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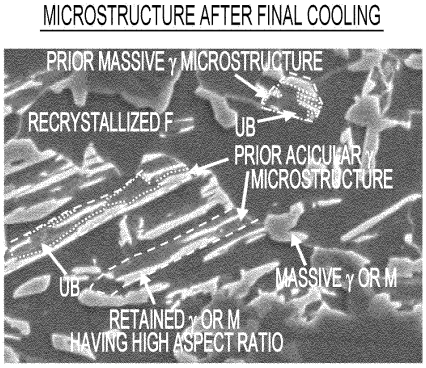
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(54) STEEL SHEET, MEMBER, AND METHOD FOR MANUFACTURING STEEL SHEET AND MEMBER

(57) Provided are a steel sheet having high strength, high ductility, excellent stretch flange formability, and good chemical convertibility; a related member; and methods for manufacturing them.

The steel sheet has a chemical composition including specific amounts of C, Si, Mn, P, S, sol. Al, and N in mass%. The steel sheet has a specific ratio of the total of polygonal ferrite, upper bainite, retained γ, fresh martensite, tempered martensite, and lower bainite, and a specific ratio of a remaining microstructure. The steel sheet has a specific ratio of the number of fresh martensite grains and retained γ grains having an equivalent circular diameter of less than 1.2 um, and has a specific ratio of the number of fresh martensite grains and retained γ grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 um or more.

(a)

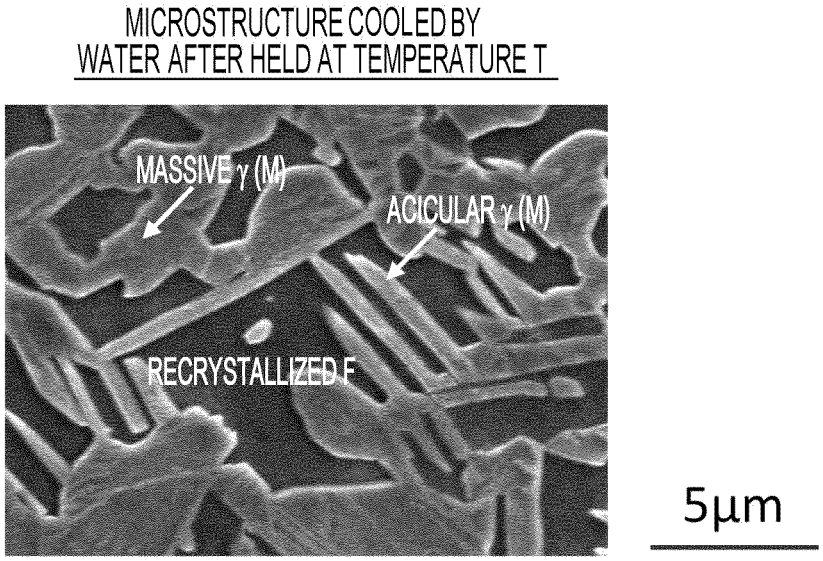


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FIG. 1

(b)


MICROSTRUCTURE COOLED BY
WATER AFTER HELD AT TEMPERATURE T
MASSIVE γ (M)
ACICULAR γ (M)
RECRYSTALLIZED F
5μm

Description

Technical Field

[0001] The present invention relates to a steel sheet that is suitably press formed into complicated shapes through a press-forming process for use in, for example, automobiles and home appliances and has excellent chemical convertibility; to a member obtained using the steel sheet; and to methods for manufacturing them.

Background Art

[0002] Against the backdrop of heightened regulations on CO_2 emissions worldwide, there is a growing demand for weight reduction of automobile bodies by increasing the strength of steel sheets for automobiles. Thus, 590 MPa or higher grade high strength steel sheets are increasingly applied to bodies and seat parts in place of the existing 440 MPa grade cold rolled steel sheets. In general, increasing the strength of steel sheets is accompanied by a decrease in press formability, such as ductility and stretch flangeability, thereby increasing the occurrence frequency of cracking at the time of press forming and lowering the degree of freedom in shape. Thus, the application is limited to parts having simple shapes. In order to apply high strength steel sheets to parts having complicated shapes, it is important to increase the strength of steel sheets while maintaining or enhancing formability.

[0003] Against the above background, techniques for enhancing the ductility of steel sheets have led to the development of TRIP steel in which retained austenite (retained γ) is dispersed in the microstructure of the steel sheets.

[0004] For example, Patent Literature 1 discloses a manufacturing method involving austempering treatment (a treatment in which the steel is cooled from a single-phase annealing temperature or a two-phase annealing temperature to a bainite transformation temperature and is isothermally held to form retained γ while utilizing bainite transformation during the isothermal holding or the cooling). Specifically, a steel sheet including C: 0.10 to 0.45%, Si: 0.5 to 1.8%, and Mn: 0.5 to 3.0% is annealed and is thereafter subjected to aging treatment at a temperature in the range of 350 to 500°C for 1 to 30 minutes to form retained γ. According to the disclosure, a high ductility steel sheet having TS: 80 kgf/mm^2 or more and TS $\times$ EL: 2500 kgf/mm^{2}·% or more can be obtained.

[0005] Patent Literature 2 discloses that a steel sheet containing C: 0.10 to 0.25%, Si: 1.0 to 2.0%, and Mn: 1.5 to 3.0% is annealed, cooled to 450 to 300°C at a rate of 10°C/s or more, and held for 180 to 600 seconds. In this manner, the microstructure is controlled so that the volume fraction of retained γ will be 5% or more and the area fractions of bainitic ferrite and polygonal ferrite will be 60% or more and 20% or less, respectively. According to the disclosure, a steel sheet excellent in both ductility: El and stretch flangeability: λ can be obtained.

[0006] In the above techniques, the steel sheet contains a large amount of silicon in order to promote efficient concentration of carbon into non-transformed γ. On the other hand, steel sheets used for press-formed members are incorporated into structures, such as automobiles, after being painted. Chemical conversion is performed on the steel sheets in order to impart good paintability. In chemical conversion, the presence of an oxide film as a superficial portion of the steel sheet destroys the uniformity of crystal grains that are attached by the chemical conversion and serves as a factor that deteriorates paintability. To address this, a continuous annealing furnace used in the manufacturing of steel sheets usually performs pickling treatment as a pretreatment to improve the chemical convertibility. Depending on the components in the steel sheets, particularly when the steel sheets have a high Si content, the chemical convertibility is significantly deteriorated by a Si-containing surface oxide layer that remains after pickling.

[0007] In order to solve the above problem, for example, Patent Literature 3 discloses that excellent chemical convertibility can be imparted even to a high-Si content steel sheet by steps in which the steel sheet is pickled by being continuously immersed in a mixed acid solution including an oxidative first acid and a non-oxidative second acid, and is repickled by being continuously immersed in an acid solution including a non-oxidative third acid.

Citation List

Patent Literature

[0008]

PTL 1: Japanese Examined Patent Application Publication No. 6-35619
PTL 2: Japanese Patent No. 4411221
PTL 3: Japanese Patent No. 6041079

Summary of Invention

Technical Problem

[0009] Press forming of simple-shaped parts completes with uniform elongation alone, but local elongation is also important in the case of complex-shaped parts. While the conventional TRIP steel described in Patent Literature 1 has excellent El, its stretch flange formability is very low.

[0010] In the technique described in Patent Literature 2, the microstructure is mainly bainitic ferrite and includes a small amount of ferrite. Because of this composition, the steel sheet is excellent in stretch flange formability but is not necessarily high in ductility. Furthermore, the technique is directed to steel sheets having a low yield ratio and is difficult to apply to body frame members and energy absorbing members. Furthermore, Patent Literature 1 and Patent Literature 2 do not take chemical convertibility into consideration. It is expected that chemical convertibility will be low depending on the Si content or the annealing conditions.

[0011] The technique described in Patent Literature 3 can provide a steel sheet having high ductility and excellent stretch flangeability by holding the steel sheet in the course of post-annealing cooling in such a manner that upper bainite transformation is utilized, and by subsequently performing the Q&P treatment and reheating followed by bainite transformation. However, the improvement in local elongation is still insufficient to satisfy both bend formability and bulging formability that are required simultaneously in the formation of hard-to-form parts. Because the steel sheet contains a large amount of silicon to promote the partitioning of carbon from martensite to non-transformed γ in the Q&P treatment, a pickling technique, such as one described in Patent Literature 3, is required. Because of this fact, a pickling facility needs to be added and the running costs are increased. Thus, the establishment of other techniques has been desired.

[0012] As discussed above, the conventional techniques are still insufficient and are incapable of imparting excellent chemical convertibility to steel sheets while ensuring high ductility and excellent stretch flange formability at the same time.

[0013] The present invention has been made to solve the problems discussed above. It is therefore an object of the present invention to provide a steel sheet that has 590 MPa or higher tensile strength and achieves good chemical convertibility as well as high ductility and excellent stretch flange formability; a related member; and methods for manufacturing them.

[0014] Here, 590 MPa or higher tensile strength means that a JIS No. 5 test piece for tensile test has a tensile strength of 590 MPa or more when tested by a tensile test in accordance with the provisions of JIS Z2241 (2011) in the tensile direction perpendicular to the rolling direction at a crosshead speed of 10 mm/min.

[0015] Furthermore, high ductility means that a JIS No. 5 test piece for tensile test satisfies tensile strength (TS) $\times$ total elongation (T. El) $\geq$ 22000 MPa-% or more when tested by a tensile test in accordance with the provisions of JIS Z2241 (2011) in the tensile direction perpendicular to the rolling direction at a crosshead speed of 10 mm/min.

[0016] Furthermore, excellent stretch flange formability means that a steel satisfies (A1) or (A2) below when tested by a hole expansion test in accordance with JFST (The Japan Iron and Steel Federation Standard) 1001.

(A1) The hole expansion ratio λ is 60% or more when the tensile strength is 590 MPa or more and less than 780 MPa.
(A2) The hole expansion ratio λ is 35% or more when the tensile strength is 780 MPa or more.

[0017] Furthermore, good chemical convertibility means that a steel sheet is covered with a chemical conversion coating microstructure on all the faces when the steel sheet is subjected to sulfuric acid electrolytic pickling for 2 seconds at a current density of 20 to 35 A/dm^2, degreasing (treatment temperature: 40°C, treatment time: 120 seconds, spray degreasing), surface conditioning (pH: 9.5, treatment temperature: room temperature, treatment time: 20 seconds), and chemical conversion using a zinc phosphate chemical conversion solution (temperature of the chemical conversion solution: 35°C, treatment time: 120 seconds).

Solution to Problem

[0018] The present inventors carried out extensive studies on approaches to imparting high ductility and excellent stretch flange formability even to a low-Si steel sheet and have obtained the following conclusions. Here, the term low-Si means, although not particularly limited to, that the Si content is less than 1.60 mass%.

[0019] In the austempering treatment, bainite transformation at around 400°C causes partitioning of carbon into non-transformed austenite until T_0 composition is reached in which the free energies of fcc phases and bcc phases are equal to each other, and thereafter the bainite transformation stops. The coarse and thermally unstable non-transformed austenite forms a hard martensite microstructure or mechanically unstable retained γ at the time of final cooling, with the result that stretch flangeability is deteriorated. Thus, it is generally difficult to satisfy ductility and stretch flangeability at the same time by the austempering treatment.

[0020] As a result of extensive studies on the heating process before annealing, the present inventors have identified

the cold rolling conditions, the components in the steel, and the heating conditions that allow a soft ferrite microstructure and, adjacent thereto, acicular austenite to be formed by recrystallization in the heating process. The present inventors have found that this acicular austenite contributes to the partitioning of carbon and the formation of retained austenite in microstructure formation during the cooling process.

[0021] In view of the fact that the To composition expands to a higher carbon region as the bainite transformation temperature is lowered, the present inventors further investigated austempering treatment in two stages, on high-temperature side and on low-temperature side, and have found that by such a treatment, carbon partitioning is further promoted and uniform elongation can be enhanced. Furthermore, the present inventors have found that bainite transformation proceeds to a later stage without stopping, and consequently the amount and the size of coarse and hard fresh martensite formed in the final cooling are reduced. Furthermore, bainite transformation from acicular austenite provides retained austenite having a high aspect ratio and high working stability. The present inventors have found that the above configuration lessens the stress concentration at the time of press forming and eliminates or reduces the formation of voids, thereby enhancing local elongation. Such acicular austenite does not deteriorate flangeability even when it is transformed into a hard fresh martensite microstructure during the final cooling.

[0022] Steel sheets manufactured based on the above guidelines can attain enhancements in uniform elongation and local elongation at the same time even when the alloy design involves a reduced amount of silicon. The low-Si design eliminates the need of costly pickling treatment required to impart chemical convertibility and significantly saves the process costs. Incidentally, the chemical convertibility described here is a characteristic with which the steel sheet after a general pickling process can be treated to attain a coating weight and uniformity offering satisfactory paintability. The general pickling process is, for example, sulfuric acid pickling, but the pickling process is not limited.

[0023] The present inventors have found that the configuration described above allows even a low-Si steel sheet to exhibit excellent ductility and stretch flangeability. The findings are based on the outlines below.

[0024]

(i) In the cold rolling step, steel is cold rolled with a cold rolling reduction ratio (rolling reduction ratio) in the first pass of 5% or more and less than 25% so as to suppress the development of shear texture. A cold rolled steel sheet is thus produced in which the preferred rolling orientation (texture) and the rotated cube orientation are developed (a cold rolled steel sheet in which the area fraction of the total of microstructures having {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation is 35% or more and 75% or less relative to all the bcc phase microstructures).

(ii) Before soaking and holding in the annealing step, the temperature is increased from 500°C to Ac1 at a heating rate (average heating rate) of 15°C/sec or less to induce sufficient recrystallization of microstructures in the cold rolled steel sheet that have been cold rolled with a rolling reduction ratio of 30% or more, thus developing a recrystallized texture.

(iii) Subsequently, austenite (γ) transforms at temperatures of Ac1 and above forms from recrystallized bcc phase grain boundaries or from residual carbides as nuclei. The transformed austenite has a specific crystallographic orientation relationship with the surrounding bcc phases. Thus, the degree of interfacial matching is high, and the grain growth accompanied by interfacial migration is retarded. Part of the interfaces preferentially migrate in order to approach an equilibrium state, and acicular austenite (acicular γ) is thus formed. In order to make use of this acicular γ, annealing is performed at a two-phase annealing temperature. Annealing is performed at an annealing temperature T that satisfies $(T - Ac1)/(Ac3 - Ac1) < 1.0$.

(iv) During cooling after soaking and holding in the annealing step, the steel is isothermally held (first holding) in a range of temperatures of 400 to 550°C. In this manner, acicular γ is transformed into upper bainite containing little precipitates, and non-transformed austenite (non-transformed γ) having a high solute C content is formed.

By performing austempering treatment in two stages, on high-temperature side and on low-temperature side, carbon is partitioned efficiently without stopping of bainite transformation and acicular austenite is caused to occur in the microstructure before cooling. In this manner, retained austenite having a high aspect ratio and high working stability can be formed after the bainite transformation. As a result, it is possible to manufacture a steel sheet that achieves excellent uniform elongation and local elongation at the same time. Such steel sheets are not cracked even when worked by press forming that involves both bulging and stretch flanging and can be applied to the production of complicated shapes.

[0025] As described above, excellent uniform elongation and local elongation can be obtained at the same time by making use of acicular austenite formed in the heating process and by utilizing two-stage bainite transformation on high-temperature side and on low-temperature side. As a result, high ductility and excellent stretch flange formability can be imparted to even a steel sheet having a reduced amount of silicon. Furthermore, the chemical convertibility of the steel sheet can also be improved.

[0026] The present invention has been made based on the above knowledge. Specifically, the present invention

provides the following:

[1] A steel sheet having a chemical composition including, in mass%:

C: 0.06 to 0.24%,
Si: 0.4% or more and less than 1.60%,
Mn: 1.5 to 3.2%,
P: 0.02% or less,
S: 0.01% or less,
sol. Al: less than 1.0%, and
N: less than 0.015%,
the chemical composition satisfying formula (1) below,
the balance being Fe and incidental impurities,
the steel sheet including a microstructure in which:

the area fraction of polygonal ferrite is 20% or more and 85% or less,
the area fraction of upper bainite is 9% or more and 45% or less,
the volume fraction of retained austenite is 3% or more and 15% or less,
the area fraction of fresh martensite is 3% or more and 15% or less,
the total of the area fractions of tempered martensite and lower bainite is 50% or less (including 0%), and
the area fraction of a remaining microstructure is 5% or less,
the microstructure being such that:

the ratio of the total number of fresh martensite grains and retained austenite grains having an equivalent circular diameter of less than 1.2 um is 50% or more relative to the number of all fresh martensite grains and all retained austenite grains, and
the ratio of the total number of fresh martensite grains and retained austenite grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 um or more is 40% or more relative to the number of fresh martensite grains and retained austenite grains having an equivalent circular diameter of 1.2 um or more,

$$Si/Mn < 0.50 \cdots \quad \text{Formula (1)}$$

wherein in formula (1), Si and Mn indicate the Si content (mass%) and the Mn content (mass%), respectively.

[2] The steel sheet described in [1], wherein
the chemical composition further includes, in mass%, one, or two or more selected from:

Nb: 0.2% or less,
Ti: 0.2% or less,
V: 0.2% or less,
B: 0.01% or less,
Cu: 0.2% or less,
Ni: 0.2% or less,
Cr: 0.4% or less, and
Mo: 0.15% or less.

[3] The steel sheet described in [1] or [2], wherein
the chemical composition further includes, in mass%, one, or two or more selected from:

Mg: 0.0050% or less,
Ca: 0.0050% or less,
Sn: 0.10% or less,
Sb: 0.10% or less, and
REM: 0.0050% or less.

[4] A member obtained using the steel sheet described in any of [1] to [3].

[5] A method for manufacturing a steel sheet, including, after hot rolling and pickling are performed on a steel slab having the chemical composition described in any of [1] to [3], a cold rolling step of performing a cold rolling treatment on the hot rolled steel sheet to produce a cold rolled steel sheet, and

an annealing step of performing an annealing treatment on the cold rolled steel sheet to produce a steel sheet, the cold rolling step being such that the cold rolled steel sheet is obtained by performing the cold rolling treatment in such a manner that:

the cumulative cold rolling reduction ratio is 30 to 85%, and
the rolling reduction ratio in a first pass is 5% or more and less than 25%, thereby controlling the area fraction of the total of microstructures having {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation to 35% or more and 75% or less relative to all bcc phase microstructures,
the annealing step being such that the annealing treatment includes:

heating the cold rolled steel sheet at an average heating rate of 0.5 to 15°C/sec in a range of temperatures of 500°C or above and Ac1 or below, to an annealing temperature T being 840°C or below and satisfying $0.5 \leq (T - Ac1)/(Ac3 - Ac1) < 1.0$;
after the heating, soaking and holding the steel sheet at the annealing temperature T in a furnace atmosphere having a dew point Td of -50°C or above and -30°C or below, thereby producing a steel sheet having a number density of acicular austenite microstructures of 5 microstructures/1000 μm^2 or more;
subsequently performing first cooling of cooling the steel sheet at an average cooling rate of 6.0°C/sec or more in a range of temperatures of 750 to 550°C, to a first cooling stop temperature Tc1 of 550°C or below and 400°C or above; after the first cooling, subjecting the steel sheet to first holding at the first cooling stop temperature Tc1 for 25 seconds or more;
after the first holding, performing second cooling of cooling the steel sheet to a second cooling stop temperature Tc2 that is equal to or lower than the first cooling stop temperature Tc1 and is 450°C or below and 300°C or above; subjecting the steel sheet to second holding at the second cooling stop temperature Tc2 for 20 to 3000 seconds; and after the second holding, performing third cooling of cooling the steel sheet.

[6] A method for manufacturing a member, including a step of subjecting the steel sheet described in any of [1] to [3] to at least one working of forming and joining to produce a member.

Advantageous Effects of Invention

[0027] According to the present invention, a steel sheet is provided that has 590 MPa or higher tensile strength and achieves high ductility, excellent stretch flange formability, and good chemical convertibility. A related member, and methods for manufacturing them are also provided.
The steel sheet of the present invention is suitably used in press forming of complicated shapes produced in the press forming process for use in, for example, automobiles and home appliances.

Brief Description of Drawings

[0028]

[Fig. 1] Fig. 1 is a set of views illustrating SEM images of a microstructure after final cooling (after third cooling in the annealing step) in the present invention and of a microstructure of acicular austenite (acicular γ) observed in a microstructure cooled by water after held at a temperature T.
[Fig. 2] Fig. 2 is a schematic view of acicular austenite (acicular γ) and explains the definition of the aspect ratio of acicular γ.

Description of Embodiments

[0029] Hereinafter, the present invention will be described in detail. The present invention is not limited to the embodiments discussed below.
[0030] A steel sheet of the present invention has a chemical composition including, in mass%, C: 0.06 to 0.24%, Si:

0.4% or more and less than 1.60%, Mn: 1.5 to 3.2%, P: 0.02% or less, S: 0.01% or less, sol. Al: less than 1.0%, and N: less than 0.015%, the chemical composition satisfying formula (1) below, the balance being Fe and incidental impurities. The steel sheet includes a microstructure in which the area fraction of polygonal ferrite is 20% or more and 85% or less, the area fraction of upper bainite is 9% or more and 45% or less, the volume fraction of retained austenite is 3% or more and 15% or less, the area fraction of fresh martensite is 3% or more and 15% or less, the total of the area fractions of tempered martensite and lower bainite is 50% or less (including 0%), and the area fraction of a remaining microstructure is 5% or less. The microstructure is such that the ratio of the total number of fresh martensite grains and retained austenite grains having an equivalent circular diameter of less than 1.2 um is 50% or more relative to the number of all fresh martensite grains and all retained austenite grains, and the ratio of the total number of fresh martensite grains and retained austenite grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 um or more is 40% or more relative to the number of fresh martensite grains and retained austenite grains having an equivalent circular diameter of 1.2 um or more.

$$Si/Mn < 0.50 \quad \cdots \text{Formula (1)}$$

[0031] In formula (1), Si and Mn indicate the Si content (mass%) and the Mn content (mass%), respectively.
The steel sheet of the present invention will be described below in the order of its chemical composition and its steel microstructure.

[0032] The steel sheet of the present invention includes the components described below. In the following description, the unit "%" for the contents of components means "mass%".

C: 0.06 to 0.24%

[0033] Carbon is added in order to control the hardenability of the steel sheet, the strength of martensite, and the volume fraction of retained γ to desired ranges. If the C content is less than 0.06%, the strength of the steel sheet and the ductility of the steel sheet cannot be sufficiently ensured. Thus, the C content is limited to 0.06% or more. The C content is preferably 0.08% or more, and more preferably 0.10% or more. If the C content exceeds 0.24%, the toughness of welds is deteriorated. Furthermore, the desired area fraction of fresh martensite cannot be obtained if the C content exceeds 0.24%. Thus, the C content is limited to 0.24% or less. In order to enhance ductility and the toughness of spot welds, the C content is preferably 0.21% or less. In order to further improve the toughness of spot welds, the C content is more preferably 0.20% or less.

Si: 0.4% or more and less than 1.60%

[0034] Silicon is added in order to increase the strength of ferrite microstructure and in order to effectively stabilize retained γ and thereby enhance ductility by suppressing the formation of carbides in martensite and bainite. From these points of view, the Si content is limited to 0.4% or more. In order to enhance ductility, the Si content is preferably 0.5% or more. The Si content is more preferably 0.6% or more. If the Si content is 1.60% or more, chemical convertibility is significantly deteriorated. Thus, the Si content is limited to less than 1.60%. The Si content is preferably 1.30% or less, and more preferably 1.20% or less. The Si content is still more preferably less than 1.0%.

Mn: 1.5 to 3.2%

[0035] Manganese ensures predetermined hardenability, suppresses ferrite transformation, and ensures the desired area fraction of tempered martensite and/or bainite to ensure strength. Furthermore, manganese is concentrated into γ during ferrite/γ two-phase annealing and lowers the Ms temperature of non-transformed γ to stabilize retained γ and thereby to improve ductility. Furthermore, similarly to silicon, manganese suppresses the formation of carbides in bainite and enhances ductility. Furthermore, manganese increases the volume fraction of retained γ to enhance ductility. From these points of view, manganese is an important element in the present invention. In order to obtain these effects, the Mn content is limited to 1.5% or more. In order to enhance hardenability, the Mn content is preferably 1.7% or more. The Mn content is more preferably 1.9% or more. If, on the other hand, the Mn content exceeds 3.2%, bainite transformation is significantly retarded to make it difficult to ensure high ductility. Furthermore, more than 3.2% manganese makes it difficult to suppress the formation of massive coarse γ and deteriorates stretch flange formability. Thus, the Mn content is limited to 3.2% or less. In order to ensure high ductility by promoting bainite transformation, the Mn content is preferably 3.0% or less, and more preferably 2.8% or less.

$$Si/Mn < 0.50 \cdots \text{Formula (1)}$$

[0036] Silicon oxide is a surface oxide on the steel sheet that significantly deteriorates chemical convertibility. Thus, Si/Mn is limited to less than 0.50 in order to form Mn-containing oxide that is readily soluble in an acid solution. That is, in the present invention, formula (1) is limited to Si/Mn < 0.50. In formula (1), Si and Mn indicate the Si content (mass%) and the Mn content (mass%), respectively. When the formula is satisfied, chemical convertibility can be imparted in the dew point range of - 50°C or above and -30°C or below. Si/Mn is preferably 0.40 or less, and more preferably 0.35 or less.

P: 0.02% or less

[0037] Phosphorus is an element that strengthens steel, but much phosphorus deteriorates spot weldability. Thus, the P content is limited to 0.02% or less. In order to improve spot weldability, the P content is preferably 0.01% or less. The P content may be nil. From the point of view of manufacturing cost, the P content is preferably 0.001% or more.

S: 0.01% or less

[0038] Sulfur is an element that is effective in improving scale exfoliation in hot rolling and effective in suppressing nitridation during annealing, but sulfur deteriorates spot weldability and local elongation. To eliminate or reduce the deterioration, the S content is limited to 0.01% or less. In the present invention, the contents of C, Si, and Mn are high and spot weldability tends to be deteriorated. In order to improve spot weldability, the S content is preferably 0.0020% or less, and more preferably less than 0.0010%. The S content may be nil. From the point of view of manufacturing cost, the S content is preferably 0.0001% or more.

sol. Al: less than 1.0%

[0039] Aluminum is added for the purpose of deoxidization or for the purpose of stabilizing retained γ as a substitute for silicon. The lower limit of the sol. Al content is not particularly limited. For stable deoxidization, the sol. Al content is preferably 0.01% or more. On the other hand, 1.0% or more sol. Al significantly lowers the strength of the base material and also deteriorates chemical convertibility. Furthermore, large amounts of aluminum oxides are formed during steel-making to cause a significant decrease in bendability. Thus, the sol. Al content is limited to less than 1.0%. In order to obtain high strength, the sol. Al content is preferably less than 0.50%, and more preferably 0.10% or less.

N: less than 0.015%

[0040] Nitrogen is an element that forms nitrides, such as BN, AlN, and TiN, in steel. This element lowers the hot ductility of steel and lowers the surface quality. Furthermore, in B-containing steel, nitrogen has a harmful effect in eliminating the effect of boron through the formation of BN. The surface quality is significantly deteriorated if the N content is 0.015% or more. Thus, the N content is limited to less than 0.015%. The N content may be nil. From the point of view of manufacturing cost, the N content is preferably 0.0001% or more.

[0041] The balance after the above components is Fe and incidental impurities. The steel sheet of the present invention preferably has a chemical composition that contains the basic components described above, with the balance consisting of Fe and incidental impurities.

[0042] In addition to the above components, the chemical composition of the steel sheet of the present invention may appropriately include, in place of part of Fe and incidental impurities, one or two of optional elements selected from the following (A) and (B):

(A) in mass%, one, or two or more selected from Nb: 0.2% or less, Ti: 0.2% or less, V: 0.2% or less, B: 0.01% or less, Cu: 0.2% or less, Ni: 0.2% or less, Cr: 0.4% or less, and Mo: 0.15% or less;
(B) in mass%, one, or two or more selected from Mg: 0.0050% or less, Ca: 0.0050% or less, Sn: 0.10% or less, Sb: 0.10% or less, and REM: 0.0050% or less.

Nb: 0.2% or less

[0043] Niobium is preferably added in order to reduce the size of the microstructure and enhance the defect resisting characteristics of spot welds. Furthermore, niobium may be added to produce an effect of reducing the size of the steel microstructure and increasing the strength, an effect of promoting bainite transformation through the grain size reduction, an effect of improving bendability, and an effect of enhancing delayed fracture resistance. In order to obtain these effects,

the Nb content is preferably 0.002% or more, but the lower limit is not particularly limited. The Nb content is more preferably 0.004% or more, and still more preferably 0.010% or more. However, adding much niobium results in excessive precipitation strengthening and low ductility. Furthermore, the rolling load is increased and castability is deteriorated. Thus, when niobium is added, the Nb content is limited to 0.2% or less. The Nb content is preferably 0.1% or less, more preferably 0.05% or less, and still more preferably 0.03% or less.

Ti: 0.2% or less

[0044] Titanium is preferably added in order to reduce the size of the microstructure and enhance the defect resisting characteristics of spot welds. Furthermore, titanium fixes nitrogen in steel as TiN to produce an effect of enhancing hot ductility and an effect of allowing boron to produce its effect of enhancing hardenability. In order to obtain these effects, the Ti content is preferably 0.002% or more, but the lower limit is not particularly limited. In order to fix nitrogen sufficiently, the Ti content is more preferably 0.008% or more. The Ti content is still more preferably 0.010% or more. On the other hand, more than 0.2% titanium causes an increase in rolling load and a decrease in ductility by an increased amount of precipitation strengthening. Thus, when titanium is added, the Ti content is limited to 0.2% or less. The Ti content is preferably 0.1% or less, and more preferably 0.05% or less. In order to ensure high ductility, the Ti content is still more preferably 0.03% or less.

V: 0.2% or less

[0045] Vanadium may be added to produce an effect of enhancing the hardenability of steel, an effect of suppressing the formation of carbides in martensite and upper/lower bainite, an effect of reducing the size of the microstructure, and an effect of improving delayed fracture resistance through the precipitation of carbide. In order to obtain these effects, the V content is preferably 0.003% or more, but the lower limit is not particularly limited. The V content is more preferably 0.005% or more, and still more preferably 0.010% or more. On the other hand, much vanadium significantly deteriorates castability. Thus, when vanadium is added, the V content is limited to 0.2% or less. The V content is preferably 0.1% or less. The V content is more preferably 0.05% or less, and still more preferably 0.03% or less.

B: 0.01% or less

[0046] Boron advantageously facilitates the formation of a predetermined area fraction of tempered martensite and/or bainite. Furthermore, residual solute boron enhances delayed fracture resistance. In order to obtain these effects of boron, the B content is preferably 0.0002% or more. The B content is more preferably 0.0005% or more. The B content is still more preferably 0.0010% or more. If, on the other hand, the B content exceeds 0.01%, the effects are saturated, and further hot ductility is significantly lowered to invite surface defects. Thus, when boron is added, the B content is limited to 0.01% or less. The B content is preferably 0.0050% or less. The B content is more preferably 0.0030% or less.

Cu: 0.2% or less

[0047] Copper enhances the corrosion resistance in automobile use environments. Furthermore, corrosion products of copper cover the surface of the steel sheet and effectively suppress penetration of hydrogen into the steel sheet. Copper is an element that is mixed when scraps are used as raw materials. By accepting copper contamination, recycled materials can be used as raw materials and thereby manufacturing costs can be reduced. From these points of view and further from the point of view of enhancing delayed fracture resistance, the Cu content is preferably 0.05% or more, but the lower limit is not particularly limited. The Cu content is more preferably 0.10% or more. On the other hand, too much copper invites surface defects. Thus, when copper is added, the Cu content is limited to 0.2% or less.

Ni: 0.2% or less

[0048] Similarly to copper, nickel is an element that acts to enhance corrosion resistance. Furthermore, nickel also acts to eliminate or reduce the occurrence of surface defects that tend to occur when the steel contains copper. In order to obtain these effects, the Ni content is preferably 0.01% or more, but the lower limit is not particularly limited. The Ni content is more preferably 0.04% or more, and still more preferably 0.06% or more. On the other hand, adding too much nickel can instead cause surface defects because scales are formed nonuniformly in a heating furnace, and also increases the cost. Thus, when nickel is added, the Ni content is limited to 0.2% or less.

Cr: 0.4% or less

[0049] Chromium may be added to produce an effect of enhancing the hardenability of steel and an effect of suppressing the formation of carbides in martensite and upper/lower bainite. In order to obtain these effects, the Cr content is preferably 0.01% or more, but the lower limit is not particularly limited. The Cr content is more preferably 0.03% or more, and still more preferably 0.06% or more. On the other hand, too much chromium deteriorates pitting corrosion resistance. Thus, when chromium is added, the Cr content is limited to 0.4% or less.

Mo: 0.15% or less

[0050] Molybdenum may be added to produce an effect of enhancing the hardenability of steel and an effect of suppressing the formation of carbides in martensite and upper/lower bainite. In order to obtain these effects, the Mo content is preferably 0.01% or more. The Mo content is more preferably 0.03% or more, and still more preferably 0.06% or more. On the other hand, molybdenum significantly deteriorates the chemical convertibility of the cold rolled steel sheet. Thus, when molybdenum is added, the Mo content is limited to 0.15% or less.

Mg: 0.0050% or less

[0051] Magnesium fixes oxygen as MgO and contributes to improvement in delayed fracture resistance. Thus, the Mg content is preferably 0.0002% or more. The Mg content is more preferably 0.0004% or more, and still more preferably 0.0006% or more. On the other hand, much magnesium deteriorates surface quality and bendability. Thus, when magnesium is added, the Mg content is limited to 0.0050% or less. The Mg content is preferably 0.0025% or less, and more preferably 0.0010% or less.

Ca: 0.0050% or less

[0052] Calcium fixes sulfur as CaS and contributes to improvements in bendability and delayed fracture resistance. Thus, the Ca content is preferably 0.0002% or more. The Ca content is more preferably 0.0005% or more, and still more preferably 0.0010% or more. On the other hand, much calcium deteriorates surface quality and bendability. Thus, when calcium is added, the Ca content is limited to 0.0050% or less. The Ca content is preferably 0.0035% or less, and more preferably 0.0020% or less.

Sn: 0.10% or less

[0053] Tin suppresses oxidation and nitridation of a superficial portion of the steel sheet and thereby eliminates or reduces the loss of the C and B contents in the superficial portion. Furthermore, the elimination or reduction of the loss of the C and B contents leads to suppressed formation of ferrite in the superficial portion of the steel sheet, thus increasing strength and improving fatigue resistance. From these points of view, the Sn content is preferably 0.002% or more. The Sn content is more preferably 0.004% or more, and still more preferably 0.006% or more. The Sn content is further preferably 0.008% or more, and still further preferably 0.01% or more.
If, on the other hand, the Sn content exceeds 0.10%, castability is deteriorated. Furthermore, tin is segregated at prior γ grain boundaries to deteriorate delayed fracture resistance. Thus, when tin is added, the Sn content is limited to 0.10% or less. The Sn content is preferably 0.04% or less, and more preferably 0.03% or less.

Sb: 0.10% or less

[0054] Antimony suppresses oxidation and nitridation of a superficial portion of the steel sheet and thereby eliminates or reduces the loss of the C and B contents in the superficial portion. Furthermore, the elimination or reduction of the loss of the C and B contents leads to suppressed formation of ferrite in the superficial portion of the steel sheet, thus increasing strength and improving fatigue resistance. From these points of view, the Sb content is preferably 0.002% or more. The Sb content is more preferably 0.004% or more, and still more preferably 0.006% or more. If, on the other hand, the Sb content exceeds 0.10%, castability is deteriorated and segregation occurs at prior γ grain boundaries to deteriorate delayed fracture resistance. Thus, when antimony is added, the Sb content is limited to 0.10% or less. The Sb content is preferably 0.04% or less, and more preferably 0.03% or less.

REM: 0.0050% or less

[0055] Rare earth metals are elements that spheroidize the shape of sulfides and thereby eliminate or reduce adverse

effects of sulfides on stretch flange formability, thus improving stretch flange formability. In order to obtain this effect, the REM content is preferably 0.0005% or more. The REM content is more preferably 0.0010% or more, and still more preferably 0.0020% or more.

If, on the other hand, the REM content exceeds 0.0050%, the effect of improving stretch flange formability is saturated. Thus, when rare earth metals are added, the REM content is limited to 0.0050% or less.

In the present invention, the rare earth metals indicate scandium (Sc) with atomic number 21, yttrium (Y) with atomic number 39, and lanthanide elements from lanthanum (La) with atomic number 57 to lutetium (Lu) with atomic number 71. The REM concentration in the present invention is the total content of one, or two or more elements selected from the above rare earth metals.

[0056] When the content of any of the above optional components is below the lower limit, the optional element present below the lower limit does not impair the advantageous effects of the present invention. Thus, such an optional element below the lower limit content is regarded as an incidental impurity.

[0057] Next, the steel microstructure of the steel sheet of the present invention will be described.

Area fraction of polygonal ferrite: 20% or more and 85% or less

[0058] In order to ensure high ductility, the area fraction of polygonal ferrite is limited to 20% or more. The polygonal ferrite is preferably 25% or more, and more preferably 30% or more. The polygonal ferrite is preferably 35% or more, and more preferably 40% or more.

On the other hand, the area fraction of polygonal ferrite is limited to 85% or less in order to obtain predetermined strength. The polygonal ferrite is more preferably 82% or less. The polygonal ferrite is preferably 80% or less, and more preferably 78% or less.

Area fraction of upper bainite: 9% or more and 45% or less

[0059] Upper bainite is bainite involving a low level of carbide precipitation. Upper bainite partitions carbon into surrounding non-transformed γ and thus can be used to form retained γ with high working stability. In addition, upper bainite has hardness intermediate between those of ferrite and martensite, and the presence of such a microstructure having intermediate hardness enhances local elongation. Thus, 9% or more upper bainite is required at a strength level where the tensile strength (TS) is 590 MPa or more. The area fraction of upper bainite is therefore limited to 9% or more. The area fraction of upper bainite is preferably 12.0% or more, and more preferably 15.0% or more. In the present invention, on the other hand, the steel is soaked and held at a two-phase temperature and a ferrite microstructure is formed in a large amount. In view of this fact, the area fraction of upper bainite is limited to 45% or less in order to reduce the decrease in strength. The upper bainite is preferably 38% or less, and more preferably 30% or less.

Volume fraction of retained austenite (retained γ): 3% or more and 15% or less

[0060] In order to ensure high ductility, the volume fraction of retained γ is limited to 3% or more relative to the whole of the steel microstructure. The volume fraction of retained γ (the amount of retained γ) is preferably 3.0% or more, more preferably 5% or more, and still more preferably 7% or more. This amount of retained γ includes the amounts of retained γ generated adjacent to upper bainite and retained γ generated adjacent to martensite and lower bainite. If the amount of retained γ is excessively large, strength is lowered and stretch flange formability is significantly lowered. Thus, the volume fraction of retained γ is limited to 15% or less. The volume fraction of retained γ is preferably 13% or less. Incidentally, the "volume fraction" may be regarded as the "area fraction".

Area fraction of fresh martensite: 3% or more and 15% or less

[0061] Fresh martensite is a microstructure that lowers local elongation, but can offer enhanced strength when formed within a range not detrimental to bendability and flangeability. From this point of view, the lower limit and the upper limit of the area fraction of fresh martensite are limited to 3% and 15%, respectively.

Total of area fractions of tempered martensite and lower bainite: 50% or less (including 0%)

[0062] In the present invention, tempered martensite and lower bainite are formed when the steel is held on the low-temperature side in the two-stage austempering.

While upper bainite involves a low level of carbide precipitation, tempered martensite and lower bainite have carbides precipitated in their microstructures. Thus, the amount of carbon partitioned to non-transformed γ is reduced. Tempered martensite and lower bainite, however, broaden the T_0 composition at low temperatures to bring about enrichment of

carbon to non-transformed γ or further reduce the amount of fresh martensite occurring during the final cooling. It is therefore necessary to control these microstructures to obtain retained γ having high working stability.

Depending on the chemical composition of the steel or the austempering temperature, the holding on the low-temperature side in the two-stage austempering takes place at or below the Ms temperature, and part of non-transformed γ undergoes martensite transformation and is tempered by the subsequent holding. If the total of the area fractions of tempered martensite and lower bainite exceeds 50%, the precipitation of carbides is so promoted that the required amount of retained γ cannot be obtained and further the strength is excessively increased to make it impossible to obtain desired ductility. Thus, in the present invention, the total of the area fractions of tempered martensite and lower bainite is limited to 50% or less. The total of these area fractions is preferably 40% or less, and more preferably 35% or less. The total of these area fractions is still more preferably 30% or less, and further preferably 25% or less. On the other hand, the total of the area fractions of tempered martensite and lower bainite may be 0% as long as the area fractions of polygonal ferrite, upper bainite, retained γ, and fresh martensite can be controlled to the desired ranges.

Area fraction of remaining microstructure: 5% or less

[0063] The remaining microstructure is a microstructure other than polygonal ferrite, upper bainite, retained austenite, fresh martensite, tempered martensite, and lower bainite, and includes, for example, pearlite. A pearlite microstructure inhibits efficient carbon partitioning and suppresses the formation of retained γ, thus causing a decrease in ductility. In the annealing step in which the two-stage austempering treatment is performed, part of the austenite (γ) microstructure undergoes pearlite transformation depending on the chemical composition of the steel or the austempering temperature. In the present invention, influences on the material quality can be ignored as long as the area fraction of the remaining microstructure is 5% or less. Thus, the upper limit of the area fraction of the remaining microstructure is limited to 5%. The area fraction of the remaining microstructure is preferably 5.0% or less. The area fraction of the remaining microstructure may be 0%.

[0064] Ratio of the total number of fresh martensite grains and retained γ grains having an equivalent circular diameter of less than 1.2 um relative to the number of all fresh martensite grains and all retained austenite grains: 50% or more

[0065] Fresh martensite grains and retained γ grains having an equivalent circular diameter of less than 1.2 um are unlikely to serve as stress concentration sites during local deformation and do not contribute to void formation. Thus, they are microstructures that do not deteriorate local ductility and flangeability.

Excellent local elongation and flangeability can be obtained in the present invention when the total number of fresh martensite grains and retained γ grains having an equivalent circular diameter of less than 1.2 um is 50% or more of the number of all fresh martensite grains and all retained γ grains.

Thus, in the present invention, the ratio of the total number of fresh martensite grains and retained γ grains having an equivalent circular diameter of less than 1.2 um is limited to 50% or more relative to the number of all fresh martensite grains and all retained austenite grains. That is, formula (A) below is satisfied.

$$100 \times (\text{total number of fresh martensite grains and retained } \gamma \text{ grains having an equivalent circular diameter of less than } 1.2\ \mu m) / (\text{number of all fresh martensite grains and all retained } \gamma \text{ grains}) \geq 50\ (\%) \quad \text{Formula (A)}$$

The ratio of the left side specified by formula (A) is preferably 55% or more.

In order to obtain the above microstructure, upper bainite, tempered martensite, and lower bainite may be formed in the microstructure by the two-stage austempering process.

[0066] Ratio of the total number of fresh martensite grains and retained γ grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 um or more relative to the number of fresh martensite grains and retained γ grains having an equivalent circular diameter of 1.2 um or more: 40% or more

[0067] Stress concentration during local deformation can be reduced and the formation of voids can be suppressed to attain enhancements in local ductility and flangeability when fresh martensite grains and/or retained γ grains having an equivalent circular diameter of 1.2 um or more have a high aspect ratio of the fresh martensite grains and/or the retained γ grains.

The area fraction of such fresh martensite grains and/or retained γ grains can be increased by ensuring that acicular γ formed in the heating process and surrounded by a soft ferrite microstructure is transformed into bainite in the subsequent cooling process. In the present invention, desired formability can be obtained when the grains having an equivalent circular diameter of 1.2 um or more and an aspect ratio of 2.5 or more represent 40% or more of the total number of fresh martensite grains and retained γ grains having an equivalent circular diameter of 1.2 um or more.

Thus, in the present invention, the ratio of the total number of fresh martensite grains and retained austenite grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 um or more is limited to 40% or more relative to the number of fresh martensite grains and retained austenite grains having an equivalent circular diameter

of 1.2 um or more. That is, formula (B) below is further satisfied in addition to formula (A) described hereinabove. The ratio of the left side specified by formula (B) is preferably 45% or more.

$$100 \times (\text{total number of fresh martensite grains and retained } \gamma \text{ grains having an aspect ration of 2.5 or more and an equivalent circular diameter of 1.2 } \mu m \text{ or more}) / (\text{number of fresh martensite grains and retained } \gamma \text{ grains having an equivalent circular diameter of 1.2 } \mu m \text{ or more}) \geq 40 \quad \text{Formula (B)}$$

[0068] The microstructure of the steel sheet obtained is measured in the following manner.

Measurement of area fractions of steel microstructure

[0069] The steel sheet is cut to give an observation specimen so that a cross section perpendicular to the steel sheet surface and parallel to the rolling direction will be observed. The through-thickness cross section is etched with 1 vol% Nital. Microstructure images of 3000 μm^2 or larger regions are photographed at thickness t/4 locations with a scanning electron microscope (SEM) at a magnification of 2000 times. The images are analyzed to determine items (i) to (iv) below. Incidentally, the letter t indicates the sheet thickness and the letter w indicates the sheet width.

(i) Polygonal ferrite and upper bainite

[0070] Polygonal ferrite (recrystallized F) and upper bainite (UB) are both gray in SEM images but can be distinguished by their shapes. An exemplary SEM image is illustrated in Fig. 1 together with a SEM image of a microstructure cooled by water after held at a temperature T. The regions indicated by the dashed line in Fig. 1(a) are acicular γ microstructures formed by treatments up to soaking and holding at an annealing temperature T in the range of the present invention in the annealing step. Upper bainite (UB) is formed within the acicular γ microstructures and is surrounded by retained γ or fresh martensite (M) having a high aspect ratio. Similar microstructures are also seen in massive γ microstructures formed by treatments up to soaking and holding at an annealing temperature T. The area fractions of polygonal ferrite and upper bainite were measured by a point count method in accordance with ASTM E562-11 (2014). The area fraction of polygonal ferrite and the area fraction of upper bainite are each an average of values measured at five locations.

(ii) Fresh martensite and retained γ

[0071] Fresh martensite and retained γ are both white in SEM images and cannot be distinguished. Thus, retained γ was measured separately by a method described later. The total area fraction of fresh martensite and retained γ is measured from the SEM image by a point count method in accordance with ASTM E562-11 (2014), and the area fraction of retained γ measured by the method described later is subtracted from the total area fraction to determine the area fraction of fresh martensite. The total area fraction of fresh martensite and retained γ is measured by the point count method with respect to 5 locations, the measurement results being averaged, and the volume fraction of retained γ measured by the method described later is subtracted from the average value to give the area fraction of fresh martensite.

(iii) Tempered martensite and/or lower bainite

[0072] Tempered martensite and lower bainite are carbide-containing microstructures that are seen as white fine microstructures in SEM images. These two microstructures can be distinguished by more microscopic observation but are difficult to distinguish by SEM images. Thus, in the present invention, tempered martensite and lower bainite are defined as the same microstructure, and the total area fraction of tempered martensite and lower bainite is measured by a point count method in accordance with ASTM E562-11 (2014). The results measured at 5 locations are averaged to give the total area fraction of tempered martensite and lower bainite.

(iv) Remaining microstructure

[0073] The area fractions of polygonal ferrite, upper bainite, fresh martensite, retained γ, tempered martensite, and lower bainite measured by the above methods are subtracted from 100%. The difference is defined as the area fraction of the remaining microstructure.

Measurement of volume fraction of retained γ

[0074] The steel sheet is polished by 1/4 sheet thickness and is further polished by 0.1 mm by chemical polishing.

The exposed face is analyzed with an X-ray diffractometer using MoKα radiation to measure the integrated reflection intensities of (200) plane, (220) plane, and (311) plane of FCC iron (γ), and of (200) plane, (211) plane, and (220) plane of BCC iron (ferrite). The volume fraction of retained γ is determined from the intensity ratio of the integrated reflection intensity of the planes of FCC iron (γ) to the integrated reflection intensity of the planes of BCC iron (ferrite). In the present invention, the volume fraction of retained γ can be regarded as the area fraction of retained γ.

Equivalent circular diameter and aspect ratio of fresh martensite grains and/or retained γ grains

[0075] The steel sheet is cut to give an observation specimen so that a cross section parallel to the rolling direction will be observed. The microstructure on the through-thickness cross section is exposed by etching with LePera etchant. Microstructure images of 10000 μm^2 or larger regions are photographed at thickness t/4 locations with a laser microscope (LM) at a magnification of 1000 times. Lepera etching is color etching. Fresh martensite grains and/or retained γ grains are extracted by showing fresh martensite and/or retained γ in white contrast, and image analysis is performed to measure the equivalent circular diameter and the aspect ratio of the fresh martensite grains and/or the retained γ grains.
Of all the grains obtained, the number of grains having an equivalent circular diameter of less than 1.2 um is determined, and the ratio of those grains to the number of all the grains is calculated.
Of all the grains obtained, the number of grains having an equivalent circular diameter of 1.2 um or more is measured. Of those grains, the number of grains having an aspect ratio of 2.5 or more is determined. The ratio is calculated of the grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 um or more to all the grains having an equivalent circular diameter of 1.2 um or more.

[0076] Next, an embodiment of a method for manufacturing a steel sheet of the present invention will be described in detail. Unless otherwise specified, the temperatures of heating or cooling of steel, for example, a steel slab (a steel material) or a steel sheet, described below means the surface temperature of the steel, for example, the steel slab (the steel material) or the steel sheet.

[0077] A method for manufacturing a steel sheet of the present invention includes, after hot rolling and pickling are performed on a steel slab having the chemical composition described hereinabove, a cold rolling step of performing a cold rolling treatment on the hot rolled steel sheet to produce a cold rolled steel sheet, and an annealing step of performing an annealing treatment on the cold rolled steel sheet to produce a steel sheet. The cold rolling step is such that the cold rolled steel sheet is obtained by performing the cold rolling treatment in such a manner that the cumulative cold rolling reduction ratio is 30 to 85%, and the rolling reduction ratio in a first pass is 5% or more and less than 25%, thereby controlling the area fraction of the total of microstructures having {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation to 35% or more and 75% or less relative to all bcc phase microstructures. The annealing step is such that the annealing treatment includes heating the cold rolled steel sheet at an average heating rate of 0.5 to 15°C/sec in a range of temperatures of 500°C or above and Ac1 or below, to an annealing temperature T being 840°C or below and satisfying $0.5 \leq (T - Ac1)/(Ac3 - Ac1) < 1.0$; after the heating, soaking and holding the steel sheet at the annealing temperature T in a furnace atmosphere having a dew point Td of -50°C or above and -30°C or below, thereby producing a steel sheet having a number density of acicular austenite microstructures of 5 microstructures/1000 μm^2 or more; subsequently performing first cooling of cooling the steel sheet at an average cooling rate of 6.0°C/sec or more in a range of temperatures of 750 to 550°C, to a first cooling stop temperature Tc1 of 550°C or below and 400°C or above; after the first cooling, subjecting the steel sheet to first holding at the first cooling stop temperature Tc1 for 25 seconds or more; after the first holding, performing second cooling of cooling the steel sheet to a second cooling stop temperature Tc2 that is equal to or lower than the first cooling stop temperature Tc1 and is 450°C or below and 300°C or above; subjecting the steel sheet to second holding at the second cooling stop temperature Tc2 for 20 to 3000 seconds; and after the second holding, performing third cooling of cooling the steel sheet.

[0078] The steps will be described below.

Hot rolling step

[0079] In the present invention, for example, hot rolling in the hot rolling step may be performed in such a manner that the steel slab having the chemical composition described hereinabove is reheated and then rolled, that the steel slab from continuous casting is subjected to hot direct rolling without heating, or that the steel slab from continuous casting is heat treated for a short time and then rolled. The hot rolling may be performed in accordance with a conventional procedure. For example, the slab heating temperature may be 1100°C or above and 1300°C or below; the soaking time may be 20 to 30 minutes; the finish rolling temperature may be Ar3 transformation temperature (°C) or above and Ar3 transformation temperature (°C) + 200°C or below; and the coiling temperature may be 400 to 720°C. In order to eliminate or reduce thickness variations and to ensure high strength stably, the coiling temperature is preferably 430 to 620°C.

[0080] The steel slab (the steel material) may be produced by any smelting method without limitation. A known smelting technique, such as a converter or an electric arc furnace, may be used. Secondary refining may be performed in a

vacuum degassing furnace.

Pickling treatment step

[0081] In the pickling treatment step, the hot rolled steel sheet from the hot rolling step is subjected to a pickling treatment. The pickling treatment conditions are not particularly limited, and pickling treatment conditions in known production methods may be adopted.

Cold rolling step

Cumulative cold rolling reduction ratio: 30 to 85%

[0082] If the rolling reduction ratio (the cumulative cold rolling reduction ratio) in the cold rolling treatment is less than 30%, recrystallization is not sufficiently promoted and acicular γ discussed in the present invention is not formed sufficiently. Furthermore, the microstructure after the annealing step becomes nonuniform. Thus, the rolling reduction ratio (the cumulative cold rolling reduction ratio) in cold rolling needs to be 30% or more and is preferably 40% or more, and more preferably 50% or more. On the other hand, the rolling reduction ratio (the cumulative cold rolling reduction ratio) is 85% or less from the point of view of cold rolling load or further from the point of view of material quality.

[0083] In the cold rolling step, the number of passes is not particularly limited, and may be, for example, 5 passes.

The cumulative cold rolling reduction ratio (the thickness reduction ratio) indicates (1 - (sheet thickness after cold rolling (after final pass)/sheet thickness before cold rolling) $\times$ 100.

Rolling reduction ratio in the first pass: 5% or more and less than 25%

[0084] From the point of view of operability, the rolling reduction ratio in the first pass is limited to 5% or more. Because the sheet temperature at the time of the first pass of cold rolling is low, 25% or more rolling reduction ratio in the first pass applies shear strain components to the material being cold rolled, and the desired texture does not develop and acicular γ is not formed. Thus, the rolling reduction ratio in the first pass is limited to 5% or more and less than 25%.

[0085] Incidentally, the rolling reduction ratio (the thickness reduction ratio) in the first pass indicates (1 - (sheet thickness after first pass of cold rolling)/(sheet thickness before cold rolling)) $\times$ 100.

[0086] The rolling temperature (the sheet temperature) in the first pass is preferably 20°C or above and 40°C or below. The rolling temperature in the first pass is determined by measuring the temperature of a portion of the steel sheet surface free from the lubricating oil after the first pass with a radiation thermometer. If the rolling temperature in the first pass is below 20°C or if the rolling temperature in the first pass is above 40°C, the desired texture described hereinabove may not develop and acicular γ may not be formed. Thus, the rolling temperature in the first pass is preferably 20°C or above and 40°C or below.

[0087] Microstructure of the cold rolled steel sheet after cold rolling: The area fraction of the total of microstructures having {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation is 35% or more and 75% or less relative to all the bcc phase microstructures.

[0088] Acicular γ has a specific crystallographic orientation relationship (Near Kurdjumov-Sachs relationship) with ferrite surrounding its nucleation sites.

By ensuring that the cold rolled steel sheet after cold rolling has a certain or higher area fraction of the total of microstructures having specific orientations, specifically, {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation, relative to all the bcc phase microstructures, reverse transformed γ having the above specific orientations is formed easily between surrounding ferrite grains, and, as a result, a large amount of acicular γ is formed. In order to form a desired amount of acicular γ, it is necessary that the area fraction of the total of microstructures having {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation be 35% or more relative to all the bcc phase microstructures. The fraction is preferably 40% or more. If, on the other hand, the area fraction of the total of microstructures having {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation is more than 75% relative to all the bcc phase microstructures, anisotropy occurs in the material quality of the steel sheet. Thus, the area fraction of the total of microstructures having the above specific orientations is limited to 75% or less relative to all the bcc phase microstructures. The fraction is preferably 70% or less, and more preferably 65% or less.

[0089] In the present invention, the ratio of the area fraction of the total of microstructures having the above specific orientations to the area fraction of all the bcc phase microstructures can be brought to the desired range by subjecting

the hot rolled steel sheet having the chemical composition described hereinabove to the cold rolling treatment with a cold rolling reduction ratio of 30 to 85% while controlling the rolling reduction ratio in the first pass to 5% or more and less than 25%.

Method for measuring the texture of the cold rolled microstructure

[0090] The cold rolled steel sheet from the cold rolling step is cut to give a measurement specimen so that a cross section parallel to the rolling direction will be a measurement face. The measurement face is mechanically polished or electrolytically polished, and 80000 μm^2 or larger regions are analyzed by SEM-EBSD (measurement conditions: WD: 20 mm, acceleration voltage: 20 kV). The area fraction is quantified of bcc phase microstructures in which {ND plane} <RD direction> rolling orientations are {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation. The area fraction is expressed as a ratio to the area fraction of the bcc phases of all the orientations to evaluate the texture of the cold rolled steel sheet.

Annealing step

[0091] The annealing step in the present invention includes an annealing treatment described below. The cold rolled steel sheet from the cold rolling step is heated at an average heating rate (HR1) of 0.5 to 15°C/sec in a range of temperatures of 500°C or above and Ac1 or below, to an annealing temperature T that is 840°C or below and satisfies $0.5 \leq (T - Ac1)/(Ac3 - Ac1) < 1.0$. After the heating, the steel sheet is soaked and held at the annealing temperature T in a furnace atmosphere having a dew point Td of -50°C or above and -30°C or below, thereby giving a steel sheet having a number density of acicular austenite microstructures of 5 microstructures/1000 μm^2 or more. Subsequently, first cooling is performed in which the steel sheet is cooled at an average cooling rate of 6.0°C/sec or more in a range of temperatures of 750 to 550°C, to a first cooling stop temperature Tc1 of 550°C or below and 400°C or above. After the first cooling, the steel sheet is subjected to first holding at the first cooling stop temperature Tc1 for 25 seconds or more. After the first holding, second cooling is performed in which the steel sheet is cooled to a second cooling stop temperature Tc2 that is equal to or lower than the first cooling stop temperature Tc1 and is 450°C or below and 300°C or above. The steel sheet is then subjected to second holding at the second cooling stop temperature Tc2 for 20 to 3000 seconds. After the second holding, third cooling is performed in which the steel sheet is cooled.

Average heating rate in a range of temperatures of 500°C or above and Ac1 or below: 0.5 to 15°C/sec or less

[0092] In the present invention, the cold rolled sheet that has the microstructure after the cold rolling step described above is heated at an appropriate heating rate to sufficiently promote recrystallization and thereafter acicular γ is formed by heating of the steel sheet to the temperature T or by holding of the steel sheet at the temperature T. For this purpose, the average heating rate in a range of temperatures of 500°C or above and Ac1 or below where γ transformation does not occur is limited to 15°C/sec or less. The average heating rate is preferably 10°C/sec or less.
From the point of view of operation, the average heating rate is limited to 0.5°C/sec or more. The average heating rate is preferably 1.0°C/sec or more, and more preferably 1.5°C/sec or more.

[0093] Here, the average heating rate (°C/s) is calculated from ((Ac1 (°C) - 500°C)/(heating time (sec) from 500°C to Ac1 (°C)).

Heating to the annealing temperature T that is 840°C or below and satisfies $0.5 \leq (T - Ac1)/(Ac3 - Ac1) < 1.0$ After the heating, soaking and holding at the annealing temperature T in a furnace atmosphere having a dew point Td of -50°C or above and -30°C or below

[0094] In the present invention, acicular γ described later can be formed by heating the steel sheet to the annealing temperature T described later or by further holding the steel sheet at the annealing temperature T. If the steel sheet is heated to an γ single-phase region of Ac3 (°C) or above, acicular γ coalesces with adjacent γ, and the γ morphology becomes equiaxed. Thus, in the present invention, the annealing needs to be two-phase annealing.
If the annealing temperature T is such that (T - Ac1)/(Ac3 - Ac1) is less than 0.5, reverse transformation to γ does not occur sufficiently and acicular γ is not formed, and equiaxed γ is exclusively formed along recrystallized ferrite grain boundaries. Thus, the annealing temperature T is limited to satisfy $0.5 \leq (T - Ac1)/(Ac3 - Ac1) < 1.0$. If the temperature T is above 840°C, good chemical convertibility cannot be obtained. Thus, the temperature T is limited to 840°C or below.
If the dew point Td is below -50°C, the formation of Si oxide that adversely affects chemical convertibility is promoted and good chemical convertibility cannot be obtained. If the dew point Td is above -30°C, internal oxide layers are formed by selective oxide formation at grain boundaries in the microstructure, thus adversely affecting properties, such as corrosion resistance. Thus, the dew point Td is limited to -50°C or above and -30°C or below. The dew point Td is

preferably -48°C or above, and more preferably -46°C or above. The dew point Td is preferably -32°C or below, and more preferably -34°C or below.

[0095] The soaking time at the annealing temperature T is not particularly limited but is preferably 25 to 350 seconds, and more preferably 50 to 300 seconds from the point of view of element partitioning during the two-phase annealing.

[0096] Incidentally, Ac1 (°C) may be calculated from the formula below based on empirical rules.

$$Ac1\ (°C) = 723 + 22 \times [Si\%] - 18 \times [Mn\%] + 17 \times [Cr\%] + 4.5[Mo\%] + 16 \times [V\%]$$

[0097] Ac3 (°C) may be calculated from the formula below based on empirical rules.

$$Ac3\ (°C) = 910 - 203 \times [C\%]^{1/2} + 44.7 \times [Si\%] - 30 \times [Mn\%] + 700 \times [P\%] + 400 \times [sol.\ Al\%] - 20 \times [Cu\%] + 31.5 \times [Mo\%] + 104 \times [V\%] + 400 \times [Ti\%]$$

Incidentally, [X%] in the above formulas is the content (mass%) of component element X in the steel sheet and is "0" when the content is nil.

Number density of acicular γ microstructures formed by the soaking and holding treatment: 5 microstructures/1000 μm^2 or more

[0098] In the present invention, acicular γ is utilized to impart desired formability. Plenty of acicular austenite (acicular γ) promotes the formation of a large amount of retained γ having a high aspect ratio. In order to ensure that the present invention realizes the desired formability, the number density of acicular γ microstructures formed by the heating to and the soaking and holding at the annealing temperature T needs to be 5 microstructures/1000 μm^2 or more. By the nature of acicular γ, the upper limit is not limited and a larger number of acicular γ grains is more preferable.

[0099] In the present invention, the number density of acicular γ microstructures can be brought to the desired range by heating the cold rolled steel sheet having the chemical composition and the microstructure described hereinabove to the annealing temperature T in such a manner that the average heating rate in the range of temperatures of 500°C or above and Ac1 or below is 0.5 to 15°C/sec or less, and by soaking and holding the steel sheet at the annealing temperature T in a furnace atmosphere satisfying the dew point Td.

Number density of acicular γ microstructures

[0100] When a microstructure formed at a high temperature is to be evaluated, a common practice is to freeze the microstructure by water-cooling and evaluate the microstructure that has been formed. In the present invention, it is important that acicular γ formed by treatments in the annealing step up to the soaking and holding at the annealing temperature T contribute, in the subsequent cooling process, to the formation of retained γ having a high aspect ratio and high working stability.

Thus, the number density of the acicular γ microstructures is measured. The steel sheet is cut to give an observation specimen so that a cross section parallel to the rolling direction will be observed. The through-thickness cross section is etched with 1 vol% Nital. Microstructure images of 3000 μm^2 or larger regions are photographed at thickness t/4 locations with a scanning electron microscope (SEM) at a magnification of 2000 times. The SEM image illustrated in Fig. 1(b) is a photograph of a microstructure cooled by water after the steel sheet is held at a temperature T within the range of the present invention in the annealing step. The image shows that acicular γ, massive γ, and a ferrite microstructure are formed. Fig. 2 illustrates a schematic view of the measurement of the aspect ratio of acicular γ. Here, acicular γ is defined as austenite having an aspect ratio of 3.0 or more and surrounded by recrystallized ferrite having the same orientation. The tip of acicular austenite may be in contact with other austenite grains. In that case, the identicalness in orientation to the adjacent ferrite grains is confirmed by electron back-scatter diffractometry (EBSD). According to this definition, the number of acicular γ in the steel sheet that has been soaked and held at the annealing temperature T in the annealing step is measured with respect to 5 fields of view, and the number of acicular γ is divided by the total of the observed areas to give the number density of acicular γ (microstructures/1000 μm^2).

[0101] First cooling: The steel sheet is cooled at an average cooling rate of 6.0°C/sec or more in a range of temperatures of 750 to 550°C, to a first cooling stop temperature Tc1 of 550°C or below and 400°C or above.

First holding: After the first cooling, the steel sheet is held at the first cooling stop temperature Tc1 for 25 seconds or more.

[0102] In the first cooling, ferrite transformation and pearlite transformation occur predominantly in a range of temperatures of 750 to 550°C. In excessive ferrite transformation or pearlite transformation, acicular γ is transformed into ferrite. Thus, ferrite transformation is suppressed by controlling the average cooling rate in the range of temperatures of 750 to 550°C to 6.0°C/sec or more. The average cooling rate is preferably 8.0°C/sec or more, and more preferably 10.0°C/sec

or more.

[0103] Here, the average cooling rate (°C/sec) is calculated from (750°C (cooling start temperature) - 550°C (cooling stop temperature))/(cooling time (sec) from cooling start temperature to cooling stop temperature).

[0104] The first cooling stop temperature Tc1 in the first cooling corresponds to a temperature on the high-temperature side in two-stage austempering and is limited to 550°C or below and 400°C or above that is a range of temperatures where upper bainite transformation with less carbide precipitation occurs. If the first cooling stop temperature Tc1 is above 550°C, ferrite transformation or pearlite transformation occurs and local carbon partitioning does not take place, with the result that retained γ having high working stability is not formed. If, on the other hand, the first cooling stop temperature Tc1 is below 400°C, lower bainite transformation accompanied by carbide precipitation occurs. Furthermore, depending on the chemical composition or the annealing conditions, part of the microstructure undergoes martensite transformation and forms tempered martensite in the subsequent holding. Thus, the partitioning of carbon to non-transformed γ is retarded, and retained γ having high working stability is not formed. Thus, the first cooling stop temperature Tc1 is limited to 550°C or below and 400°C or above. The first cooling stop temperature Tc1 is preferably 420°C or above, and more preferably 450°C or above. The first cooling stop temperature Tc1 is preferably 530°C or below, and more preferably 510°C or below.

In the first holding after the first cooling, the first cooling stop temperature Tc1 may be modulated as long as the temperature is in the range of 550°C or below and 400°C or above. The holding time is 25 seconds or more. In this manner, upper bainite transformation can occur sufficiently. The holding time in the first holding is preferably 30 seconds or more, and more preferably 35 seconds or more.

The holding time in the first holding is preferably 60 seconds or less, and more preferably 55 seconds or less.

[0105] Second cooling: The steel sheet is cooled to a second cooling stop temperature Tc2 that is equal to or lower than the first cooling stop temperature Tc1 and is 450°C or below and 300°C or above.

[0106] Second holding: The steel sheet is held at the second cooling stop temperature Tc2 for 20 to 3000 seconds.

[0107] In the second cooling, the second cooling stop temperature Tc2 corresponds to a temperature on the low-temperature side in two-stage austempering. By holding the steel sheet at the second cooling stop temperature Tc2, low-solute C and thermally unstable non-transformed γ that will be transformed into martensite at the time of final cooling is caused to undergo bainite transformation after the holding at the first cooling stop temperature Tc1. This makes sure that retained γ will be formed. If the second cooling stop temperature Tc2 is a temperature that is equal to or lower than the first cooling stop temperature Tc1 and is above 450°C, bainite transformation stops and non-transformed γ is excessively transformed into martensite during the final cooling. If, on the other hand, the second cooling stop temperature Tc2 is below 300°C, non-transformed γ undergoes martensite transformation followed by holding. This promotes the formation of carbides and thus inhibits the formation of retained γ having a high solute C content and high working stability. Thus, the second cooling stop temperature Tc2 is limited to be equal to or lower than Tc1 and to be 450°C or below and 300°C or above. The second cooling stop temperature Tc2 is preferably 320°C or above, and more preferably 340°C or above. The second cooling stop temperature Tc2 is preferably 430°C or below, and more preferably 410°C or below.

In the second holding after the second cooling, the second cooling stop temperature Tc2 may be modulated as long as the temperature is equal to or lower than the first cooling stop temperature Tc1 and is 450°C or below and 300°C or above. The formation of retained γ having high working stability is promoted as long as the holding time is 20 seconds or more. From the point of view of operation, on the other hand, the holding time in the second holding is limited to 3000 seconds or less. The holding time in the second holding is preferably 100 seconds or more, and more preferably 200 seconds or more. The holding time in the second holding is preferably 2500 seconds or less, and more preferably 2000 seconds or less.

[0108] Third cooling: After the second holding, the steel sheet is cooled.

[0109] After the second holding, the steel sheet is cooled to room temperature (10 to 30°C). The steel sheet of the present invention is thus obtained.

[0110] After the annealing step, for example, temper rolling with an elongation ratio of 0.05 to 0.5% may be performed. However, the post treatment is not particularly limited thereto.

[0111] The steel sheet of the present invention that is obtained by the steel sheet manufacturing method of the present invention preferably has a thickness of 0.5 mm or more. The thickness is preferably 2.0 mm or less.

[0112] Next, a member and a method for manufacture thereof according to the present invention will be described.

[0113] The member of the present invention is obtained by subjecting the steel sheet of the present invention to at least one working of forming and joining. The method for manufacturing a member of the present invention includes a step of subjecting the steel sheet of the present invention to at least one working of forming and joining to produce a member.

[0114] The steel sheet of the present invention has a tensile strength of 590 MPa or more and has high ductility, excellent stretch flange formability, and good chemical convertibility. Thus, the member that is obtained using the steel sheet of the present invention also has high strength and has high ductility, excellent stretch flange formability, and good

chemical convertibility compared to the conventional high-strength members. Furthermore, weight can be reduced by using the member of the present invention. Thus, for example, the member of the present invention may be suitably used in an automobile body frame part. The member of the present invention also includes a welded joint.

[0115] The forming may be performed using any common working process, such as press working, without limitation. Furthermore, the joining may be performed using common welding, such as spot welding or arc welding, or, for example, riveting or caulking without limitation.

EXAMPLES

[0116] EXAMPLES of the present invention will be described below.

[0117] Steel slabs having a thickness of 250 mm and a chemical composition described in Table 1 were hot rolled (slab heating temperature: 1250°C, soaking time: 30 minutes, finish rolling temperature: Ar3 + 50°C, coiling temperature: 550°C) and pickled. The hot rolled steel sheets obtained were cold rolled under conditions described in Table 2. Cold rolled steel sheets were thus manufactured.

Next, the cold rolled steel sheets were annealed in a continuous annealing line under conditions described in Table 2 and were then temper rolled with an elongation ratio of 0.2 to 0.4%. Evaluation steel sheets were thus manufactured. The second holding at the second cooling stop temperature Tc2 was followed by third cooling, and the steel sheets were cooled to room temperature (20°C).

[Table 1]

Steel No.	Chemical composition (mass%)									Remarks
	C	Si	Mn	P	S	sol.Al	N	Others	Formula (1)*1	
A	0.089	1.11	2.43	0.006	0.0003	0.035	0.0030	Cu:0.14, REM: 0.0034	0.457	Compliant steel
B	0.112	0.77	1.82	0.005	0.0040	0.032	0.0028	B:0.0018, Ti:0.019	0.423	Compliant steel
C	0.132	1.42	2.88	0.003	0.0005	0.029	0.0019	-	0.493	Compliant steel
D	0.175	0.84	1.95	0.006	0.0004	0.022	0.0035	B:0.0018, Ti:0.021, Sb:0.021	0.431	Compliant steel
E	0.089	0.64	3.12	0.004	0.0013	0.292	0.0022	Nb:0.020, Sb:0.009	0.205	Compliant steel
F	0.203	0.55	1.92	0.003	0.0011	0.031	0.0031	Sn:0.022	0.286	Compliant steel
G	0.183	1.04	2.38	0.011	0.0005	0.070	0.0035	V:0.012, Mg: 0.0008	0.437	Compliant steel
H	0.091	0.92	2.78	0.007	0.0002	0.066	0.0044	Ni:0.07, Mo:0.11	0.331	Compliant steel
I	0.110	0.96	2.67	0.012	0.0009	0.032	0.0023	Cr:0.17, Ca: 0.0016	0.273	Compliant steel
J	0.132	0.88	3.52	0.013	0.0016	0.039	0.0013	-	0.370	Comparative steel
K	0.261	1.04	2.38	0.011	0.0005	0.070	0.0035	B:0.0034, Ti:0.022	0.437	Comparative steel
L	0.091	1.65	3.11	0.007	0.0002	0.070	0.0044	-	0.531	Comparative steel

(continued)

Steel No.	Chemical composition (mass%)									Remarks
	C	Si	Mn	P	S	sol.Al	N	Others	Formula (1)*1	
M	0.054	0.78	2.56	0.005	0.0014	0.029	0.0033	-	0.305	Comparative steel
N	0.119	1.12	2.12	0.002	0.0022	0.038	0.0022	-	0.528	Comparative steel

·The balance other than the above components is Fe and incidental impurities.
Note: Underlines indicate being outside of the range of the present invention.
*1: Formula (1): Si/Mn (In formula (1), Si and Mn indicate the Si content (mass%) and the Mn content (mass%), respectively.)

[Table 2]

No.	Steel No.	Cold rolling conditions				Annealing conditions													Remarks
		Rolling reduction ratio in first pass (%)	Rolling temp. in first pass (°C)	Cumulative cold rolling reduction ratio (%)	Area fraction of BCC microstructures having specific orientations*2 (%)	Dew point Td (°C)	Annealing heating rate*3 (°C/s)	Annealing temp. T (°C)	Ac1 (°C)	Ac3 (°C)	(T-Ac1)/(Ac3-Ac1)	Soaking time (s)	Number density of acicular austenite*4 (microstructures /1000μm^2)	Average cooling rate*5 (°C/s)	First finish cooling temp. Tc1 (°C)	Holding time at first finish cooling temp. Tc1 (sec)	Second finish cooling temp. Tc2 (°C)	Holding time at second finish cooling temp. Tc2 (sec)	
1	A	15	35	55	41.3	-40	12	815	704	842	0.8	220	6.7	15.2	540	30	430	650	INV. EX.
2	A	15	34	55	42.3	-40	12	875	704	842	1.2	220	0.0	19.8	540	30	430	650	COMP. EX.
3	A	15	33	55	49.3	-40	12	820	704	842	0.8	220	8.2	2.1	540	30	430	650	COMP. EX.
4	A	15	35	55	55.9	-40	12	805	704	842	0.7	220	7.4	14.7	540	30	430	650	INV. EX.
5	B	15	36	55	56.3	-35	12	810	707	846	0.7	220	6.5	15.5	570	45	430	650	COMP. EX.
6	B	32	55	55	27.0	-35	12	805	707	846	0.7	220	3.2	34.2	520	45	430	650	COMP. EX.
7	B	15	32	55	52.1	-35	25	800	707	846	0.7	220	2.2	15.2	520	45	350	650	COMP. EX.
8	B	15	33	55	44.3	-35	10	810	707	846	0.7	220	6.3	24.5	520	45	350	480	INV. EX.
9	C	29	52	55	12.8	-40	10	795	702	827	0.7	220	1.2	25.2	520	25	350	480	COMP. EX.
10	C	15	32	55	55.3	-40	10	795	702	827	0.7	220	9.4	19.6	520	25	350	1200	INV. EX.
11	C	15	33	55	42.1	-40	10	805	702	827	0.8	220	8.3	8.4	300	25	350	650	COMP. EX.
12	C	15	32	55	39.8	-49	10	800	702	827	0.8	220	5.6	20.2	450	25	300	480	INV. EX.
13	D	15	31	55	46.3	-49	10	805	706	826	0.8	220	6.2	30.2	450	30	300	15	COMP. EX.
14	D	15	36	55	44.8	-45	10	810	706	826	0.9	220	5.3	14.8	550	30	450	650	INV. EX.

(continued)

No.	Steel No.	Cold rolling conditions				Annealing conditions													Remarks
		Rolling reduction ratio in first pass (%)	Rolling temp. in first pass (°C)	Cumulative cold rolling reduction ratio (%)	Area fraction of BCC microstructures having specific orientations*2 (%)	Dew point Td (°C)	Annealing heating rate*3 (°C/s)	Annealing temp. T (°C)	Ac1 (°C)	Ac3 (°C)	(T-Ac1)/(Ac3-Ac1)	Soaking time (s)	Number density of acicular austenite*4 (microstructures /1000μm^2)	Average cooling rate*5 (°C/s)	First finish cooling temp. Tc1 (°C)	Holding time at first finish cooling temp. Tc1 (sec)	Second finish cooling temp. Tc2 (°C)	Holding time at second finish cooling temp. Tc2 (sec)	
15	D	15	30	55	42.9	-45	35	800	706	826	0.8	220	2.3	14.3	550	30	520	650	COMP. EX.
16	E	15	32	40	37.5	-45	10	850	681	904	0.8	220	5.6	15.2	400	35	280	650	COMP. EX.
17	E	15	33	40	35.9	-45	10	820	681	904	0.6	220	6.4	13.4	400	35	450	650	INV. EX.
18	E	15	34	40	38.3	-32	15	810	681	904	0.6	220	6.3	14.8	450	35	450	650	INV. EX.
19	E	15	30	40	40.2	-32	15	810	681	904	0.6	220	6.3	14.8	450	35	450	650	INV. EX.
20	E	10	31	22	10.3	-32	15	820	681	904	0.6	220	3.1	18.9	450	35	450	600	COMP. EX.
21	F	15	31	48	42.6	-55	15	810	701	800	1.1	220	0.0	21.0	570	30	450	600	COMP. EX.
22	F	15	31	48	45.6	-43	18	810	701	800	1.1	220	0.0	21.0	500	30	450	600	COMP. EX.
23	F	15	32	48	51.2	-43	18	780	701	800	0.8	220	6.9	5.0	500	30	450	600	COMP. EX.
24	F	15	33	48	42.3	-43	18	820	701	800	1.2	220	0.0	10.5	500	30	450	650	COMP. EX.
25	G	15	34	75	52.3	-45	12	790	703	835	0.7	220	7.8	14.5	500	60	450	650	INV. EX.
26	G	15	33	75	54.3	-40	12	800	703	835	0.7	120	6.5	14.3	500	60	400	650	INV. EX.
27	G	15	36	75	52.3	-40	12	810	703	835	0.8	120	8.3	14.5	500	60	400	650	INV. EX.
28	H	15	35	75	49.6	-40	12	830	694	841	0.9	120	7.9	15.5	500	30	400	650	INV. EX.

(continued)

No.	Steel No.	Cold rolling conditions				Annealing conditions												Remarks
		Rolling reduction ratio in first pass (%)	Rolling temp. in first pass (°C)	Cumulative cold rolling reduction ratio (%)	Area fraction of BCC microstructures having specific orientations*2 (%)	Dew point Td (°C)	Annealing heating rate*3 (°C/s)	Annealing temp. T (°C)	Ac1 (°C)	Ac3 (°C)	(T-Ac1)/(Ac3-Ac1)	Soaking time (s)	Number density of acicular austenite*4 (microstructures /1000μm^2)	Average cooling rate*5 (°C/s)	First finish cooling temp. Tc1 (°C)	Holding time at first finish cooling temp. Tc1 (sec)	Second finish cooling temp. Tc2 (°C)	Holding time at second finish cooling temp. Tc2 (sec)	
29	H	15	34	75	53.6	-40	12	820	694	841	0.9	120	6.5	14.9	500	30	400	650	INV. EX.
30	I	15	36	48	43.3	-40	12	790	699	827	0.7	220	7.7	15.5	500	30	400	600	INV. EX.
31	I	15	36	48	41.3	-40	12	805	699	827	0.8	220	9.3	15.2	500	30	400	650	INV. EX.
32	J	15	35	48	43.2	-40	12	805	679	795	1.1	220	0.0	15.0	500	30	400	650	COMP. EX.
33	K	15	33	48	42.2	-40	12	805	703	826	0.8	220	5.9	15.0	500	30	400	650	COMP. EX.
34	L	15	29	48	41.3	-40	12	805	703	862	0.6	220	8.4	15.0	500	30	400	650	COMP. EX.
35	M	15	36	48	43.3	-40	12	805	694	836	0.8	220	6.5	15.0	500	30	400	650	COMP. EX.
36	N	15	33	48	40.3	-40	12	805	709	843	0.7	220	6.6	15.0	500	30	400	650	COMP. EX.
37	A	15	34	55	42.3	-40	12	830	704	842	0.9	220	5.5	15.2	580	30	280	650	COMP. EX.

Note: Underlines indicate being outside of the range of the present invention.

*2: Ratio of the total of bcc phase microstructures having {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation relative to all the bcc phase microstructures

*3: Average heating rate in a range of temperatures of 500°C to Ac1

*4: Number density of acicular austenite microstructures formed by treatments up to heating and holding at the temperature T

*5: Average cooling rate in a range of temperatures of 750 to 550°C

[0118] The steel sheets obtained were evaluated in the following manner.

(1) Measurement of area fractions of steel microstructure and measurement of number density of acicular γ

[0119] The steel sheet was cut to give an observation specimen so that a cross section perpendicular to the steel sheet surface and parallel to the rolling direction would be observed. The through-thickness cross section was etched with 1 vol% Nital. Microstructure images of 3000 μm^2 or larger regions were photographed at thickness t/4 locations with a scanning electron microscope (SEM) at a magnification of 2000 times. The images were analyzed to determine items (i) to (iv) below. The results are described in Table 3. Incidentally, the letter t indicates the sheet thickness and the letter w indicates the sheet width.

(i) Polygonal ferrite and upper bainite

[0120] Polygonal ferrite (recrystallized F) and upper bainite (UB) are both gray in SEM images but can be distinguished by their shapes. An exemplary SEM image is illustrated in Fig. 1 together with a SEM image of a microstructure cooled by water after held at a temperature T. The regions indicated by the dashed line in Fig. 1(a) are acicular γ microstructures formed by treatments up to soaking and holding at an annealing temperature T in the range of the present invention in the annealing step. Upper bainite (UB) is formed within the acicular γ microstructures and is surrounded by retained γ or fresh martensite (M) having a high aspect ratio. Similar microstructures are also seen in massive γ microstructures formed by treatments up to soaking and holding at an annealing temperature T. The area fractions of polygonal ferrite and upper bainite were measured by a point count method in accordance with ASTM E562-11 (2014). The area fraction of polygonal ferrite and the area fraction of upper bainite were each obtained as an average of values measured at five locations.

(ii) Fresh martensite and retained γ

[0121] Fresh martensite and retained γ are both white in SEM images and cannot be distinguished. Thus, retained γ was measured separately by a method described later. The total area fraction of fresh martensite and retained γ was measured from the SEM image by a point count method in accordance with ASTM E562-11 (2014), and the area fraction of retained γ measured by the method described later was subtracted from the total area fraction to determine the area fraction of fresh martensite. The total area fraction of fresh martensite and retained γ was measured by the point count method with respect to 5 locations, the measurement results being averaged, and the volume fraction of retained γ measured by the method described later was subtracted from the average value to give the area fraction of fresh martensite.

(iii) Tempered martensite and/or lower bainite

[0122] Tempered martensite and lower bainite are carbide-containing microstructures that are seen as white fine microstructures in SEM images. These two can be distinguished by more microscopic observation but are difficult to distinguish by SEM images. Thus, in the present invention, tempered martensite and lower bainite were defined as the same microstructure, and the total area fraction of tempered martensite and lower bainite was measured by a point count method in accordance with ASTM E562-11 (2014). The results measured at 5 locations were averaged to give the total area fraction of tempered martensite and lower bainite.

(iv) Remaining microstructure

[0123] The area fractions of polygonal ferrite, upper bainite, fresh martensite, retained γ, tempered martensite, and lower bainite measured by the above methods were subtracted from 100%. The difference was defined as the area fraction of the remaining microstructure.

(2) Measurement of volume fraction of retained γ

[0124] The steel sheet was polished by 1/4 sheet thickness and was further polished by 0.1 mm by chemical polishing. The exposed face was analyzed with an X-ray diffractometer using MoKα radiation to measure the integrated reflection intensities of (200) plane, (220) plane, and (311) plane of FCC iron (γ), and of (200) plane, (211) plane, and (220) plane of BCC iron (ferrite). The volume fraction of retained γ was determined from the intensity ratio of the integrated reflection intensity of the planes of FCC iron (γ) to the integrated reflection intensity of the planes of BCC iron (ferrite). In the present invention, the volume fraction of retained γ can be regarded as the area fraction of retained γ.

(3) Number density of acicular γ microstructures

[0125] When a microstructure formed at a high temperature is to be evaluated, a common practice is to freeze the microstructure by water-cooling and evaluate the microstructure that has been formed. In the present invention, it is important that acicular γ formed by treatments in the annealing step up to the soaking and holding at the annealing temperature T contribute, in the subsequent cooling process, to the formation of retained γ having a high aspect ratio and high working stability.

[0126] Thus, the number density of the acicular γ microstructures was measured. The steel sheet was cut to give an observation specimen so that a cross section parallel to the rolling direction would be observed. The through-thickness cross section was etched with 1 vol% Nital. Microstructure images of 3000 μm^2 or larger regions were photographed at thickness t/4 locations with a scanning electron microscope (SEM) at a magnification of 2000 times. The SEM image illustrated in Fig. 1(b) is a photograph of a microstructure cooled by water after the steel sheet was held at a temperature T within the range of the present invention in the annealing step. The image shows that acicular γ, massive γ, and a ferrite microstructure were formed. Fig. 2 illustrates a schematic view of the measurement of the aspect ratio of acicular γ. Here, acicular γ is defined as austenite having an aspect ratio of 3.0 or more and surrounded by recrystallized ferrite having the same orientation. The tip of acicular austenite may be in contact with other austenite grains. In that case, the identicalness in orientation to the adjacent ferrite grains is confirmed by electron back-scatter diffractometry (EBSD). According to this definition, the number of acicular γ in the steel sheet that had been soaked and held at the annealing temperature T in the annealing step was measured with respect to 5 fields of view, and the number of acicular γ was divided by the total of the observed areas to give the number density of acicular γ (microstructures/1000 μm^2).

[0127] The results are described in Table 3.

(4) Equivalent circular diameter and aspect ratio of fresh martensite grains and/or retained γ grains

[0128] The steel sheet was cut to give an observation specimen so that a cross section parallel to the rolling direction would be observed. The microstructure on the through-thickness cross section was exposed by etching with LePera etchant. Microstructure images of 10000 μm^2 or larger regions were photographed at thickness t/4 locations with a laser microscope (LM) at a magnification of 1000 times. Lepera etching is color etching. Fresh martensite grains and/or retained γ grains were extracted by showing fresh martensite and/or retained γ in white contrast, and image analysis was performed to measure the equivalent circular diameter and the aspect ratio of the fresh martensite grains and/or the retained γ grains.

Of all the grains obtained, the number of grains having an equivalent circular diameter of less than 1.2 um was determined, and the ratio of those grains to the number of all the grains was calculated.

Of all the grains obtained, the number of grains having an equivalent circular diameter of 1.2 um or more was measured. Of those grains, the number of grains having an aspect ratio of 2.5 or more was determined. The ratio was calculated of the grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 um or more to all the grains having an equivalent circular diameter of 1.2 um or more. The results are described in Table 3.

(5) Texture of cold rolled microstructure

[0129] The cold rolled steel sheet from the cold rolling step was cut to give a measurement specimen so that a cross section parallel to the rolling direction would be a measurement face. The measurement face was mechanically polished or electrolytically polished, and 80000 μm^2 or larger regions were analyzed by SEM-EBSD (measurement conditions: WD: 20 mm, acceleration voltage: 20 kV). The area fraction was quantified of bcc phase microstructures in which {ND plane} <RD direction> rolling orientations were {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation. The area fraction was expressed as a ratio to the area fraction of the bcc phases of all the orientations to evaluate the texture of the cold rolled steel sheet.

(6) Tensile test

[0130] JIS No. 5 test pieces for tensile test were fabricated from the steel sheet so that the tensile direction would be perpendicular to the rolling direction. The test pieces were each subjected to a tensile test in accordance with the provisions of JIS Z2241 (2011). The crosshead speed in the tensile test was 10 mm/min. The measurement was performed twice, and the measured values were averaged to give the tensile strength (TS) of the steel sheet.

(7) Hole expansion test

[0131] 100 mm × 100 mm test specimens were sampled. Three test specimens from each sampling location were

tested by a hole expansion test in accordance with JFST (The Japan Iron and Steel Federation Standard) 1001. The results of the three measurements were averaged (total of values of three measurements (%)/3) to give the hole expansion ratio λ (%).

(8) Evaluation

[0132] In the present invention, the steel sheets were evaluated as having high strength when the tensile strength (TS) was 590 MPa or more.
The ductility El was evaluated as excellent when tensile strength (TS) × total elongation (T. El) ≥ 22000 MPa-% or more.
The stretch flange formability λ was evaluated as excellent when the hole expansion ratio λ (%) satisfied (A1) or (A2) below.

(A1) The hole expansion ratio λ was 60% or more when the tensile strength was 590 MPa or more and less than 780 MPa.
(A2) The hole expansion ratio λ was 35% or more when the tensile strength was 780 MPa or more.

(9) Chemical convertibility

[0133] The steel sheet after annealing was electrolytically pickled with sulfuric acid for 2 seconds at a current density of 20 to 35 A/dm^2 and was degreased and surface conditioned. Subsequently, chemical conversion was performed using a zinc phosphate chemical conversion solution. In the chemical conversion, the degreasing step involved treatment temperature: 40°C, treatment time: 120 seconds, and spray degreasing; the surface conditioning step involved pH: 9.5, treatment temperature: room temperature, and treatment time: 20 seconds; and the chemical conversion step involved chemical conversion solution temperature: 35°C and treatment time: 120 seconds. The degreasing step, the surface conditioning step, and the chemical conversion step involved the following treatment agents, respectively: degreasing agent: FC-E2011, surface conditioning agent: PL-X, and chemical conversion solution: PALBOND PB-L3065, each manufactured by Nihon Parkerizing Co., Ltd. The surface chemical conversion microstructure was observed with respect to 10000 μm^2 or larger regions by SEM at a magnification of 2000 times. The chemical convertibility was evaluated as o when the chemical conversion coating microstructure was present on all the faces, and was evaluated as × when the chemical conversion coating microstructure was visually found to be absent no matter how small the size. The results are described in Table 3.

[Table 3]

No.	Steel No.	Microstructure								Properties				Remarks
		Area fraction of polygonal ferrite (%)	Area fraction of upper bainite (%)	Volume fraction of retained γ (%)	Area fraction of fresh martensite (%)	Total area fraction of tempered martensite and lower bainite (%)	Area fraction of remaining microstructure (%)	Ratio of less than 1.2 μm grains*6 (%)	Ratio of grains with 2.5 or higher aspect ratio*7 (%)	TS (MPa)	TS*T.EL (MPa*%)	Hole expansion ratio A (%)	Chemical convertibility (rating)	
1	A	69.2	10.4	8.1	10.3	1.5	0.5	56.3	45.0	612	24664	66	○	INV. EX.
2	A	18.2	12.7	4.6	12.2	52.3	0.0	59.6	31.2	956	20363	82	×	COMP. EX.
3	A	88.2	2.7	3.6	5.5	0.0	0.0	42.0	25.6	552	22025	78	○	COMP. EX.
4	A	65.2	11.1	8.4	5.4	9.9	0.0	63.2	45.2	622	25067	71	○	INV. EX.
5	B	69.4	14.3	2.4	3.3	4.2	6.4	55.3	52.1	625	20125	45	○	COMP. EX.
6	B	72.2	13.2	6.4	3.3	4.9	0.0	52.0	32.4	632	18644	37	○	COMP. EX.
7	B	74.2	11.3	4.5	5.5	4.5	0.0	53.3	294	599	19288	44	○	COMP. EX.
8	B	68.4	9.9	6.8	4.4	8.2	2.3	67.2	49.5	622	23201	68	○	INV. EX.
9	C	58.4	13.8	12.2	10.1	5.5	0.0	56.3	21.6	812	23710	32	○	COMP. EX.
10	C	54.2	14.1	11.1	4.2	16.4	0.0	66.9	49.5	832	22797	49	○	INV. EX.
11	C	37.5	10.2	1.1	0.0	51.2	0.0	56.3	52.6	834	16513	65	○	COMP. EX.
12	C	46.8	12.3	8.4	5.4	27.1	0.0	66.4	48.3	812	22168	45	○	INV. EX.
13	D	43.2	21.3	13.1	18.3	4.1	0.0	41.2	56.3	837	20172	22	○	COMP. EX.
14	D	39.8	17.6	12.4	4.6	24.4	1.2	57.9	47.2	817	24592	48	○	INV. EX.

(continued)

No.	Steel No.	Microstructure								Properties				Remarks
		Area fraction of polygonal ferrite (%)	Area fraction of upper bainite (%)	Volume fraction of retained γ (%)	Area fraction of fresh martensite (%)	Total area fraction of tempered martensite and lower bainite (%)	Area fraction of remaining microstructure (%)	Ratio of less than 1.2 μm grains*6 (%)	Ratio of grains with 2.5 or higher aspect ratio*7 (%)	TS (MPa)	TS*T.EL (MPa*%)	Hole expansion ratio A (%)	Chemical convertibility (rating)	
15	D	45.9	12.4	11.0	12.8	17.9	0.0	56.4	26.7	834	20099	29	○	COMP. EX.
16	E	38.4	11.2	7.0	322	11.2	0.0	34.5	46.2	899	21037	22	×	COMP. EX.
17	E	42.1	18.4	8.5	11.5	19.5	0.0	56.4	56.2	813	22601	49	○	INV. EX.
18	E	53.3	12.9	9.8	5.6	18.4	0.0	67.2	49.2	832	24461	52	○	INV. EX.
19	E	49.8	11.2	11.4	11.2	16.4	0.0	56.4	45.2	792	22493	48	○	INV. EX.
20	E	44.9	13.2	10.2	12.5	18.4	0.8	45.0	38.2	821	20115	22	○	COMP. EX.
21	F	29.4	10.3	11.2	5.5	41.1	2.5	59.3	26.3	873	24619	23	×	COMP. EX.
22	F	26.4	11.7	10.3	8.4	43.2	0.0	55.4	27.4	823	18518	31	○	COMP. EX.
23	F	65.3	10.2	4.5	4.5	10.1	5.4	64.1	54.1	792	20671	43	○	COMP. EX.
24	F	24.4	21.1	8.4	11.2	34.9	0.0	63.1	18.2	821	17734	19	○	COMP. EX.
25	G	45.6	24.2	13.4	5.6	11.2	0.0	54.3	62.3	842	23071	44	○	INV. EX.
26	G	38.4	35.4	13.7	5.3	7.2	0.0	59.4	45.2	821	22988	50	○	INV. EX.
27	G	25.3	25.3	14.1	14.2	21.1	0.0	56.7	52.1	846	23519	44	○	INV. EX.
28	H	76.1	11.1	5.6	4.2	3.0	0.0	52.1	44.4	613	23539	77	○	INV. EX.

(continued)

No.	Steel No.	Microstructure								Properties				Remarks
		Area fraction of polygonal ferrite (%)	Area fraction of upper bainite (%)	Volume fraction of retained γ (%)	Area fraction of fresh martensite (%)	Total area fraction of tempered martensite and lower bainite (%)	Area fraction of remaining microstructure (%)	Ratio of less than 1.2 μm grains*6 (%)	Ratio of grains with 2.5 or higher aspect ratio*7 (%)	TS (MPa)	TS*T.EL (MPa*%)	Hole expansion ratio A (%)	Chemical convertibility (rating)	
29	H	76.7	9.1	4.1	3.9	6.2	0.0	71.1	49.3	633	24750	65	○	INV. EX.
30	I	63.1	12.8	6.3	4.3	13.5	0.0	65.4	53.1	786	22873	49	○	INV. EX.
31	I	57.4	13.2	11.9	5.4	12.1	0.0	59.3	46.2	821	23481	55	○	INV. EX.
32	J	61.1	2.1	2.4	20.5	13.9	0.0	66.3	42.1	1121	15918	31	○	COMP. EX.
33	K	55.6	6.7	4.5	31.2	2.0	0.0	71.2	53.3	1234	13697	19	○	COMP. EX.
34	L	45.6	14.3	12.1	8.8	19.2	0.0	64.2	44.3	823	23702	46	×	COMP. EX.
35	M	81.1	5.4	2.9	4.5	6.1	0.0	68.2	56.3	552	20645	81	○	COMP. EX.
36	N	55.5	12.3	10.9	8.4	12.9	0.0	64.2	44.3	823	23538	46	×	COMP. EX.
37	A	76.3	10.9	3.8	4.5	4.5	0.0	32.0	49.3	799	23091	28	○	COMP. EX.

Note: Underlines indicate being outside of the range of the present invention.
*6: Ratio of fresh martensite grains and/or retained γ grains having an equivalent circular diameter of less than 1.2 μm relative to all the grains
*7: Ratio of fresh martensite grains and/or retained γ grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 μm or more relative to all the fresh martensite grains and/or all the retained γ grains having an equivalent circular diameter of 1.2 μm or more

[0134] As described in Table 3, the steel sheets of the present invention were shown to have 590 MPa or higher tensile strength, high ductility, and excellent stretch flange formability and also to be excellent in chemical convertibility.

[0135] The steel sheets of INVENTIVE EXAMPLES have high strength, high ductility, excellent stretch flange formability, and good chemical convertibility. This has shown that members obtained by forming of the steel sheets of INVENTIVE EXAMPLES, members obtained by joining of the steel sheets of INVENTIVE EXAMPLES, and members obtained by forming and joining of the steel sheets of INVENTIVE EXAMPLES will have high strength, high ductility, excellent stretch flange formability, and good chemical convertibility similarly to the steel sheets of INVENTIVE EXAMPLES.

Claims

1. A steel sheet having a chemical composition comprising, in mass%:

C: 0.06 to 0.24%,
Si: 0.4% or more and less than 1.60%,
Mn: 1.5 to 3.2%,
P: 0.02% or less,
S: 0.01% or less,
sol. Al: less than 1.0%, and
N: less than 0.015%,
the chemical composition satisfying formula (1) below,
the balance being Fe and incidental impurities,
the steel sheet comprising a microstructure in which:

the area fraction of polygonal ferrite is 20% or more and 85% or less,
the area fraction of upper bainite is 9% or more and 45% or less,
the volume fraction of retained austenite is 3% or more and 15% or less,
the area fraction of fresh martensite is 3% or more and 15% or less,
the total of the area fractions of tempered martensite and lower bainite is 50% or less (including 0%), and
the area fraction of a remaining microstructure is 5% or less,
the microstructure being such that:

the ratio of the total number of fresh martensite grains and retained austenite grains having an equivalent circular diameter of less than 1.2 um is 50% or more relative to the number of all fresh martensite grains and all retained austenite grains, and
the ratio of the total number of fresh martensite grains and retained austenite grains having an aspect ratio of 2.5 or more and an equivalent circular diameter of 1.2 um or more is 40% or more relative to the number of fresh martensite grains and retained austenite grains having an equivalent circular diameter of 1.2 um or more,

$$Si/Mn < 0.50 \cdots \text{Formula (1)}$$

wherein in formula (1), Si and Mn indicate the Si content (mass%) and the Mn content (mass%), respectively.

2. The steel sheet according to claim 1, wherein
the chemical composition further comprises, in mass%, one, or two or more selected from:

Nb: 0.2% or less,
Ti: 0.2% or less,
V: 0.2% or less,
B: 0.01% or less,
Cu: 0.2% or less,
Ni: 0.2% or less,
Cr: 0.4% or less, and
Mo: 0.15% or less.

3. The steel sheet according to claim 1 or 2, wherein
the chemical composition further comprises, in mass%, one, or two or more selected from:

Mg: 0.0050% or less,
Ca: 0.0050% or less,
Sn: 0.10% or less,
Sb: 0.10% or less, and
REM: 0.0050% or less.

4. A member obtained using the steel sheet described in any of claims 1 to 3.

5. A method for manufacturing a steel sheet, comprising, after hot rolling and pickling are performed on a steel slab having the chemical composition described in any of claims 1 to 3, a cold rolling step of performing a cold rolling treatment on the hot rolled steel sheet to produce a cold rolled steel sheet, and

an annealing step of performing an annealing treatment on the cold rolled steel sheet to produce a steel sheet, the cold rolling step being such that the cold rolled steel sheet is obtained by performing the cold rolling treatment in such a manner that:

the cumulative cold rolling reduction ratio is 30 to 85%, and
the rolling reduction ratio in a first pass is 5% or more and less than 25%, thereby controlling the area fraction of the total of microstructures having {111} <0-11> orientation, {111} <11-2> orientation, {211} <0-11> orientation, and {100} <011> orientation to 35% or more and 75% or less relative to all bcc phase microstructures,
the annealing step being such that the annealing treatment comprises:

heating the cold rolled steel sheet at an average heating rate of 0.5 to 15°C/sec in a range of temperatures of 500°C or above and Ac1 or below, to an annealing temperature T being 840°C or below and satisfying $0.5 \leq (T - Ac1)/(Ac3 - Ac1) < 1.0$;
after the heating, soaking and holding the steel sheet at the annealing temperature T in a furnace atmosphere having a dew point Td of -50°C or above and -30°C or below, thereby producing a steel sheet having a number density of acicular austenite microstructures of 5 microstructures/1000 μm^2 or more;
subsequently performing first cooling of cooling the steel sheet at an average cooling rate of 6.0°C/sec or more in a range of temperatures of 750 to 550°C, to a first cooling stop temperature Tc1 of 550°C or below and 400°C or above;
after the first cooling, subjecting the steel sheet to first holding at the first cooling stop temperature Tc1 for 25 seconds or more;
after the first holding, performing second cooling of cooling the steel sheet to a second cooling stop temperature Tc2 that is equal to or lower than the first cooling stop temperature Tc1 and is 450°C or below and 300°C or above;
subjecting the steel sheet to second holding at the second cooling stop temperature Tc2 for 20 to 3000 seconds; and
after the second holding, performing third cooling of cooling the steel sheet.

6. A method for manufacturing a member, comprising a step of subjecting the steel sheet described in any of claims 1 to 3 to at least one working of forming and joining to produce a member.

FIG. 1

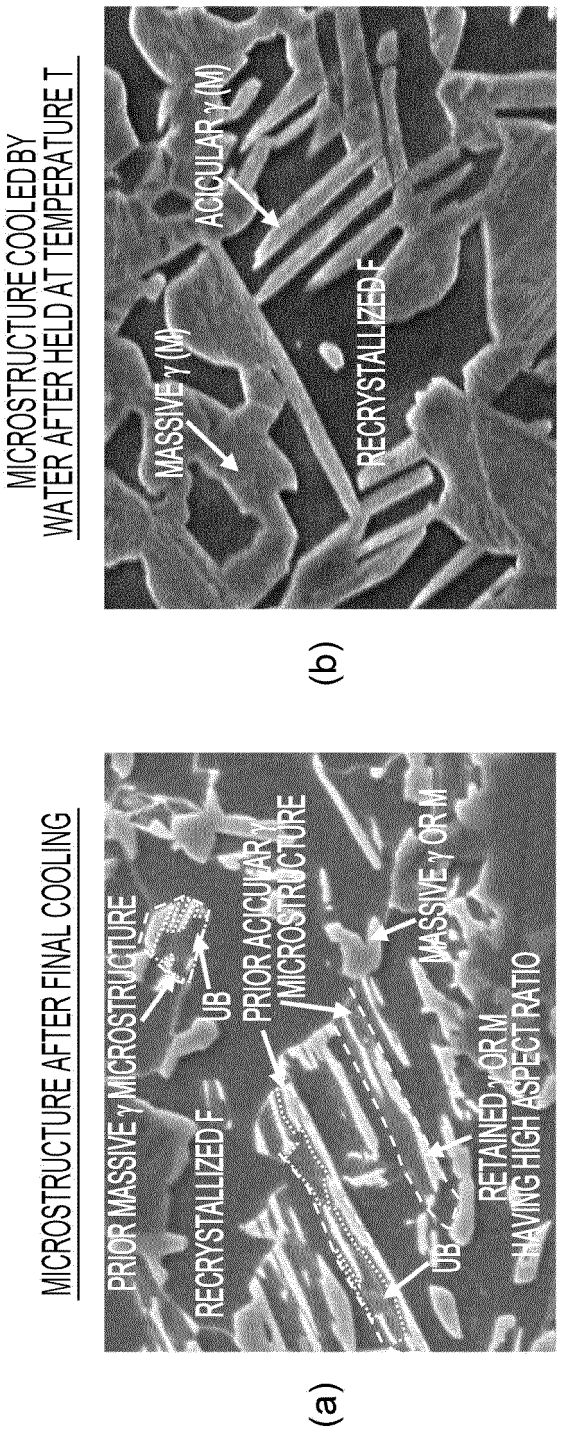
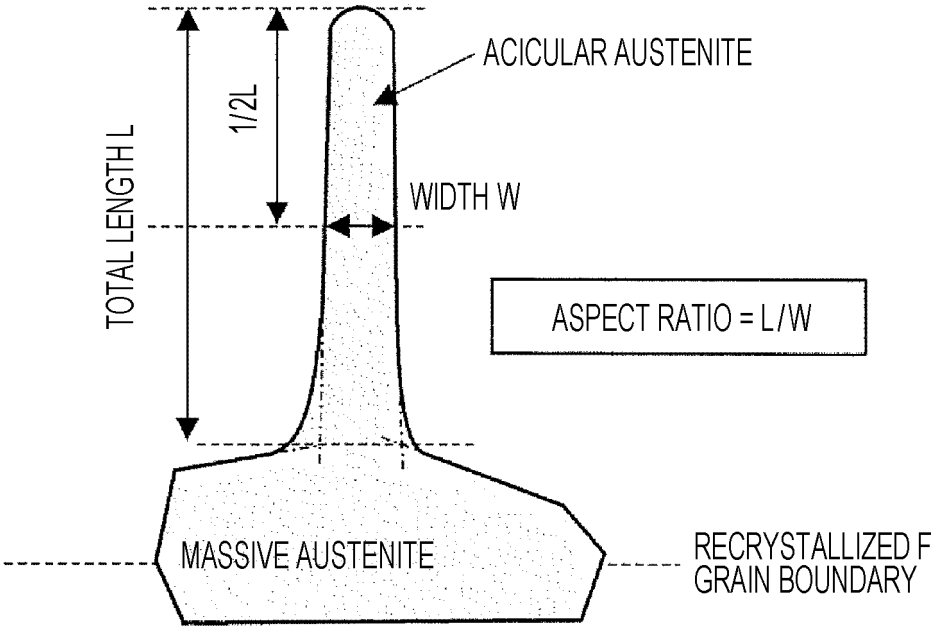

MICROSTRUCTURE AFTER FINAL COOLING
PRIOR MASSIVE γ MICROSTRUCTURE
UB
PRIOR ACICULAR γ MICROSTRUCTURE
RECRYSTALLIZED F
MASSIVE γ OR M
UB
RETAINED γ OR M HAVING HIGH ASPECT RATIO
(a)
MICROSTRUCTURE COOLED BY WATER AFTER HELD AT TEMPERATURE T
ACICULAR γ (M)
MASSIVE γ (M)
RECRYSTALLIZED F
(b)
5μm

FIG. 2


ACICULAR AUSTENITE
1/2L
TOTAL LENGTH L
WIDTH W
ASPECT RATIO = L/W
MASSIVE AUSTENITE
RECRYSTALLIZED F
GRAIN BOUNDARY

INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2022/033945

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 301S; C21D9/46 G; 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; C21D8/00-8/04; C21D9/46-9/48

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 1971-2022
Registered utility model specifications of Japan 1996-2022
Published registered utility model applications of Japan 1994-2022

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 2019-502819 A (POSCO) 31 January 2019 (2019-01-31) claims, paragraphs [0034]-[0036]	1-6
A	WO 2021/079756 A1 (JFE STEEL CORP.) 29 April 2021 (2021-04-29) claims	1-6
A	WO 2009/099251 A1 (JFE STEEL CORP.) 13 August 2009 (2009-08-13) claims	1-6

☐ 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	Date of mailing of the international search report
17 November 2022	**29 November 2022**
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.

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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/JP2022/033945

Patent document cited in search report	Publication date (day/month/year)	Patent family member(s)	Publication date (day/month/year)
JP 2019-502819 A	31 January 2019	US 2018/0355453 A1 claims, paragraphs [0068]-[0071] CN 108463570 A WO 2017/105026 A1	
WO 2021/079756 A1	29 April 2021	EP 4029957 A1 claims	
WO 2009/099251 A1	13 August 2009	EP 2243852 A1 claims US 2011/0036465 A1 CN 101939457 A	

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

REFERENCES CITED IN THE DESCRIPTION

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Patent documents cited in the description

- JP 6035619 B **[0008]**
- JP 4411221 B **[0008]**
- JP 6041079 B **[0008]**