having a flat interior and containing no inner carbides. In the microstructure images, bainitic ferrite is recessed structures having a flat interior and containing inner carbides. The values thus obtained are combined and are averaged to determine the total of ferrite and bainitic ferrite.

[0045] 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]

[0046] 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.

[0047] 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]

[0048] 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.

[0049] 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 Ar_3 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: 700°C or above and 950°C or below]

[0057] When the annealing temperature T1 is below 700°C, the area fraction of the total of ferrite and bainitic ferrite is more than 60% to make it difficult to realize 980 MPa or higher TS. 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 μm to give rise to a decrease in working embrittlement resistance. Thus, the annealing temperature T1 is limited to 700°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 more than 60%. As a result, it is difficult to achieve 980 MPa or higher TS. 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: less than 20°C/s]

[0059] When the average cooling rate from 750°C to 600°C is 20°C/s or more, the area fraction of the total of ferrite and bainitic ferrite is less than 10% to cause a decrease in EI, thereby deteriorating press formability. For the reasons above, the average cooling rate from 750°C to 600°C is limited to less than 20°C/s. The average cooling rate is preferably 15°C/s or less.

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

[0060] This configuration is a very important requirement that constitutes the present invention. When the average cooling rate from (Ms + 50°C) to a quench start temperature T2 is 5°C/s or more, the area fraction of the total of ferrite and bainitic ferrite is less than 10% to cause a decrease in EI, thereby deteriorating press formability. For the reasons above, the average cooling rate from (Ms + 50°C) to a quench start temperature T2 is limited to less than 5°C/s. The average cooling rate is preferably 4°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 cause a decrease in press formability. 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. 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):

$$Ms = 519 - 474 \times [\% C] - 30.4 \times [\% Mn] - 12.1 \times [\% Cr] - 7.5 \times [\% Mo] - 17.7 \times [\% Ni] - T1/80 \quad (1)$$

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]

[0062] 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 cause a decrease in press formability. 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]

- 5 **[0063]** 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, martensite is excessively tempered to make it difficult to achieve 980 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]

- 15 **[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 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, martensite is excessively tempered to make it difficult to achieve 980 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.

[0065] 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.

- 25 **[0066]** 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.

[0067] When the high strength steel sheet is a product that is traded, the steel sheet is usually traded after being cooled to room temperature.

- 30 **[0068]** The high strength steel sheet may be subjected to coating treatment during annealing or after annealing.

- [0069]** 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 less than 20°C/s. 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.

- 40 **[0070]** After the coating treatment after annealing, the steel sheet may be worked again 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.

EXAMPLES

- 50 **[0071]** 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 5. 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.

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[Table 1]

Steels	Chemical composition (mass%)													
	C	Si	Mn	P	S	N	O	Al	Ti	B	Nb	Cu	Others	
A	0.215	0.280	2.24	0.006	0.0011	0.005	0.006	0.047						INV. EX.
B	0.217	0.298	1.98	0.005	0.0009	0.006	0.005	0.053						INV. EX.
C	0.192	0.251	2.04	0.009	0.0008	0.002	0.003	0.015						INV. EX.
D	0.111	1.332	2.04	0.011	0.0007	0.004	0.005	0.024						INV. EX.
E	0.113	1.464	2.22	0.010	0.0014	0.006	0.002	0.018						INV. EX.
F	0.048	0.262	2.25	0.008	0.0011	0.004	0.002	0.030						INV. EX.
G	<u>0.021</u>	0.168	2.30	0.007	0.0013	0.003	0.005	0.036						COMP. EX.
H	<u>0.468</u>	0.166	2.09	0.009	0.0012	0.006	0.002	0.059						INV. EX.
I	<u>0.522</u>	0.248	2.00	0.010	0.0015	0.001	0.007	0.015						COMP. EX.
J	0.212	0.073	2.13	0.015	0.0010	0.005	0.006	0.052						INV. EX.
K	0.191	<u>0.002</u>	2.23	0.013	0.0006	0.002	0.006	0.020						COMP. EX.
L	0.210	2.339	2.22	0.006	0.0010	0.006	0.007	0.054						INV. EX.
M	0.205	<u>2.532</u>	1.98	0.012	0.0007	0.002	0.006	0.025						COMP. EX.
N	0.203	<u>0.273</u>	0.27	0.006	0.0007	0.001	0.004	0.055						INV. EX.
O	0.187	0.307	<u>0.08</u>	0.012	0.0009	0.007	0.005	0.033						COMP. EX.
P	0.191	0.271	4.98	0.012	0.0011	0.005	0.002	0.052						INV. EX.
Q	0.197	0.162	<u>5.12</u>	0.007	0.0007	0.005	0.004	0.040						COMP. EX.
R	0.216	0.314	<u>2.12</u>	0.099	0.0005	0.002	0.003	0.012						INV. EX.
S	0.219	0.333	2.06	<u>0.121</u>	0.0010	0.004	0.003	0.049						COMP. EX.
T	0.219	0.173	2.25	0.006	0.0182	0.006	0.006	0.024						INV. EX.
U	0.205	0.192	2.06	0.010	<u>0.0222</u>	0.004	0.002	0.024						COMP. EX.
V	0.205	0.165	2.18	0.011	0.0010	0.002	0.003	0.976						INV. EX.
W	0.195	0.165	2.10	0.015	0.0009	0.005	0.002	<u>1.135</u>						COMP. EX.
X	0.214	0.204	2.06	0.011	0.0009	0.0089	0.007	0.059						INV. EX.
Y	0.192	0.186	1.93	0.012	0.0007	<u>0.0112</u>	0.002	0.032						COMP. EX.

(continued)

Steels	Chemical composition (mass%)													
	C	Si	Mn	P	S	N	O	Al	Ti	B	Nb	Cu	Others	
Z	0.182	0.229	2.22	0.011	0.0007	0.004	0.0090	0.037						INV. EX.
AA	0.206	0.271	2.19	0.009	0.0015	0.006	<u>0.0110</u>	0.040						COMP. EX.
AB	0.186	0.331	1.91	0.007	0.0010	0.005	0.001	0.038	0.002					INV. EX.
AC	0.190	0.327	2.29	0.013	0.0008	0.006	0.002	0.027	0.187					INV. EX.
AD	0.206	0.341	2.26	0.011	0.0012	0.006	0.003	0.038	<u>0.223</u>					COMP. EX.
AE	0.182	0.258	2.06	0.008	0.0009	0.005	0.005	0.038		0.0002				INV. EX.
AF	0.189	0.309	2.12	0.006	0.0005	0.007	0.003	0.048		0.0088				INV. EX.
AG	0.206	0.239	2.16	0.008	0.0009	0.003	0.005	0.031		<u>0.0121</u>				COMP. EX.
AH	0.187	0.164	2.27	0.006	0.0013	0.002	0.004	0.016			0.002			INV. EX.
AI	0.216	0.308	2.20	0.008	0.0011	0.002	0.003	0.031			0.189			INV. EX.
AJ	0.190	0.189	2.05	0.012	0.0006	0.003	0.007	0.051			<u>0.203</u>			COMP. EX.
AK	0.183	0.345	2.03	0.009	0.0013	0.002	0.005	0.023				0.03		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		
AL	0.205	0.293	1.98	0.013	0.0012	0.005	0.002	0.057				0.90		INV. EX.
AM	0.190	0.317	2.29	0.007	0.0010	0.003	0.007	0.052				1.11		COMP. EX.
AN	0.214	0.261	2.25	0.007	0.0010	0.005	0.004	0.050					V:0.070	INV. EX.
AO	0.205	0.309	2.02	0.007	0.0006	0.002	0.006	0.028					Ta:0.05	INV. EX.
AP	0.213	0.215	2.26	0.014	0.0006	0.006	0.002	0.048					W:0.03	INV. EX.
AQ	0.187	0.263	2.10	0.015	0.0007	0.005	0.003	0.036					Cr:0.87	INV. EX.
AR	0.184	0.228	2.06	0.009	0.0011	0.002	0.004	0.029					Mo:0.13	INV. EX.
AS	0.193	0.288	2.29	0.012	0.0005	0.006	0.004	0.028					Co:0.008	INV. EX.
AT	0.219	0.242	2.21	0.007	0.0012	0.002	0.003	0.034					Ni:0.33	INV. EX.
AU	0.185	0.245	1.92	0.012	0.0011	0.007	0.004	0.056					Sn:0.012	INV. EX.
AV	0.185	0.301	2.08	0.009	0.0006	0.005	0.004	0.026					Sb:0.005	INV. EX.
AW	0.196	0.274	1.90	0.010	0.0012	0.004	0.002	0.056					Ca:0.0015	INV. EX.
AX	0.218	0.339	1.95	0.009	0.0009	0.004	0.004	0.022					Mg:0.0086	INV. EX.
AY	0.186	0.343	2.010	0.005	0.0008	0.002	0.002	0.051					Zr:0.083	INV. EX.
AZ	0.217	0.240	1.990	0.008	0.0013	0.007	0.003	0.024					Te:0.092	INV. EX.
BA	0.102	1.364	2.270	0.010	0.0014	0.006	0.005	0.016					Hf:0.05	INV. EX.
BB	0.128	1.390	1.960	0.012	0.0005	0.005	0.004	0.037					REM:0.0092	INV. EX.
BC	0.137	1.402	2.020	0.012	0.0007	0.003	0.005	0.030					Bi:0.164	INV. EX.
BD	0.132	1.328	1.940	0.007	0.0006	0.004	0.003	0.012					Zn:0.03	INV. EX.
BE	0.113	1.482	2.090	0.008	0.0007	0.005	0.007	0.041					Pb:0.016	INV. EX.
BF	0.111	1.374	2.000	0.007	0.0008	0.005	0.002	0.011					As:0.040	INV. EX.
BG	0.116	1.362	2.080	0.011	0.0011	0.006	0.006	0.012					Ge:0.090	INV. EX.
BH	0.133	1.387	2.220	0.009	0.0009	0.001	0.003	0.052					Sr:0.065	INV. EX.
BI	0.108	1.310	2.150	0.007	0.0012	0.001	0.004	0.037					Cs:0.082	INV. EX.
BJ	0.198	0.870	2.700	0.010	0.0003	0.004	0.001	0.045	0.007	0.0017	0.014	0.18	Ni:0.05	INV. EX.

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(continued)

Steels	Chemical composition (mass%)													
	C	Si	Mn	P	S	N	O	Al	Ti	B	Nb	Cu	Others	
BK	0.218	0.326	2.060	0.009	0.0008	0.007	0.007	0.012						INV. EX.
BL	0.108	1.352	1.920	0.013	0.0011	0.003	0.004	0.056						INV. EX.
BM	0.105	1.331	2.030	0.007	0.0006	0.004	0.002	0.049						INV. EX.
BN	0.207	1.374	1.910	0.007	0.0013	0.003	0.001	0.057						INV. EX.
BO	0.189	1.414	2.020	0.009	0.0015	0.003	0.003	0.050						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. 13 (°C)	Holding time t3 (s)	Type"
1	A	796	322	13	2	330	250	321	905	203	854	CR
2	B	783	348	7	3	337	257	323	966	189	994	CR
3	B	717	427	12	2	337	257	320	875	172	961	CR
4	B	<u>692</u>	255	10	2	337	257	327	957	215	989	COMP. EX.
5	B	927	274	7	3	336	256	322	862	151	606	CR
6	B	<u>965</u>	399	12	3	335	255	316	882	186	555	COMP. EX.
7	B	758	63	9	3	337	257	320	914	188	784	CR
8	B	777	<u>8</u>	6	3	336	256	322	863	199	706	COMP. EX.
9	B	785	896	11	2	336	256	324	888	193	931	CR
10	B	783	<u>1015</u>	11	4	336	256	322	1000	206	590	COMP. EX.
11	B	753	311	17	4	336	256	324	960	198	929	CR
12	B	924	404	<u>25</u>	2	335	255	321	872	212	563	COMP. EX.
13	B	798	223	12	4	336	256	322	831	162	596	CR
14	B	792	243	9	2	336	256	327	836	198	954	CR
15	B	759	207	7	3	335	255	328	878	215	780	CR
16	B	785	335	11	3	335	255	318	832	203	862	CR
17	B	779	381	9	4	337	257	327	842	151	693	CR
18	B	804	203	12	<u>6</u>	337	257	331	865	177	714	CR
19	B	790	241	7	3	336	256	261	928	156	687	CR

(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. 13 (°C)	Holding time t3 (s)	Type"
20	B	803	260	7	4	336	256	20	847	158	671	CR COMP. EX.
21	B	761	335	7	3	336	256	409	894	165	571	CR COMP. EX.
22	B	797	376	14	4	336	256	631	978	168	603	CR COMP. EX.
23	B	771	369	13	3	336	256	316	312	212	510	CR INV. EX.
24	B	755	222	10	2	336	256	326	284	201	832	CR COMP. EX.
25	B	764	414	10	3	336	256	323	34	211	741	CR COMP. EX.
26	B	788	404	9	3	337	257	331	915	167	879	CR INV. EX.
27	B	768	222	11	3	336	256	329	995	111	996	CR INV. EX.
28	B	768	437	15	3	337	257	325	846	110	639	CR INV. EX.
29	B	784	321	8	2	337	257	322	910	389	763	CR INV. EX.
30	B	772	266	13	4	336	256	321	883	398	707	CR INV. EX.
31	B	755	465	8	3	337	257	328	821	161	23	CR INV. EX.
32	B	790	311	15	3	336	256	326	897	173	12	CR INV. EX.
33	B	786	304	8	4	336	256	317	854	210	9860	CR INV. EX.
34	B	779	485	7	3	336	256	324	811	196	9878	CR INV. EX.
35	B	751	282	7	4	336	256	330	951	196	726	CR INV. EX.
36	B	808	294	14	3	336	256	328	956	178	828	CR INV. EX.
37	C	799	211	14	3	336	256	20	899	152	622	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 (s)	Type"	
38	D	760	208	10	3	294	214	<u>544</u>	875	219	917	CR	COMP. EX.
39	D	719	490	6	3	295	215	281	818	195	563	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. 13 (°C)	Holding time t3 (s)	Type*
40	D	935	452	14	3	294	214	276	870	159	565	CR
41	D	798	77	10	3	294	214	287	938	193	689	CR
42	D	756	903	15	3	294	214	281	827	171	585	CR
43	D	805	421	19	3	295	215	278	837	216	737	CR
44	D	794	453	11	4	294	214	277	876	159	969	CR
45	D	786	473	6	3	294	214	279	962	151	846	CR
46	D	770	485	14	4	295	215	285	878	215	607	CR
47	D	805	453	9	2	294	214	216	877	191	949	CR
48	D	793	380	12	3	294	214	287	972	187	527	CR
49	D	751	328	11	4	295	215	282	324	177	723	CR
50	D	796	292	6	4	294	214	282	822	215	652	CR
51	D	766	311	10	2	295	215	288	857	114	508	CR
52	D	808	328	12	3	295	215	281	966	391	882	CR
53	D	790	421	13	2	295	215	286	890	216	12	CR
54	D	764	295	11	3	295	215	278	980	198	9910	CR
55	D	795	344	13	3	294	214	283	802	177	855	CR
56	D	804	332	13	3	295	215	284	846	160	813	CR
57	D	778	361	11	3	295	215	276	836	208	558	CR
58	D	766	334	8	4	295	215	21	869	169	742	COMP. EX.
59	D	794	330	7	2	294	214	347	962	188	501	COMP. EX.
60	D	786	472	13	4	295	215	289	971	168	859	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. 13 (°C)	Holding time t3 (s)	Type*
61	D	756	347	7	4	295	215	277	857	174	747	GA INV. EX.
62	D	778	220	10	2	295	215	280	845	178	636	EG INV. EX.
63	D	799	304	7	3	295	215	287	888	199	977	GA INV. EX.
64	D	801	233	15	2	294	214	288	860	177	518	CR INV. EX.
65	E	776	328	9	3	298	218	280	905	167	662	CR INV. EX.
66	F	768	335	10	2	411	331	404	986	167	603	GA INV. EX.
67	G	766	285	9	3	420	340	413	991	204	546	GA COMP. EX.
68	H	775	383	12	3	213	133	199	818	189	812	GI INV. EX.
69	I	806	292	10	3	183	103	172	915	190	559	GA COMP. EX.
70	J	768	465	11	3	334	254	320	970	164	534	GA INV. EX.
71	K	794	423	8	3	329	249	317	952	212	852	GA COMP. EX.
72	L	787	397	10	4	342	262	331	994	174	783	GA INV. EX.
73	M	773	409	8	4	336	256	325	943	163	646	GI COMP. EX.
74	N	785	205	10	3	393	313	383	974	176	537	GA INV. EX.
75	O	783	203	14	2	408	328	390	896	209	789	GA COMP. EX.
76	P	779	359	13	3	272	192	261	813	164	618	GA INV. EX.
77	Q	759	297	6	3	260	180	253	957	183	728	GA COMP. EX.

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(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. 13 (°C)	Holding time t3 (s)	Type*	INV. EX.
78	R	766	394	8	3	329	249	322	879	158	860	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 (s)	Type*	
79	S	784	454	13	3	341	261	327	930	189	933	GI	COMP. EX.
80	T	801	331	12	2	341	261	325	901	215	887	GA	INV.EX.
81	U	768	416	6	3	347	267	334	919	185	770	GA	COMP. EX.
82	V	771	480	11	3	336	256	327	824	187	944	GA	INV.EX.
83	W	758	214	12	3	352	272	345	984	164	764	GA	COMP. EX.
84	X	801	499	13	3	352	272	335	934	171	697	CR	INV.EX.
85	Y	790	401	6	3	341	261	326	971	194	829	CR	COMP. EX.
86	Z	791	235	5	3	339	259	333	890	203	868	GA	INV.EX.
87	AA	775	413	10	3	329	249	312	865	174	528	GA	COMP. EX.
88	AB	787	208	6	3	350	270	333	992	193	882	GA	INV.EX.
89	AC	798	455	14	3	345	265	340	942	170	856	GA	INV.EX.
90	AD	777	477	7	4	336	256	330	977	150	508	GA	COMP. EX.
91	AE	802	364	11	3	348	266	335	906	156	648	GA	INV.EX.
92	AF	799	203	12	2	336	258	331	1000	211	653	GA	INV.EX.
93	AG	752	276	5	2	326	246	313	979	168	583	CR	COMP. EX.
94	AH	804	473	14	3	331	251	320	980	185	523	CR	INV.EX.
95	AI	786	370	13	2	345	265	336	846	182	795	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 (s)	Type*	
96	AJ	803	309	10	3	340	260	329	977	220	587	CR	COMP. EX.
97	AK	807	472	14	2	340	260	323	900	152	738	CR	INV. EX.
98	AL	798	340	13	3	340	260	329	834	152	641	CR	INV. EX.
99	AM	755	292	10	3	325	245	313	843	156	644	CR	COMP. EX.
100	AN	705	370	5	3	351	271	340	924	194	645	CR	INV. EX.
101	AO	927	268	9	4	339	259	320	856	181	676	CR	INV. EX.
102	AP	773	51	11	2	346	266	338	878	214	580	CR	INV. EX.
103	AQ	767	864	7	2	318	236	298	1000	211	849	CR	INV. EX.
104	AR	788	208	18	3	326	246	313	942	168	501	CR	INV. EX.
105	AS	804	235	11	4	340	260	333	948	170	967	CR	INV. EX.
106	AT	801	287	14	3	321	241	314	888	164	610	CR	INV. EX.
107	AU	752	236	15	4	349	269	332	963	171	868	CR	INV. EX.
108	AV	771	260	7	2	333	253	332	807	187	716	CR	INV. EX.
109	AW	785	472	13	3	332	252	261	921	206	904	CR	INV. EX.
110	AX	807	354	14	3	345	265	335	325	151	504	CR	INV. EX.
111	AY	758	450	13	3	333	253	326	833	182	679	CR	INV. EX.
112	AZ	803	289	11	3	338	258	321	901	106	597	CR	INV. EX.
113	BA	790	344	15	2	283	203	274	953	394	563	CR	INV. EX.
114	BB	782	379	13	3	295	215	277	977	155	23	CR	INV. EX.
115	BC	797	237	9	4	283	203	273	918	200	9851	CR	INV. EX.
116	BD	799	406	11	4	284	204	265	940	196	667	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 (s)	Type*
117	BE	799	205	8	4	306	226	292	914	168	663	CR
116	BF	769	466	8	2	290	210	277	844	213	732	CR
119	BG	768	326	12	2	287	207	273	812	153	757	CR
120	BH	769	333	7	3	292	212	287	811	150	829	CR
121	BI	794	352	13	2	293	213	281	949	207	878	CR
122	BJ	880	310	19	3	331	251	420	1000	180	800	COMP. EX.
123	BK	758	416	8	3	349	269	330	994	166	744	CR
124	BL	800	400	14	4	293	213	287	843	181	895	CR
125	BM	800	493	6	2	284	204	267	882	207	965	CR
126	BN	797	330	6	3	392	312	381	931	209	677	CR
127	BO	810	366	8	2	403	323	392	842	181	528	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

[0072] 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.

(Microstructure observation)

[0073] 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.

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

[0074] The average proportion of packets having the largest area in prior austenite grains was determined by the method described hereinabove.

(Tensile test)

[0075] 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. TS and EI were thus measured. In the present invention, 980 MPa or higher TS was determined to be acceptable.

(Press formability)

[0076] A hole expansion test was performed in accordance with JIS Z 2256 (2010). The steel sheets obtained were each cut to 100 mm $\times$ 100 mm. A 10 mm diameter hole was punched with a clearance of $12\% \pm 1\%$. While holding the steel sheet on a die having an inner diameter of 75 mm with a blank holder force of 9 tons (88.26 kN), a conical punch with an apex angle of 60° was pushed into the hole to measure the critical hole diameter at the occurrence of cracking. The limiting hole expansion ratio λ (%) was determined from the formula below, and the flangeability was evaluated based on the value of limiting hole expansion ratio.

Limiting hole expansion ratio: λ (%) = $\{(D_f - D_0)/D_0\} \times 100$

wherein D_f is the hole diameter (mm) at the occurrence of cracking and D_0 is the initial hole diameter (mm).

[0077] Based on the tensile strength (TS), the total elongation (EI), and the hole expansion ratio (λ) obtained as described above, $TS \times EI \times \lambda^{0.5}/1000$ was calculated. The steel sheet was evaluated as "excellent in press formability" when the calculated value was 80 or more.

(Flatness in the width direction)

[0078] 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)

[0079] 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

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$vE_{0\%}/vE_{10\%}$ 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.

[0080] The results are described in Tables 6 to 8. As shown in the tables, INVENTIVE EXAMPLES achieved 980 MPa or higher TS, excellent press formability, excellent flatness in the width direction, and excellent working embrittlement resistance. In contrast, COMPARATIVE EXAMPLES were unsatisfactory in one or more of TS, press formability, flatness in the width direction, and working embrittlement resistance.

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[Table 6]

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Ferrite (area%)	Bainitic ferrite (area%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior γ grain size (μm)	TS (MPa)	EI (%)	λ (%)	$\text{TS} \times \text{E-}$ $\text{I} \times \lambda^{0.5} / 1000$	Flatness in width direction	Working embrittlement resistance	
1	A	55	1	36	10	46	50	15	1602	10	59	123	⊙	⊙	INV. EX.
2	B	59	1	33	10	43	53	12	1791	8	59	110	⊙	⊙	INV. EX.
3	B	41	0	49	7	56	47	10	1006	15	42	98	⊙	⊙	INV. EX.
4	B	<u>34</u>	0	60	6	<u>66</u>	50	12	<u>827</u>	18	56	111	⊙	⊙	COMP. EX.
5	B	57	1	46	4	50	60	20	1758	8	42	91	⊙	○	INV. EX.
6	B	52	0	38	6	44	56	<u>24</u>	1480	10	58	113	⊙	<u>×</u>	COMP. EX.
7	B	43	0	50	7	57	50	12	1012	15	50	107	⊙	⊙	INV. EX.
8	B	<u>35</u>	0	58	8	<u>66</u>	58	14	<u>896</u>	17	60	118	⊙	⊙	COMP. EX.
9	B	51	1	39	6	45	55	18	1425	11	57	118	⊙	○	INV. EX.
10	B	54	0	41	8	49	55	<u>21</u>	1540	10	44	102	⊙	<u>×</u>	COMP. EX.
11	B	85	0	11	3	14	50	8	1947	7	40	86	⊙	⊙	INV. EX.
12	B	<u>94</u>	1	1	6	<u>7</u>	52	10	2031	5	50	<u>72</u>	⊙	⊙	COMP. EX.
13	B	51	1	38	9	47	55	10	1471	10	57	111	⊙	⊙	INV. EX.
14	B	57	1	42	7	49	49	10	1687	9	59	117	⊙	⊙	INV. EX.
15	B	59	0	42	8	50	49	14	1752	8	51	100	⊙	⊙	INV. EX.
16	B	54	1	36	9	45	55	13	1545	10	47	106	⊙	⊙	INV. EX.
17	B	84	1	13	7	20	50	14	1973	6	51	85	⊙	⊙	INV. EX.
18	B	<u>92</u>	0	2	2	<u>4</u>	55	9	2294	4	57	<u>69</u>	⊙	⊙	INV. EX.

(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Ferrite (area%)	Bainitic ferrite (area%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior γ grain size (μm)	TS (MPa)	EI (%)	λ (%)	$\text{TS} \times \text{E-}$ $I \times \lambda^{0.5} / 1000$	Flatness in width direction	Working embrittlement resistance	
19	B	49	2	40	7	47	57	14	1390	11	33	88	⊙	⊙	INV. EX.
20	B	52	<u>6</u>	32	10	42	47	10	1522	10	25	<u>76</u>	⊙	⊙	COMP. EX.
21	B	53	1	39	8	47	<u>95</u>	10	1557	9	57	106	$\times$ <u>—</u>	$\times$ <u>—</u>	COMP. EX.
22	B	57	0	33	9	42	<u>78</u>	11	1732	9	49	109	$\times$ <u>—</u>	$\times$ <u>—</u>	COMP. EX.
23	B	56	2	42	6	48	56	12	1621	9	34	85	⊙	⊙	INV. EX.
24	B	52	<u>6</u> <u>—</u>	40	6	46	59	12	1458	10	24	<u>71</u>	⊙	⊙	COMP. EX.
25	B	45	<u>7</u> <u>—</u>	37	5	42	46	8	1128	14	21	<u>72</u>	⊙	⊙	COMP. EX.
26	B	52	0	37	7	44	50	12	1509	10	56	113	⊙	⊙	INV. EX.
27	B	55	1	43	5	48	50	11	1728	9	45	104	⊙	○	INV. EX.
28	B	54	1	34	7	41	48	11	1684	9	54	111	⊙	○	INV. EX.
29	B	53	0	32	9	41	53	14	1221	12	43	96	⊙	⊙	INV. EX.
30	B	59	0	35	7	42	52	14	1477	10	42	96	⊙	⊙	INV. EX.
31	B	52	0	37	9	46	49	9	1518	10	46	103	⊙	○	INV. EX.
32	B	52	1	41	7	48	47	10	1500	10	49	105	⊙	○	INV. EX.
33	B	50	1	37	7	44	46	13	1354	11	43	98	⊙	⊙	INV. EX.
34	B	52	1	37	8	45	58	14	1465	10	54	108	⊙	⊙	INV. EX.
35	B	57	0	36	9	45	56	10	1690	9	54	112	⊙	⊙	INV. EX.
36	B	51	1	39	4	43	53	13	1447	11	56	119	⊙	⊙	INV. EX.

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(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Ferrite (area%)	Bainitic ferrite (area%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior γ grain size (μm)	TS (MPa)	EI (%)	λ (%)	$\text{TS} \times \text{E-}$ $\text{I} \times \lambda^{0.5} / 1000$	Flatness in width direction	Working embrittlement resistance	
37	C	52	<u>5</u>	37	9	46	50	13	1480	10	23	<u>71</u>	⊙	⊙	COMP. EX.
38	D	54	1	36	6	42	<u>93</u>	9	1384	11	48	105	<u>×</u>	<u>×</u>	COMP. EX.
39	D	41	0	54	5	59	58	14	1035	15	56	116	⊙	⊙	INV. EX.
40	D	54	0	44	4	48	48	18	1474	10	42	96	⊙	○	INV. EX.
41	D	42	0	54	4	58	56	10	983	15	42	96	⊙	⊙	INV. EX.
42	D	58	0	37	4	41	57	16	1636	9	58	112	⊙	○	INV. EX.
43	D	84	0	11	2	13	50	14	2138	6	51	92	⊙	⊙	INV. EX.
44	D	56	0	38	10	48	56	12	1564	9	50	100	⊙	⊙	INV. EX.
45	D	54	0	41	3	44	56	13	1486	10	52	107	⊙	⊙	INV. EX.
46	D	88	1	14	3	17	59	12	2120	6	49	89	⊙	⊙	INV. EX.
47	D	49	2	42	4	46	59	14	1201	12	34	84	⊙	⊙	INV. EX.
Underlines indicate being outside the range of the present invention.															

[Table 7]

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Ferrite (area%)	Bainitic ferrite (area%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior γ grain size (μm)	TS (MPa)	EI (%)	λ (%)	TSxEI $\lambda^{0.5}$ /1000	Flatness in width direction	Working embrittlement resistance
48	D	52	1	42	4	46	56	14	1342	11	48	102	⊙	INV. EX.
49	D	57	2	40	8	48	49	14	1582	9	33	82	⊙	INV. EX.
50	D	55	0	37	8	45	47	11	1435	10	48	99	⊙	INV. EX.
51	D	55	1	43	3	46	46	9	1586	10	55	118	⊙	INV. EX.
52	D	50	0	37	5	42	52	13	1046	14	48	101	⊙	INV. EX.
53	D	53	0	35	8	43	52	8	1343	11	53	108	⊙	INV. EX.
54	D	55	1	39	4	43	55	15	1060	13	52	99	⊙	INV. EX.
55	D	58	1	37	9	46	47	9	1627	9	42	95	⊙	INV. EX.
56	D	52	0	38	4	42	49	14	1382	11	55	113	⊙	INV. EX.
57	D	52	1	39	5	44	59	9	1310	11	52	104	⊙	INV. EX.
58	D	45	<u>5</u>	36	7	43	57	8	1054	14	26	<u>75</u>	⊙	COMP. EX.
59	D	49	0	39	6	45	<u>88</u>	9	1205	12	50	102	x	COMP. EX.
60	D	52	0	33	10	43	55	14	1370	11	52	109	⊙	INV. EX.
61	D	57	1	32	9	41	54	13	1586	10	60	123	⊙	INV. EX.
62	D	51	1	36	7	43	50	15	1310	12	42	102	⊙	INV. EX.
63	D	51	1	39	6	45	50	13	1279	12	45	103	⊙	INV. EX.
64	D	52	0	43	3	46	56	15	1357	11	54	110	⊙	INV. EX.
65	E	59	1	33	7	40	59	10	1712	9	56	115	⊙	INV. EX.
66	F	42	0	50	5	55	59	10	1036	14	51	104	⊙	INV. EX.
67	G	<u>33</u>	1	59	5	<u>64</u>	51	11	<u>819</u>	18	57	111	⊙	COMP. EX.

(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Ferrite (area%)	Bainitic ferrite (area%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior γ grain size (μm)	TS (MPa)	EI (%)	λ (%)	TSxEI $\lambda^{0.5}$ /1000	Flatness in width direction	Working embrittlement resistance	
68	H	53	1	40	8	48	54	11	2022	7	56	106	⊙	○	INV.EX.
69	I	51	0	46	3	49	55	12	2035	7	50	101	⊙	⊗	COMP. EX.
70	J	56	0	35	7	42	48	10	1078	11	52	86	⊙	⊙	INV.EX.
71	K	55	1	37	4	41	54	12	<u>820</u>	10	52	<u>59</u>	⊙	⊙	COMP. EX.
72	L	57	2	38	10	48	57	9	1869	8	33	86	⊙	⊙	INV.EX.
73	M	44	<u>6</u>	39	9	48	52	10	1286	12	26	<u>79</u>	⊙	⊙	COMP. EX.
74	N	42	1	50	10	60	55	12	1096	13	57	108	⊙	⊙	INV.EX.
75	O	<u>35</u>	0	61	4	<u>65</u>	51	10	<u>687</u>	22	49	106	⊙	⊙	COMP. EX.
76	P	59	0	38	4	42	59	15	1984	8	54	117	⊙	○	INV.EX.
77	Q	55	1	39	6	45	56	9	1789	9	54	118	⊙	⊗	COMP. EX.
78	R	52	1	36	6	42	52	9	1529	10	41	98	⊙	○	INV.EX.
79	S	53	0	38	6	44	48	14	1533	10	48	106	⊙	⊗	COMP. EX.
80	T	58	0	38	7	45	47	11	1719	9	47	106	⊙	○	INV.EX.
81	U	57	0	32	9	41	57	12	1679	9	45	101	⊙	⊗	COMP. EX.
82	V	40	1	48	7	55	54	9	1066	14	48	103	⊙	⊙	INV.EX.
83	W	<u>30</u>	0	64	6	<u>70</u>	45	15	<u>792</u>	19	51	107	⊙	⊙	COMP. EX.

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(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Ferrite (area%)	Bainitic ferrite (area%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior γ grain size (μm)	TS (MPa)	EI (%)	λ (%)	TSxEI $\lambda^{0.5}$ /1000	Flatness in width direction	Working embrittlement resistance	
84	X	54	0	40	8	48	51	13	1586	9	53	104	⊙	○	INV. EX.
85	Y	51	0	38	6	44	45	14	1361	11	43	98	⊙	× <u>—</u>	COMP. EX.
86	Z	56	1	43	5	48	53	9	1576	10	48	109	⊙	○	INV. EX.
87	AA	57	1	37	10	47	46	9	1713	9	49	108	⊙	× <u>—</u>	COMP. EX.
88	AB	53	0	41	4	45	51	14	1449	10	41	93	⊙	⊙	INV. EX.
89	AC	50	0	44	3	47	55	14	1496	10	53	109	⊙	○	INV. EX.
90	AD	58	1	39	9	48	57	14	1533	10	54	113	⊙	× <u>—</u>	COMP. EX.
91	AE	59	1	38	5	43	55	8	1772	9	45	107	⊙	⊙	INV. EX.
92	AF	54	1	43	3	46	55	13	1816	8	54	107	⊙	○	INV. EX.
93	AG	54	0	38	9	47	54	10	1850	8	59	114	⊙	× <u>—</u>	COMP. EX.
Underlines indicate being outside the range of the present invention.															

[Table 8]

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Ferrite (area%)	Bainitic ferrite (area%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior γ grain size (μm)	TS (MPa)	EI (%)	A (%)	$\text{TS} \times \text{E-}$ $\text{I} \times \lambda^{0.5} / 1000$	Flatness in width direction	Working embrittlement resistance	
94	AH	57	1	37	5	42	51	13	1656	9	51	106	⊙	⊙	INV.EX.
95	AI	53	0	33	10	43	47	14	1696	9	49	107	⊙	○	INV.EX.
96	AJ	54	1	39	4	43	47	9	1742	9	56	117	⊙	× <u>—</u>	COMP. EX.
97	AK	58	1	44	4	48	48	11	1739	8	55	103	⊙	⊙	INV.EX.
98	AL	53	0	42	8	50	55	14	1771	8	57	107	⊙	○	INV.EX.
99	AM	54	0	40	8	48	57	10	1810	8	60	112	⊙	× <u>—</u>	COMP. EX.
100	AN	40	1	48	7	55	50	9	1038	14	47	100	⊙	⊙	INV.EX.
101	AO	52	1	41	4	45	55	16	1466	10	58	112	⊙	○	INV.EX.
102	AP	43	1	47	9	56	59	8	1039	15	48	108	⊙	⊙	INV.EX.
103	AQ	57	1	34	7	41	54	18	1613	9	43	95	⊙	○	INV.EX.
104	AR	84	1	11	6	17	49	10	2081	6	50	88	⊙	⊙	INV.EX.
105	AS	52	1	41	9	50	52	15	1476	10	56	110	⊙	⊙	INV.EX.
106	AT	58	1	37	4	41	56	15	1798	9	47	111	⊙	⊙	INV.EX.
107	AU	84	1	14	5	19	48	11	2071	6	56	93	⊙	⊙	INV.EX.
108	AV	50	2	37	8	45	49	10	1331	11	45	98	⊙	⊙	INV.EX.
109	AW	55	0	40	6	46	51	8	1536	10	31	86	⊙	⊙	INV.EX.
110	AX	50	2	40	9	49	47	9	1444	10	34	84	⊙	⊙	INV.EX.
111	AY	55	1	36	9	45	53	13	1564	9	53	102	⊙	⊙	INV.EX.
112	AZ	54	0	42	4	46	47	13	1682	9	57	114	⊙	○	INV.EX.
113	BA	50	1	40	7	47	51	9	1041	15	49	109	⊙	⊙	INV.EX.
114	BB	56	1	39	7	46	47	8	1603	9	56	108	⊙	○	INV.EX.

(continued)

Nos.	Steels	Tempered martensite (area%)	Retained austenite (vol%)	Ferrite (area%)	Bainitic ferrite (area%)	Total of ferrite and bainitic ferrite (area%)	Proportion of largest packets in prior austenite grains (area%)	Prior γ grain size (μm)	TS (MPa)	EI (%)	A (%)	$\text{TS} \times \text{E-}$ $\text{I} \times \lambda^{0.5} / 1000$	Flatness in width direction	Working embrittlement resistance	
115	BC	52	1	32	9	41	59	15	1378	11	54	111	⊙	⊙	INV. EX.
116	BD	58	1	33	8	41	53	14	1633	9	42	95	⊙	⊙	INV. EX.
117	BE	52	0	37	9	46	52	9	1389	11	52	110	⊙	⊙	INV. EX.
118	BF	51	1	38	6	44	47	12	1257	12	53	110	⊙	⊙	INV. EX.
119	BG	57	1	39	6	45	50	14	1632	9	48	102	⊙	⊙	INV. EX.
120	BH	59	1	40	4	44	54	11	1774	9	40	101	⊙	⊙	INV. EX.
121	BI	54	0	36	4	40	60	8	1403	11	57	117	⊙	⊙	INV. EX.
122	BJ	89	0	9	1	10	<u>88</u>	9	1520	9	45	92	$\times$ <u>—</u>	$\times$ <u>—</u>	COMP. EX.
123	BK	57	1	37	5	42	45	13	1743	9	49	110	⊙	⊙	INV. EX.
124	BL	52	1	42	4	46	54	8	1339	11	47	101	⊙	⊙	INV. EX.
125	BM	51	1	43	4	47	57	8	1255	12	56	113	⊙	⊙	INV. EX.
126	BN	50	1	36	6	42	58	14	1406	11	52	112	⊙	⊙	INV. EX.
127	BO	59	0	34	9	43	46	10	1828	8	51	104	⊙	⊙	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 38% or more and less than 90%,

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

an area fraction of the total of ferrite and bainitic ferrite is 10% or more and 60% or less,

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

an average of 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 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 according to 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 700°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:

the average cooling rate from 750°C to 600°C is less than 20°C/s,

an average cooling rate from (Ms + 50°C) to a quench start temperature T2 is less than 5°C/s 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,

$$Ms = 519 - 474 \times [\% \text{ C}] - 30.4 \times [\% \text{ Mn}] - 12.1 \times [\% \text{ Cr}] - 7.5 \times [\% \text{ Mo}] - 17.7 \times [\% \text{ Ni}] - T1/80 \quad (1)$$

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.

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FIG. 1

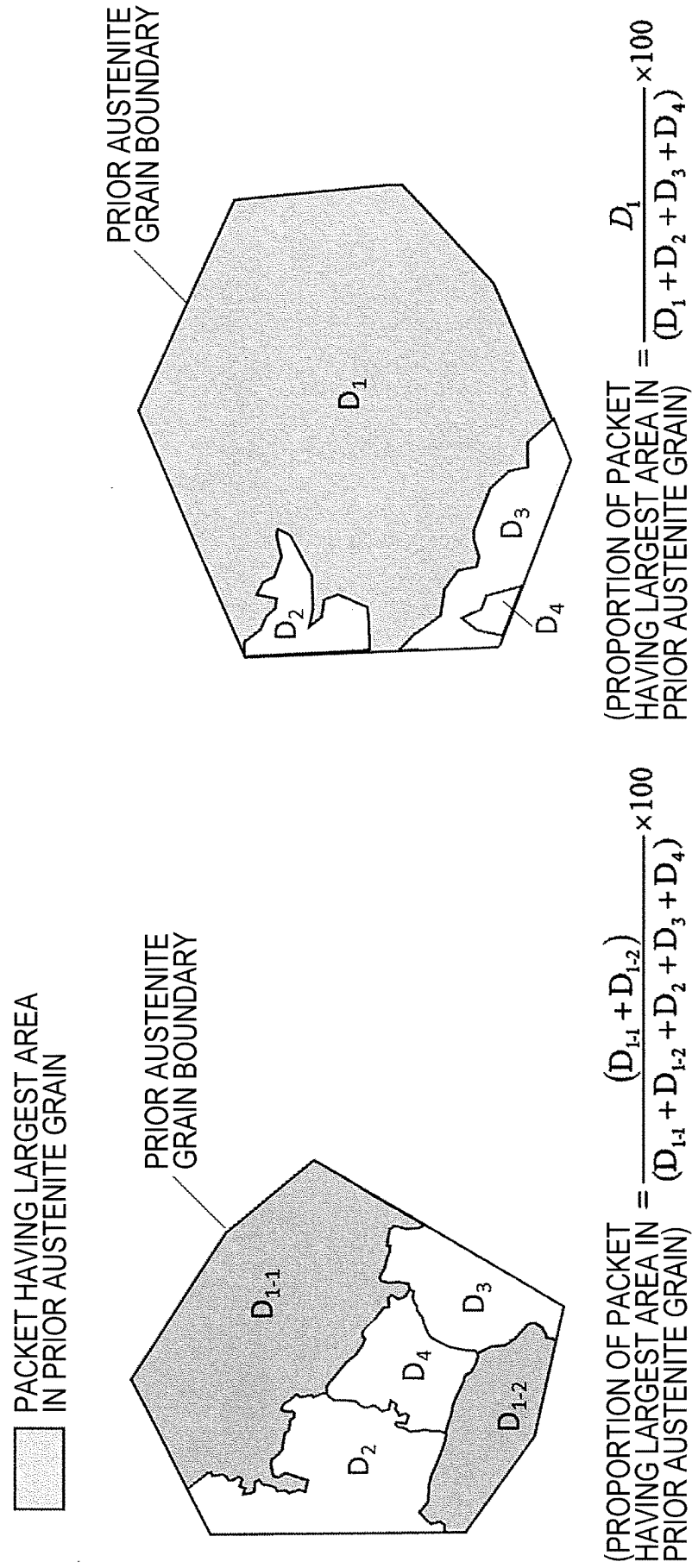
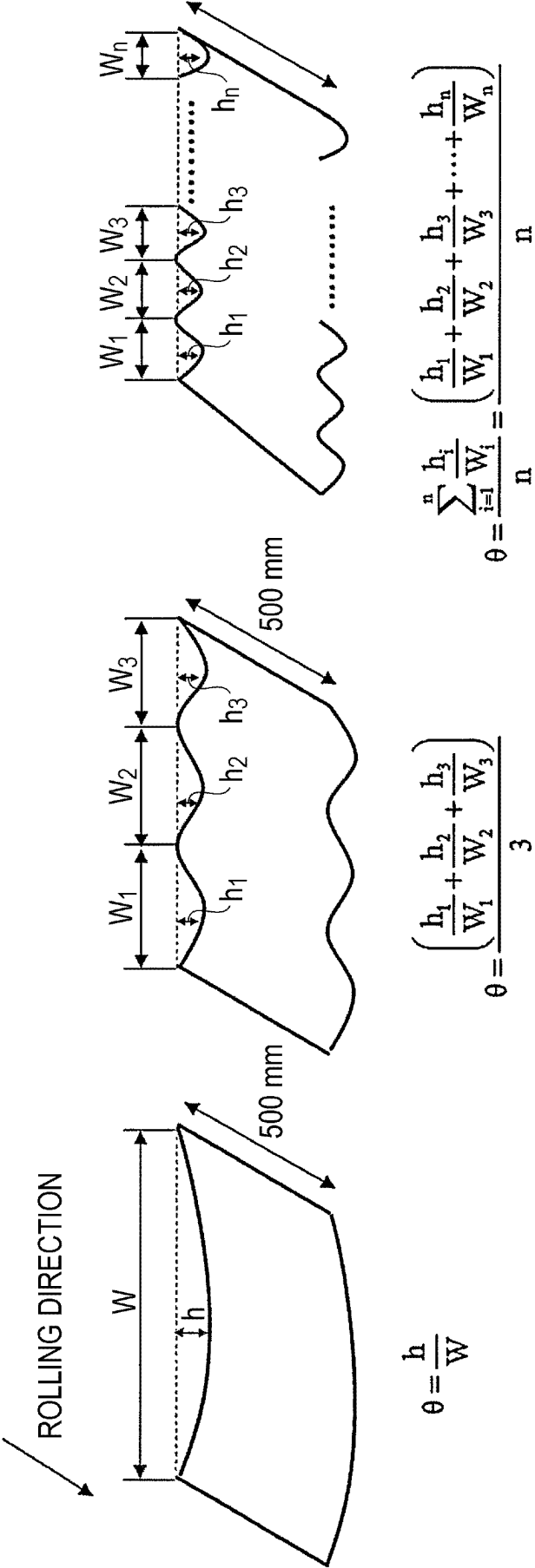


FIG. 2



INTERNATIONAL SEARCH REPORT

International application No.

PCT/JP2023/002915

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; C21D9/46 F; C21D9/46 J; C22C38/00 301T; C22C38/06; C22C38/60

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	JP 2009-287102 A (JFE STEEL CORP) 10 December 2009 (2009-12-10) claims	1-5
A	WO 2014/129379 A1 (KOBE STEEL LTD) 28 August 2014 (2014-08-28) claims	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.

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

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“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

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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/JP2023/002915

Patent document cited in search report			Publication date (day/month/year)	Patent family member(s)	Publication date (day/month/year)
JP 2009-287102 A			10 December 2009	(Family: none)	
WO 2014/129379 A1			28 August 2014	EP 2960353 A1	
				claims	
				US 2015/0368742 A1	
				CN 105074040 A	
				JP 2014-159610 A	
WO 2022/209519 A1			06 October 2022	(Family: none)	

REFERENCES CITED IN THE DESCRIPTION

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Patent documents cited in the description

- JP 4947176 B [0006]
- JP 6777272 B [0006]

Non-patent literature cited in the description

- *Journal of Smart Processing*, 2013, vol. 2 (3), 110-118
[0007]