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(54) **THIN STEEL SHEET AND PROCESS FOR PRODUCING SAME**

(57) A thin steel sheet having sheet thickness ≤ 1.6 mm, but tensile strength ≥ 780 MPa and Young's modulus ≥ 240 GPa in transverse direction is provided, where the steel sheet has composition including, in mass%, C: 0.06-0.12 %, Si: 0.5-1.5 %, Mn: 1.0-3.0 %, P: 0.05 % or less, S: 0.01 % or less, Al: 0.5 % or less, N: 0.01 % or less, Ti: 0.02-0.20 %, and the balance being Fe and incidental impurities, where the composition satisfies relations of Formula (1) and (2), and microstructure such that ferrite phase has area ratio ≥ 60 % and martensite phase has area ratio of 15-35 %, ferrite and martensite phases are 95 % or more in total, average grain size of ferrite is ≤ 4.0 μm and that of martensite is ≤ 1.5 μm,

$$0.11 \leq [\%C]-(12/47.9)\times[\%Ti^{*}] \leq 0.15 \quad (1)$$

, where

$$Ti^{*}=[\%Ti]-(47.9/14)\times[\%N]-(47.9/32.1)\times[\%S] \quad (2).$$

EP 2 781 615 A1

Description

TECHNICAL FIELD

[0001] The present invention relates to a high-strength thin steel sheet having excellent rigidity that is preferably and mainly used for automobile body parts, and a method for manufacturing the same. The high-strength thin steel sheet of the present invention, which is preferably applicable as structural members having a columnar or nearly columnar cross-sectional shape with a rigid sensitivity index of sheet thickness of approximately 1, such as center pillars, side sills, side frames and cross members of automobiles, has a tensile strength of 780 MPa or higher and shows excellent ductility.

BACKGROUND ART

[0002] In recent years, responding to increasing public concern about global environment issues, for example, emission regulations have been implemented for automobiles and it has been a critical issue to reduce the weight of automobile body. As such, efforts have been made to reduce the weight of body by strengthening steel sheets to reduce the sheet thickness. Currently, as a result of remarkable advances in strengthening steel sheets, there is increased use of steel sheets having a sheet thickness of less than 1.6 mm. Particularly, since steel sheets in 780 MPa and 980 MPa grades of tensile strength have been used at increasing proportions every year, it is essential to prevent a decrease in rigidity of parts due to reduced thickness at the same time in order to achieve such weight reduction through strengthening of steel sheets. The problem associated with a decrease in rigidity of parts due to reduced thickness of steel sheets becomes apparent in those steel sheets having a tensile strength of 590 MPa or higher.

[0003] Generally, in order to increase the rigidity of parts, it is thought to be effective to change the shape of parts, or alternatively, for those parts being subjected to spot welding, to change the welding conditions, such as increasing the number of welding points or switching to laser welding. However, when used as automobile parts, there is a problem that it is not easy to change the shape of the parts in a limited space in an automobile, and changes to the welding conditions are made at the expense of an increase in cost, and so on.

[0004] In view of the foregoing, to increase the rigidity of parts without changing the shape or welding conditions of the parts, it is effective to increase Young's modulus of members used for these parts. In the case of steel having body-centered cubic lattice, it is known that Young's modulus, which is strongly dependent on texture, has the highest value in <111> direction in which atoms are most densely packed, while having the smallest value in <100> direction in which atoms are less dense. It is widely known that Young's modulus of normal iron which is less anisotropic in crystal orientation is approximately 210 GPa. However, if the crystal orientation is anisotropic and the atomic density can be increased in a particular direction, Young's modulus can be increased in that direction.

[0005] Conventionally, as for Young's modulus of steel sheets, various considerations have been given to increasing Young's modulus in a particular direction by controlling texture.

For example, JP 5-255804 A (PTL 1) discloses a technique that uses steel resulting from adding Nb or Ti to ultra low carbon steel and involves controlling, in a hot rolling step, the rolling reduction ratio to be 85 % or more in a temperature range of Ar_3 to (Ar_3 + 150 °C) and thereby facilitating transformation of non-recrystallized austenite to ferrite, so that ferrite in {311}<011> and {332}<113> orientations is allowed to grow at the stage of hot-rolled sheet, and Young's modulus is increased in a direction perpendicular to the rolling direction through the subsequent cold rolling and recrystallization annealing whereby {211}<011> orientation is made into the primary orientation.

[0006] In addition, JP 8-311541 A (PTL 2) discloses a method for manufacturing a hot-rolled steel sheet with an increased Young's modulus by adding Nb, Mo and B to low carbon steel having C content of 0.02 % to 0.15 % and controlling the rolling reduction ratio to be 50 % or more in a temperature range of Ar_3 to 950 °C, thereby causing growth in {211}<011> orientation.

[0007] Further, JP 2006-183131 A (PTL 3) and JP 2005-314792 A (PTL4) disclose techniques that use steel resulting from adding Nb to low carbon steel, define the content of C that is not fixed as carbonitride, and involve controlling, in a hot rolling step, the total rolling reduction ratio to be 30 % or more at 950 °C or lower to facilitate transformation of non-recrystallized austenite to ferrite so that ferrite in {113}<110> orientation is allowed to grow at the stage of hot-rolled sheet and Young's modulus is increased in a direction perpendicular to the rolling direction through the subsequent cold rolling and recrystallization annealing whereby {112}<110> orientation is made into the primary orientation.

CITATION LIST

Patent Literature

[0008]

PTL 1: JP 5-255804 A
PTL 2: JP 8-311541 A
PTL 3: JP 2006-183131 A
PTL 4: JP 2005-314792 A

SUMMARY OF INVENTION

(Technical Problem)

[0009] However, the above-mentioned conventional techniques have the following problems.

[0010] That is, while the technique disclosed in PTL 1 increases Young's modulus of a steel sheet by using ultra low carbon steel having C content of 0.01 % or less and controlling its texture, the obtained tensile strength is about 450 MPa at most. Thus, there was a limit to further strengthening by applying this technique.

[0011] The technique disclosed in PTL 2 has a problem that it cannot utilize texture control by cold working since the target steel sheet is a hot-rolled steel sheet, where it is difficult not only to achieve even higher Young's modulus, but also to manufacture such a high strength steel sheet that has a sheet thickness of less than 2.0 mm in a stable manner by low temperature finish rolling.

[0012] While the technique disclosed in PTL 3 increases tensile strength by increasing the amount of alloying elements to be added and increasing the fraction of martensite, it was difficult to improve workability while enhancing the strength, because total elongation is decreased and strength-elongation balance (TS x El) is distorted as well.
In addition, while the techniques disclosed in PTL 3 and PTL4 increase Young's modulus by controlling the total rolling reduction ratio at 950 °C or lower to be 30 % or more in the hot rolling step, these techniques suffered a problem that it was difficult to maintain a total rolling reduction ratio of 30 % or more due to high rolling load in a temperature range of 950 °C or lower.

[0013] As such, the conventional techniques are directed to increasing Young's modulus of steel sheets, such as hot-rolled steel sheets or mild steel sheets, having a large sheet thickness, materials having high strength but poor ductility, or materials difficult to produce. Thus, it was difficult to provide a high strength steel sheet, which has a sheet thickness of 1.6 mm or less and TS of 780 MPa or higher, with both higher ductility and higher Young's modulus by using such conventional techniques.

[0014] The above-described problem is solved by the present invention. An object of the present invention is to provide a high-strength thin steel sheet having excellent rigidity that has a sheet thickness as small as 1.6 mm or less, but a tensile strength as high as 780 MPa or higher, more preferably 980 MPa or higher, in a transverse direction perpendicular to the rolling direction (hereinafter, also referred to as the "transverse direction"), and satisfies a condition that Young's modulus in the transverse direction is 240 GPa or higher, as well as an advantageous method for manufacturing the same.

(Solution to Problem)

[0015] In the case of normal steel having body-centered cubic lattice, Young's modulus of the steel, which is largely dependent on texture, is high in <111> direction in which atoms are most densely packed, while being low in <100> direction in which atoms are less dense. Accordingly, growth in (112)[1-10] orientation brings about alignment of <111> direction with the transverse direction of the steel sheet. It is thus possible to increase Young's modulus in this direction.
In addition, there are various methods for strengthening steel. For example, Dual-Phase (DP) steel, in which a soft ferrite phase is strengthened with a hard martensite phase, is known to have generally good ductility. However, in ultra high strength steel having TS of 780 MPa or higher, the volume fraction of martensite phase tends to increase in general, which results in not only deterioration in ductility, but also difficulty in causing growth in (112)[1-10] orientation, which is effective in increasing Young's modulus in the transverse direction.

[0016] To solve the above-described problem, as a result of studies on Young's modulus of a high-strength thin steel sheet having TS of 780 MPa or higher in a direction perpendicular to the rolling direction, it was found that it is possible to keep the volume fraction of martensite low even in a steel sheet having ultra high strength of TS of 780 MPa or higher by solid solution strengthening, grain refinement strengthening and precipitation strengthening, and to balance high ductility, high strength and high rigidity by increasing the accumulation of ferrite in (112)[1-10] orientation.
The present invention is based on the above-mentioned findings.

[0017] That is, the arrangement of the present invention is summarized as follows:

[1] A thin steel sheet having a composition including, in mass%, C: 0.06 % to 0.12 %, Si: 0.5 % to 1.5 %, Mn: 1.0 % to 3.0 %, P: 0.05 % or less, S: 0.01 % or less, Al: 0.5 % or less, N: 0.01 % or less, Ti: 0.02 % to 0.20 %, and the balance being Fe and incidental impurities, where the composition satisfies relations of Formula (1) and (2) below, wherein the steel sheet has a microstructure such that a ferrite phase has an area ratio of 60 % or more and a

martensite phase has an area ratio of 15 % to 35 %, where a total of the ferrite phase and the martensite phase is 95 % or more, an average grain size of ferrite is 4.0 μm or less and an average grain size of martensite is 1.5 μm or less, wherein the steel sheet has a tensile strength (TS) of 780 MPa or higher in a transverse direction perpendicular to the rolling direction, Young's modulus of 240 GPa or higher in the transverse direction, and strength-elongation balance (TS x El) of 16500 MPa·% or more in the transverse direction, the strength-elongation balance being expressed by a product of the tensile strength (TS) and total elongation (El),

$$0.05 \leq [\%C] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (1)$$

, where

$$Ti^* = [\%Ti] - (47.9/14) \times [\%N] - (47.9/32.1) \times [\%S] \text{ ----- } (2),$$

and
[%M] indicates the content (mass %) of M element.
[2] The thin steel sheet according to item [1] above, wherein the composition of the steel sheet further includes, in mass%, Nb: 0.02 % to 0.10 %, and satisfies a relation of Formula (3) below in place of the Formula (1):

$$0.05 \leq [\%C] - (12/92.9) \times [\%Nb] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (3).$$

[3] The thin steel sheet according to item [1] or [2] above, wherein the composition of the steel sheet further includes, in mass%, one or more elements selected from Cr: 0.1 % to 1.0 %, Ni: 0.1 % to 1.0 %, Mo: 0.1 % to 1.0 %, Cu: 0.1 % to 2.0 % and B: 0.0005 % to 0.0030 %.
[4] A method for manufacturing a thin steel sheet, the method comprising:

in a hot rolling process, subjecting a steel material to finish rolling and completing the finish rolling at 850 °C to 950 °C to obtain a hot-rolled steel sheet, the steel material having a composition including, in mass%, C: 0.06 % to 0.12 %, Si: 0.5 % to 1.5 %, Mn: 1.0 % to 3.0 %, P: 0.05 % or less, S: 0.01 % or less, Al: 0.5 % or less, N: 0.01 % or less, Ti: 0.02 % to 0.20 %, and the balance being Fe and incidental impurities, where the contents of C, N, S and Ti satisfy relations of Formula (1) and (2) below;
then coiling the steel sheet at 650 °C or lower;
subjecting the steel sheet to pickling; and
then subjecting the steel sheet to cold rolling at a rolling reduction ratio of 60 % or more;
in a subsequent annealing process, heating the steel sheet to a soaking temperature of 780 °C to 880 °C at an average heating rate from (Ac_1 - 100 °C) to Ac_1 of 15 °C/s or higher;
holding the steel sheet at the soaking temperature for 150 seconds or less; and
cooling the steel sheet to 350 °C or lower at an average cooling rate until at least 350 °C of 5 °C/s to 50 °C/s,

$$0.05 \leq [\%C] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (1)$$

, where

$$Ti^* = [\%Ti] - (47.9/14) \times [\%N] - (47.9/32.1) \times [\%S] \text{ ----- } (2),$$

and
[%M] indicates the content (mass %) of M element.

[5] The method for manufacturing a thin steel sheet according to item [4] above, wherein the composition of the steel material further includes, in mass%, Nb: 0.02 % to 0.10 %, and satisfies a relation of Formula (3) below in place of the Formula (1):

$$0.05 \leq [\%C] - (12/92.9) \times [\%Nb] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (3).$$

[6] The method for manufacturing a thin steel sheet according to item [4] or [5] above, wherein the composition of the steel material further includes, in mass%, one or more elements selected from Cr: 0.1 % to 1.0 %, Ni: 0.1 % to 1.0 %, Mo: 0.1 % to 1.0 %, Cu: 0.1 % to 2.0 % and B: 0.0005 % to 0.0030 %.

(Advantageous Effect of Invention)

[0018] According to the present invention, a high-strength thin steel sheet may be obtained that satisfies the conditions of a tensile strength of 780 MPa or higher, more preferably 980 MPa or higher, in the transverse direction and Young's modulus of 240 GPa or higher, more preferably 245 GPa or higher, in the transverse direction, and furthermore, TS × El = 16500 or more in the transverse direction.

DESCRIPTION OF EMBODIMENTS

[0019] The present invention will be specifically described below. Firstly, the reason why the chemical composition of a steel sheet in the present invention is limited to the above-described range will be described below.
In addition, although the unit of content of each element included in the chemical composition of the steel sheet is "mass%," it will be simply expressed by "%," unless otherwise specified.

0.06 % ≤ C ≤ 0.12 %

[0020] C is an element that stabilizes austenite and may improve quench hardenability and greatly facilitate formation of a low temperature transformation phase during a cooling step at the time of annealing after cold rolling, thereby making a significant contribution to enhancement of strength. To obtain this effect, C content needs be 0.06 % or more, more preferably 0.08 % or more. On the other hand, C content exceeding 0.12 % leads to an increased volume fraction of a hard, low temperature transformation phase, which results in not only an excessive increase in strength of steel, but also a deterioration in workability. In addition, such high C content inhibits recrystallization in an orientation in which Young's modulus is advantageously improved in an annealing process after cold rolling. Further, such high C content also leads to a deterioration in weldability. Thus, C content should be not more than 0.12 %.

0.5 % ≤ Si ≤ 1.5 %

[0021] Si is one of the important elements in the present invention. Since Si raises the Ar_3 transformation point in hot rolling, it facilitates recrystallization of worked austenite when rolling is performed at a temperature immediately above Ar_3. Thus, if Si content is excessively high exceeding 1.5 %, a crystal orientation necessary for increasing Young's modulus can no longer be obtained. Moreover, addition of a large amount of Si not only deteriorates weldability of a steel sheet, but also advances formation of fayalite on a surface of a slab during heating in a hot rolling process, thereby facilitating the occurrence of a surface pattern, which is referred to as so-called red scales. Further, when a steel sheet is used as a cold-rolled steel sheet, oxides of Si that are generated on a surface of the steel sheet deteriorates chemical convertibility, or alternatively, when a steel sheet is used as a hot-dip galvanized steel sheet, oxides of Si that are generated on a surface of the steel sheet induces absence of zinc coating. Thus, Si content should be not more than 1.5 %. In addition, in the case of a steel sheet requiring surface texture or a hot-dip galvanized steel sheet, Si content is preferably 1.2 % or less. On the other hand, Si is an element that stabilizes ferrite and is able to stabilize austenite and facilitate formation of a low temperature transformation phase by facilitating transformation to ferrite and concentrating C in austenite during a cooling step subsequent to soaking in two-phase region in an annealing process after cold rolling. Further, Si may enhance the strength of steel by solid solution strengthening. To obtain this effect, Si content should be 0.5 % or more, preferably 0.7 % or more.

1.0 % ≤ Mn ≤ 3.0 %

[0022] Mn is also one of the important elements in the present invention. Mn is an austenite-stabilizing element that may, during a heating step in an annealing process after cold rolling, lower the Ac_1 transformation point, facilitate transformation of non-recrystallized ferrite to austenite, and allow a low temperature transformation phase that is formed during a cooling step after soaking to grow in an orientation in which Young's modulus is advantageously improved, thereby inhibiting a decrease in Young's modulus associated with the formation of the low temperature transformation phase.

[0023] Mn may also improve quench hardenability and greatly facilitate formation of a low temperature transformation phase during a cooling step after soaking annealing in an annealing process, thereby making a significant contribution to enhancement of strength. Further, Mn acts as a solid-solution-strengthening element, which also contributes to enhancement of strength of steel. To obtain this effect, Mn content should be 1.0 % or more.

[0024] On the other hand, high Mn content exceeding 3.0 % severely inhibits formation of ferrite during cooling after annealing, and even higher Mn content would also deteriorate weldability of the steel sheet. Thus, Mn content is to be 3.0 % or less, more preferably 2.5 % or less.

P ≤ 0.05 %

[0025] P is an element that segregates at grain boundaries, which results in a deterioration in not only ductility and toughness, but also in weldability of a steel sheet. In addition, P causes an inconvenience that alloying is delayed when the steel sheet is used as a hot-dip galvannealed steel sheet. Thus, P content is to be 0.05 % or less.

S ≤ 0.01%

[0026] S is an element that significantly reduces ductility in hot rolling to induce hot cracking, and severely deteriorates surface texture. In addition, it is desirable to minimize S content because S deteriorates ductility and hole expansion formability by forming coarse MnS as an impurity element. These problems become more pronounced when S content exceeds 0.01 %. Thus, S content is to be 0.01 % or less. From the viewpoint of improvement of particularly hole expansion formability, S content is preferably 0.005 % or less.

Al ≤ 0.5 %

[0027] Al is a ferrite-stabilizing element that significantly raises the Ac_3 point in annealing and thus inhibits transformation of non-recrystallized ferrite to austenite, thereby interfering with the growth in an orientation in which Young's modulus is advantageously improved when ferrite is generated from austenite during cooling. Thus, Al content is to be 0.5 % or less, preferably 0.1 % or less. On the other hand, since Al is useful as a deoxidation element of steel, Al content is preferably 0.01 % or more.

N ≤ 0.01 %

[0028] High N content brings about slab cracking during hot rolling and may cause surface defects. Thus, N content should be 0.01 % or less.

0.02 % ≤ **Ti** ≤ **0.20** %

[0029] Ti is the most important element in the present invention. That is, Ti inhibits recrystallization of worked ferrite during a heating step in an annealing process so that transformation of non-recrystallized ferrite to austenite is facilitated, while allowing growth of ferrite, which is generated during a cooling step after annealing, in an orientation in which Young's modulus is advantageously improved. In addition, fine precipitates of Ti contribute to enhancement of strength, and furthermore, have an advantageous effect on refinement of ferrite and martensite. To obtain this effect, Ti content should be 0.02 % or more, preferably 0.04 % or more.

[0030] On the other hand, addition of a large amount of Ti results in not all of carbonitrides being dissolved during reheating in a normal hot rolling process and coarse carbonitrides being left, thereby impeding rather than improving the effects of enhancing strength and inhibiting recrystallization. In addition, even if hot rolling is initiated directly after continuous casting of a slab without subjecting the slab to cooling and subsequent reheating after the continuous casting, the amount of Ti added exceeding 0.20 % only makes a small contribution to the effects of enhancing strength and inhibiting recrystallization, and furthermore, leads to an increase in alloy cost. Thus, Ti content should be 0.20 % or less.

[0031] While the basic elements of the present invention have been described, it is not sufficient to only satisfy the above-described basic elements in the present invention. Rather, regarding the contents of C, N, S and Ti, it is also necessary to satisfy the following relations of Formula (1) and (2):

$$0.05 \leq [\%C] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (1)$$

, where

$$Ti^* = [\%Ti] - (47.9/14) \times [\%N] - (47.9/32.1) \times [\%S] \text{ ----- } (2),$$

and
[%M] indicates the content (mass %) of M element.

[0032] The above relations define the amount of C that is not fixed as carbide. However, if a large amount of C that is not fixed as carbide is present exceeding 0.10 %, the volume fraction of martensite increases and Young's modulus decreases, and furthermore, ductility deteriorates. Thus, the amount of C that is not fixed as carbide, as calculated by Formula (1), should be not more than 0.10 %, preferably not more than 0.09 %. However, if the amount of C that is not fixed as carbide is as small as less than 0.05 %, then the amount of C in austenite decreases during annealing in two-phase region after cold rolling, and furthermore, there will be a reduced amount of martensite phase generated after cooling, which makes it difficult to enhance strength to 780 MPa or higher. Thus, the amount of C that is not fixed as carbide should be not less than 0.05 %, preferably not less than 0.06 %.

[0033] The present invention may also contain the following elements as appropriate.

0.02 % ≤ Nb ≤ 0.10 %

[0034] Similar to Ti, Nb is also an important element of the present invention. Nb inhibits recrystallization of worked ferrite during a heating step in an annealing process after cold rolling so that transformation of non-recrystallized ferrite to austenite is facilitated and coarsening of austenite grains is inhibited, while allowing growth of ferrite, which is generated during a cooling step after annealing soaking, in an orientation in which Young's modulus is advantageously improved. Further, fine carbonitrides of Nb effectively contribute to enhancement of strength, and furthermore, have an advantageous effect on refinement of ferrite and martensite. To obtain this effect, Nb content is preferably 0.02 % or more.

[0035] However, addition of a large amount of Nb results in not all of carbonitrides being dissolved during reheating in a normal hot rolling process and coarse carbonitrides being left, thereby blocking the effects of inhibiting recrystallization of worked austenite in a hot rolling process and inhibiting recrystallization of worked ferrite in an annealing process after cold rolling. In addition, even if hot rolling is initiated directly after continuous casting of a slab without subjecting the slab to cooling and subsequent reheating after continuous casting, the amount of Nb added exceeding 0.10 % only makes a small contribution to the effect of inhibiting recrystallization, and furthermore, leads to an increase in alloy cost. Thus, Nb content is preferably not more than 0.10 %, more preferably not more than 0.08 %.

[0036] In addition, if Nb is also contained along with Ti, a relation of Formula (3) below, in place of Formula (1) above, is satisfied:

$$0.05 \leq [\%C] - (12/92.9) \times [\%Nb] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (3).$$

Nb forms carbide to reduce the amount of C that is not fixed as carbide. Accordingly, to control the amount of C that is not fixed as carbide within a range of 0.05 % to 0.10 %, if Nb is added, the value of ([%C] - (12/92.9) x [%Nb] - (12/47.9) × [%Ti*]) is controlled within a range of 0.05 % to 0.10 %, preferably 0.06 % to 0.09 %.

0.1 % ≤ Cr ≤ 1.0 %

[0037] Cr is an element that inhibits formation of cementite, thereby improving quench hardenability. Cr has an effect of greatly facilitating formation of martensite phase during a cooling step after soaking in an annealing process. To obtain this effect, Cr content is preferably 0.1 % or more. However, if a large amount of Cr is added, the effect attained by the addition will be saturated and alloy cost will also increase. Thus, Cr is preferably added in an amount of 1.0 % or less. In addition, when a steel sheet is used as a hot-dip galvanized steel sheet, oxides of Cr formed on a surface of the steel sheet induces absence of zinc coating. Thus, Cr content is preferably 0.5 % or less.

0.1 % ≤ Ni ≤ 1.0 %

[0038] Ni is an element that improves quench hardenability and may facilitate formation of martensite phase during a cooling step after soaking in an annealing process. In addition, Ni effectively contributes to enhancement of strength of steel as a solid-solution-strengthening element. Further, in the case of Cu-added steel, surface defects are induced during hot rolling due to cracking associated with a reduction in hot ductility. However, it is possible to inhibit occurrence of such surface defects by containing Ni in combination with Cu. To obtain this effect, Ni content is preferably 0.1 % or more. However, addition of a large amount of Ni interferes with formation of ferrite, which is necessary for increasing Young's modulus, during a cooling step after soaking, and furthermore, results in an increase in alloy cost. Thus, Ni content is preferably 1.0 % or less.

0.1 % ≤ Mo ≤ 1.0 %

[0039] Mo is an element that improves quench hardenability and may facilitate formation of martensite phase during a cooling step after soaking in an annealing process, thereby contributing to enhancement of strength. To obtain this effect, Mo content is preferably 0.1 % or more. However, if a large amount of Mo is added, the effect attained by the addition will be saturated at some point and alloy cost will also increase. Thus, Mo content is preferably 1.0 % or less, more preferably 0.5 % or less.

0.1 % ≤ Cu ≤ 2.0 %

[0040] Cu is an element that improves quench hardenability and facilitates formation of martensite phase during a cooling step after soaking in an annealing process, thereby contributing to enhancement of strength. To obtain this effect, Cu content is preferably 0.1 % or more. However, excessive addition of Cu deteriorates hot ductility and induces surface defects associated with cracking during hot rolling. Thus, Cu content is preferably 2.0 % or less.

0.0005 % ≤ B ≤ 0.0030 %

[0041] B is an element that improves quench hardenability by inhibiting transformation from austenite to ferrite and facilitates formation of martensite during a cooling step after soaking in an annealing process, thereby contributing to enhancement of strength.

[0042] To obtain this effect, B content is preferably 0.0005 % or more. However, excessive addition of B severely interferes with formation of ferrite during cooling after soaking and reduces Young's modulus. Thus, B content is preferably 0.0030 % or less.

[0043] Reasons for limitations on the microstructure of the present invention will now be described below.

[0044] The steel sheet of the present invention has a microstructure in which ferrite phase is the primary phase, including, in area ratio, 60 % or more of ferrite phase and 15 % to 35 % of martensite phase.

[0045] The area ratio of ferrite phase should be 60 % or more since ferrite phase is effective in causing growth of texture which is advantageous for improving Young's modulus. In addition, since strength as well as strength-elongation balance improve by containing martensite phase, the area ratio of martensite phase should be 15 % or more. However, if the area ratio of martensite phase exceeds 35 %, it is not possible to ensure appropriate Young's modulus in the transverse direction. Thus, the area ratio of martensite phase should be not more than 35 %. Further, in order to improve strength-elongation balance, a total of the area ratios of ferrite phase and martensite phase should be 95 % or more.

[0046] Phases other than the ferrite phase and the martensite phase may include pearlite, bainite and cementite phases, which are not problematic if contained in an amount of not more than 5 %, preferably not more than 3 %, more preferably not more than 1 %.

[0047] In addition, an average grain size of ferrite exceeding 4.0 μm leads to a reduction in strength, which necessitates increasing the volume fraction of martensite phase and adding more elements, and results in a decrease in Young's modulus and an increase in manufacturing cost. Thus, average grain size of ferrite should be 4.0 μm or less. Particularly, in order to satisfy a tensile strength of 780 MPa or higher in a stable manner, the average grain size of ferrite is preferably 3.5 μm or less.
Moreover, an average grain size of martensite exceeding 1.5 μm increases the potential of progress in void linking upon working/deformation, which results in a reduction in ductility of the steel sheet. Thus, the average grain size of martensite should be not more than 1.5 μm, more preferably not more than 1.0 p,m.

[0048] Area ratios of ferrite phase and martensite phase were determined by subjecting a cross-section of the steel sheet to nital etching, observing the cross-section with scanning electron microscope (SEM), taking three images of 25 μm × 30 μm regions, analyzing these images by image processing and measuring the areas of ferrite phase and martensite phase. In addition, based on the SEM images, the average grain size was calculated by dividing a total of respective areas of ferrite phase and martensite phase within the field of view by the number of grains in these phases to determine an average area of the grains, the value of which average area is then raised to the power of 1/2.

[0049] With the above-described chemical composition and microstructure, it is possible to obtain a high-strength thin steel sheet having excellent rigidity that has a tensile strength (TS) of 780 MPa or higher in the transverse direction, Young's modulus of 240 GPa or higher in the transverse direction and strength-elongation balance (TS x El) of 16500 or more in the transverse direction.

[0050] A preferred method for manufacturing the steel sheet of the present invention will now be described below.

[0051] In manufacturing the steel sheet of the present invention, steel having a chemical composition in accordance with the above-described composition is prepared by steelmaking, depending on the target strength level. Any appropriate steelmaking process may be applied, such as normal converter steelmaking process or electric furnace steelmaking process. The steel prepared by steelmaking is cast into a slab, which in turn is directly subjected to hot rolling, or alternatively subjected to cooling and subsequent heating before hot rolling, under a condition of finisher delivery temperature of 850 °C to 950 °C to obtain a hot-rolled sheet. Then, the sheet is subjected to coiling at 650 °C or lower, followed by pickling and subsequent cold rolling at a rolling reduction ratio of 60 % or more. Thereafter, in an annealing step, the sheet is heated at an average heating rate of 15 °C/s or higher within a temperature range of (Ac_1 - 100 °C) to Ac_1, held at a soaking temperature of 780 °C to 880 °C for a duration of 150 seconds or less, and then cooled to 350 °C or lower at an average cooling rate until at least 350 °C of 5 °C/s to 50 °C/s.

[0052] In the following, reasons for the above-described limitations on the manufacturing conditions will be described.

[Finisher Delivery Temperature: 850 °C to 950 °C]

[0053] By controlling finisher delivery temperature to be 950 °C or lower, transformation from non-recrystallized austenite to ferrite advances to provide fine ferrite phase, and furthermore, the degree of accumulation of the crystal grains in (112)[1-10] orientation may be increased through cold rolling and annealing. However, if the finisher delivery temperature is below 850 °C, it may more likely fall below the Ar_3 transformation point, which results in mixing hot-rolled phase with worked phase, thereby disturbing accumulation in (112)[1-10] orientation after the cold rolling and annealing. This also poses difficulty in manufacture, such as a significant increase in rolling load due to increased transformation resistance. Thus, the finisher delivery temperature should be within a range of 850 °C to 950 °C.

[Coiling Temperature: 650 °C or lower]

[0054] If the coiling temperature after the finish rolling exceeds 650 °C, carbonitrides of Ti and Nb coarsen and thus the effects of inhibiting recrystallization of ferrite and inhibiting coarsening of austenite grains are weakened during heating stage of the annealing process after the cold rolling. Thus, the coiling temperature is to be not higher than 650 °C. On the other hand, if the coiling temperature is lower than 400 °C, many hard, low temperature transformation phases are generated, which causes non-uniform deformation in the subsequent cold rolling, thereby disturbing accumulation in an orientation in which Young's modulus is advantageously improved. This results in no growth of the texture after annealing, which makes it difficult to improve Young's modulus. Further, in view of an increase in load during cold rolling after the coiling, the coiling temperature is preferably not lower than 400 °C.

[Rolling Reduction Ratio during Cold Rolling: 60 % or more]

[0055] After the above-described coiling, the sheet is subjected to pickling, followed by cold rolling at a rolling reduction ratio of 60 % or more. This cold rolling causes accumulation in (112)[1-10] orientation in which Young's modulus is effectively improved. That is, growth in (112)[1-10] orientation is caused by cold rolling to provide more ferrite grains having (112)[1-10] orientation even in the microstructure after the subsequent annealing process and improve Young's modulus. To obtain this effect, the rolling reduction ratio during cold rolling should be 60 % or more, more preferably 65 % or more. However, the rolling load becomes larger with higher rolling reduction ratio during cold rolling, so that manufacture becomes more difficult. Thus, the upper limit of the rolling reduction ratio during cold rolling is preferably 85 %.

[Average Heating Rate from (Ac) - 100 °C) to Ac_1: 15 °C/s or higher]

[0056] To improve Young's modulus of the steel sheet after annealing, it is necessary to inhibit, during a heating step in annealing, recrystallization of ferrite that has grown during cold rolling and has (112)[1-10] orientation and to cause transformation from worked ferrite to austenite. To this end, average heating rate should be 15 **°C/s** or higher.

[0057] As used herein, Ac_1 is Ac_1 transformation temperature that is determined by Formula (4) below based on the contents of C, Si, Mn, Al, Ni, Cr, Cu, Mo, Ti, Nb and B expressed in mass%:

$$Ac_1 = 750.8 - 26.6\ [\%C] + 17.6\ [\%Si] - 11.6\ [\%Mn] - 169.4\ [\%Al] - 23.0\ [\%Ni] + 24.1\ [\%Cr] - 22.9\ [\%Cu] + 22.5\ [\%Mo] - 5.7\ [\%Ti] + 232.6\ [\%Nb] - 894.7\ [\%B] \text{ ----- } (4)$$

, where [%M] indicates the content (mass%) of M element.

[Soaking Temperature: 780 °C to 880 °C, Soaking Duration: 150 seconds or less]

[0058] During soaking in the annealing process, a sufficient amount of ferrite transforms to austenite, which in turn transforms again to ferrite during cooling. This allows growth of the texture, thereby improving Young's modulus. In addition, if the soaking temperature is low, rolled textures remain and elongation decreases. Thus, the soaking temperature should be 780 °C or higher. However, an excessively high soaking temperature coarsens austenite grains, which makes it difficult for ferrite, which results from the retransformation during cooling after annealing, to accumulate in (112)[1-10] orientation. Thus, the soaking temperature should be 880 °C or lower.
[0059] In addition, coarsening of austenite grains is also caused by holding at this temperature range for a long duration. Thus, the soaking duration should be 150 seconds or less. On the other hand, to prevent remaining of the rolled texture and to improve elongation, the soaking duration is preferably 15 seconds or more.

[Average Cooling Rate from Soaking Temperature to at least 350 °C: 5 °C/s to 50 °C/s]

[0060] In the manufacturing method according to the present invention, it is important to control the cooling condition after the above-described soaking treatment.
[0061] That is, formation of ferrite during cooling after soaking allows for growth of texture which is advantageous for improving Young's modulus. Accordingly, ferrite is to be formed at an area ratio of 60 % or more during this cooling step. To this end, the upper limit of the cooling rate should be 50 °C/s. On the other hand, an excessively slow cooling rate hampers formation of martensite. Thus, the cooling rate should be not lower than 5 °C/s, preferably not lower than 10 °C/s.
[0062] In addition, a high cooling stop temperature causes formation of bainite and pearlite instead of martensite, which leads to a reduction in strength and an increase in YS/TS ratio. Alternatively, even if martensite is formed, the hardness of martensite is reduced by tempering during cooling and thus the contribution to enhancement of strength becomes small, which hampers provision of good TS-El balance. Thus, it is necessary to conduct cooling at a predetermined cooling rate until at least 350 °C. Further, for better TS-El balance, it is preferable to conduct cooling at a predetermined cooling rate until at least 300 °C.
[0063] Thereafter, the steel sheet may be subjected to the process where the steel sheet is passed through an overaging zone. In addition, if manufactured as a hot-dip galvanized steel sheet, the steel sheet may be passed through molten zinc, or alternatively, when manufactured as a hot-dip galvannealed steel sheet, the steel sheet may be subjected to an alloying process.
[0064] It should be noted that the steel sheet may be subjected to temper rolling for adjusting the shape of the steel sheet, in which case there is no significant change in Young's modulus or tensile properties if the percent elongation is not more than 0.8 %, preferably not more than 0.6 %.

EXAMPLES

[0065] Examples of the present invention will now be described below. It should be noted that the present invention is not intended to be limited to the disclosed examples.

(Example 1)

[0066] At first, Steel A having a chemical composition as shown in Table 1 was prepared by steelmaking in a vacuum melting furnace. Then, Steel A was subjected to hot rolling, pickling, cold rolling and subsequent annealing to produce a cold-rolled steel sheet. In this case, the following basic conditions were set - heating condition prior to hot rolling: 1250 °C for one hour; finisher delivery temperature of hot rolling: 880 °C; sheet thickness after hot rolling: 4.4 mm; coiling condition: process corresponding to coiling where furnace cooling was conducted after a holding time of one hour at 600 °C; rolling reduction ratio during cold rolling: 68 %, sheet thickness after cold rolling: 1.4 mm, average heating rate from (Ac_1 - 100 °C) to Ac_1: 20 °C/s, duration at soaking temperature of 830 °C: 60 seconds, average cooling rate until 300 °C: 15 °C/s, and subsequent cooling to room temperature: air cooling. These basic conditions are shown in Table 2.
[0067] Further, among these basic conditions, rolling reduction ratio during cold rolling, heating rate from (Ac_1 - 100 °C) to Ac_1, soaking temperature, quench stop temperature and cooling rate to quench stop temperature during the annealing process were changed as shown in Table 3.
[0068] After the above-described annealing, test specimens of 10 mm x 50 mm were cut from the steel sheets in a direction perpendicular to the rolling direction of the steel sheets. Then, a resonance frequency measuring device of lateral vibration type was used to measure Young's modulus (Ec) in accordance with the standard (C1259) of American Society to Testing Materials. In addition, JIS No. 5 tensile test specimens were cut from the cold-rolled steel sheets,

which had been subjected to temper rolling with percent elongation of 0.5 %, in a direction perpendicular to the rolling direction for measuring their tensile properties (tensile strength TS and elongation El).

[0069] It should be noted that the area ratio of ferrite phase (α) and the area ratio of martensite phase (M), as well as the average crystal grain size of each phase were determined by the above-mentioned method.

[0070] The obtained results are shown in Table 2 and Table 3.

[Table 1]

Steel ID	Chemical Composition (mass%)									C* (mass%)	Ac_1 (°C)	Remarks
	C	Si	Mn	P	S	Al	N	Ti	Nb			
A	0.12	1.01	2.01	0.014	0.002	0.04	0.004	0.12	0.03	0.090	741.6	Conforming Steel

C*: Amount of C not fixed as carbide
(C* = [%C] - (12/92.9) x [%Nb] - (12/47.9) x [%Ti*])
Where Ti* = [%Ti] - (47.9/14) x [%N] - (47.9/32.1) x [%S].

[Table 2]

Steel ID	Steel Sheet	Hot Rolling Condition		Cold Rolling Condition		Annealing Condition					Material Microstructure					Material Property					Remarks
		Finisher Delivery Temp. (°C)	Coiling Temp. (°C)	Rolling Reduction Ratio (%)	Sheet Thickness (mm)	Heating Rate from (Ac_1 - 100 °C) to Ac_1 (°C/s)	Soaking Temp. (°C)	Soaking Duration (sec.)	Quench Stop Temp. (°C)	Cooling Rate to Quench Stop Temp. (°C/s)	Ferrite Fraction (%)	Martensite Fraction (%)	Balance (%)	Ferrite Grain Size (μm)	Martensite Grain Size (μm)	YS (MPa)	TS (MPa)	El (%)	Ec (GPa)	TS x El (MPa·%)	
A	A1	880	600	68	1.41	20	830	60	300	15	67	33	0	2.9	0.8	732	1064	16.3	252	17343	Inventive Example

[Table 3]

Steel ID	Steel Sheet	Hot Rolling Condition: Finisher Delivery Temp. (°C)	Hot Rolling Condition: Coiling Temp. (°C)	Cold Rolling Condition: Rolling Reduction Ratio (%)	Cold Rolling Condition: Sheet Thickness (mm)	Annealing Condition: Heating Rate from (Ac1 - 100 °C) to Ac1 (°C/s)	Annealing Condition: Soaking Temp. (°C)	Annealing Condition: Soaking Duration (sec.)	Annealing Condition: Quench Stop Temp. (°C)	Annealing Condition: Cooling Rate to Quench Stop Temp. (°C/s)	Material Microstructure: Ferrite Fraction (%)	Material Microstructure: Martensite Fraction (%)	Material Microstructure: Balance (%)	Material Microstructure: Ferrite Grain Size (μm)	Material Microstructure: Martensite Grain Size (μm)	Material Property: YS (MPa)	Material Property: TS (MPa)	Material Property: El (%)	Material Property: Ec (GPa)	Material Property: TS x El (MPa·%)	Remarks
A	A2	880	600	54	2.02	20	830	60	300	15	68	32	0	3.3	0.9	702	1049	18.1	237	18987	Comparative Example
A	A3	880	600	77	1.01	20	830	60	300	15	65	35	0	2.8	0.8	741	1062	15.8	255	16780	Inventive Example
A	A4	880	600	80	0.88	20	830	60	300	15	65	35	0	2.7	0.8	765	1075	15.4	254	16555	Inventive Example
A	A5	880	600	68	1.41	10	830	60	300	15	66	34	0	4.2	1.7	701	1036	16.6	233	17198	Comparative Example
A	A6	880	600	68	1.41	20	770	60	300	15	89	11	0	3.2	1.4	796	1105	12.7	235	14034	Comparative Example
A	A7	880	600	68	1.41	20	860	60	300	15	66	34	0	3.2	0.9	714	1032	17.3	252	17854	Inventive Example
A	A8	880	600	68	1.41	20	830	60	600	15	85	0	P: 15	3.0	-	473	689	24.5	245	16881	Comparative Example
A	A9	880	600	68	1.41	20	830	60	300	3	87	13	0	4.1	2.3	520	763	23.2	254	17702	Comparative Example
A	A10	880	600	68	1.41	20	830	60	300	150	53	47	0	3.6	1.7	729	1124	12.7	239	14275	Comparative Example

P: Pearlite

[0071] A cold-rolled steel sheet (Steel Sheet A1), which was produced in accordance with the basic conditions, exhibited good strength-elongation balance and high Young's modulus, as shown in Table 2, such that TS: 1064 MPa; El: 16.3 %; TS x El: 17343 MPa·%; Ec: 252 GPa; area ratio of ferrite: 67 %; area ratio of martensite: 33 %; ferrite grain size: 2.9 μm; and martensite grain size: 0.8 μm.

[0072] In addition, even if the rolling reduction ratio during cold rolling and the annealing condition were changed, excellent properties were still obtained in each case where these conditions fall within the scope of the present invention (Steel Sheets A3, A4 and A7), such as TS of 780 MPa or higher, TS x El of 16500 or higher and Ec of 240 GPa or higher.

(Example 2)

[0073] Furthermore, Steels B to N having chemical compositions shown in Table 4 were prepared by steelmaking in a vacuum melting furnace. Then, these steels were subjected to hot rolling, pickling, cold rolling and annealing sequentially under the conditions shown in Table 5.

[0074] The cold-rolled steel sheets thus obtained were analyzed in the same way as described in Example 1. The obtained results are shown in Table 5.

[Table 4]

Steel ID	Chemical Composition (mass%)														C* (mass%)	Ac1 (°C)	Remarks
	C	Si	Mn	P	S	Al	N	Ti	Nb	Ni	Cr	Cu	Mo	B			
B	0.10	0.85	2.03	0.015	0.002	0.04	0.004	0.07	-	-	-	-	-	-	0.087	732.4	Conforming Steel
C	0.06	1.03	2.03	0.014	0.001	0.04	0.004	0.04	0.02	-	-	-	-	-	0.051	741.4	Conforming Steel
D	0.08	1.03	2.02	0.014	0.002	0.04	0.003	0.05	0.03	-	-	-	-	-	0.067	743.3	Conforming Steel
E	0.10	1.04	2.04	0.017	0.001	0.04	0.003	0.11	0.03	-	-	-	-	-	0.072	742.4	Conforming Steel
F	0.10	1.02	2.03	0.014	0.001	0.04	0.003	0.07	0.03	-	-	-	-	-	0.082	742.3	Conforming Steel
G	0.10	1.03	2.03	0.014	0.001	0.04	0.003	0.06	0.08	-	-	-	-	-	0.078	754.2	Conforming Steel
H	0.15	1.02	2.01	0.021	0.002	0.04	0.003	0.10	-	-	-	-	-	-	0.128	734.1	Comparative Steel
I	0.12	1.01	2.00	0.015	0.002	0.03	0.003	0.04	-	-	-	-	-	-	0.113	736.9	Comparative Steel
J	0.09	1.05	3.50	0.016	0.003	0.04	0.003	0.07	0.02	-	-	-	-	-	0.074	723.8	Comparative Steel
K	0.12	0.30	2.00	0.017	0.003	0.04	0.003	0.12	-	-	-	-	-	-	0.094	722.2	Comparative Steel
L	0.09	1.41	1.97	0.015	0.003	0.04	0.003	0.08	-	0.10	-	0.20	-	0.0010	0.074	735.4	Conforming Steel
M	0.11	0.81	2.21	0.018	0.003	0.03	0.003	0.12	-	-	0.20	-	-	-	0.084	735.5	Conforming Steel
N	0.11	1.18	2.21	0.014	0.002	0.04	0.003	0.12	-	-	-	-	0.15	-	0.083	738.9	Conforming Steel

C*: Amount of C not fixed as carbide

without addition of Nb: C* = [%C] - (12/47.9) x [%Ti*]

with addition of Nb: [%C] - (12/92.9) x [%Nb] - (12/47.9) x [%Ti*]

Where Ti* = [%Ti] - (47.9/14) x [%N] - (47.9/32.1) x [%S]

[Table 5]

		Hot Rolling Condition		Cold Rolling Condition		Annealing Condition						Material Microstructure					Material Property					
Steel ID	Steel Sheet	Finisher Delivery Temp. (°C)	Coiling Temp. (°C)	Rolling Reduction Ratio (%)	Sheet Thickness (mm)	Heating Rate from (Ac_1 - 100 °C) to Ac_1 (°C/s)	Soaking Temp. (°C)	Soaking Duration (sec.)	Quench Stop Temp. (°C)	Cooling Rate to Quench Stop Temp. (°C/s)	Ferrite Fraction (%)	Martensite Fraction (%)	Balance (%)	Ferrite Grain Size (μm)	Martensite Grain Size (μm)	YS (MPa)	TS (MPa)	El (%)	Ec (GPa)	TS x El (MPa·%)	Remarks
B	B	880	600	68	1.41	20	830	100	300	15	67	33	0	3.3	1.0	623	938	17.9	249	16790	Inventive Example
C	C	880	600	68	1.41	30	830	60	300	15	77	23	0	3.0	0.8	549	814	20.3	250	16522	Inventive Example
D	D	880	600	68	1.41	30	830	60	300	15	74	26	0	3.2	0.9	593	895	18.9	248	16916	Inventive Example
E	E	880	600	68	1.41	20	830	60	300	15	68	32	0	2.8	1.0	680	1012	17.0	248	17172	Inventive Example
F	F	880	600	68	1.41	20	830	60	300	15	69	31	0	2.9	0.9	668	982	17.7	247	17341	Inventive Example
G	G	880	600	68	1.41	20	830	60	300	15	72	28	0	2.6	0.8	637	942	18.4	249	17333	Inventive Example
H	H	880	600	68	1.41	20	830	60	300	15	61	39	0	3.3	1.3	834	1132	13.2	235	14942	Comparative Example
I	I	880	600	68	1.41	20	830	60	300	15	63	37	0	3.7	1.4	774	1043	13.6	236	14185	Comparative Example
J	J	880	600	68	1.41	20	800	60	300	15	59	41	0	3.5	1.6	810	1203	13.2	227	15880	Comparative Example
K	K	880	600	68	1.41	20	800	60	300	15	73	27	0	3.4	1.1	549	776	18.2	246	14123	Comparative Example
L	L	880	600	75	1.10	20	830	60	300	15	68	32	0	3.3	1.0	680	1012	16.4	252	16597	Inventive Example
M	M	880	600	75	1.10	20	830	60	300	15	69	31	0	3.2	0.9	691	1020	16.3	251	16626	Inventive Example
N	N	880	600	75	1.10	20	830	60	300	15	73	27	0	3.1	0.9	673	993	17.0	251	16881	Inventive Example

[0075] As shown in Table 5, each of the steel sheets (Steel Sheets B to G and L to N) obtained according to the present invention exhibited excellent properties, such as TS of 780 MPa or higher, TS x El of 16500 or higher and Ec of 240 GPa or higher.

[0076] In contrast, Comparative Examples (Steel Sheets H to K) having chemical compositions out of an appropriate range according to the present invention are inferior in at least one of tensile strength (TS), strength-elongation balance (TS x El) and Young's modulus (Ec).

INDUSTRIAL APPLICABILITY

[0077] According to the present invention, it is possible to provide a thin steel sheet having both high strength and high rigidity with a tensile strength of 780 MPa or higher and Young's modulus of 240 GPa or higher.

Claims

1. A thin steel sheet having a composition including, in mass%, C: 0.06 % to 0.12 %, Si: 0.5 % to 1.5 %, Mn: 1.0 % to 3.0 %, P: 0.05 % or less, S: 0.01 % or less, Al: 0.5 % or less, N: 0.01 % or less, Ti: 0.02 % to 0.20 %, and the balance being Fe and incidental impurities, where the composition satisfies relations of Formula (1) and (2) below,
wherein the steel sheet has a microstructure such that a ferrite phase has an area ratio of 60 % or more and a martensite phase has an area ratio of 15 % to 35 %, where a total of the ferrite phase and the martensite phase is 95 % or more, an average grain size of ferrite is 4.0 μm or less and an average grain size of martensite is 1.5 μm or less, wherein the steel sheet has a tensile strength (TS) of 780 MPa or higher in a transverse direction perpendicular to the rolling direction, Young's modulus of 240 GPa or higher in the transverse direction, and strength-elongation balance (TS x El) of 16500 or more in the transverse direction, the strength-elongation balance being expressed by a product of the tensile strength (TS) and total elongation (El),

$$0.05 \leq [\%C] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (1)$$

, where

$$Ti^* = [\%Ti] - (47.9/14) \times [\%N] - (47.9/32.1) \times [\%S] \text{ ----- } (2),$$

and
[%M] indicates the content (mass %) of M element.

2. The thin steel sheet according to claim 1, wherein the composition of the steel sheet further includes, in mass%, Nb: 0.02 % to 0.10 %, and satisfies a relation of Formula (3) below in place of the Formula (1):

$$0.05 \leq [\%C] - (12/92.9) \times [\%Nb] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (3).$$

3. The thin steel sheet according to claim 1 or 2, wherein the composition of the steel sheet further includes, in mass%, one or more elements selected from Cr: 0.1 % to 1.0 %, Ni: 0.1 % to 1.0 %, Mo: 0.1 % to 1.0 %, Cu: 0.1 % to 2.0 % and B: 0.0005 % to 0.0030 %.

4. A method for manufacturing a thin steel sheet, the method comprising:

   in a hot rolling process; subjecting a steel material to finish rolling and completing the finish rolling at 850 °C to 950 °C to obtain a hot-rolled steel sheet, the steel material having a composition including, in mass%, C: 0.06 % to 0.12 %, Si: 0.5 % to 1.5 %, Mn: 1.0 % to 3.0 %, P: 0.05 % or less, S: 0.01 % or less, Al: 0.5 % or less, N: 0.01 % or less, Ti: 0.02 % to 0.20 %, and the balance being Fe and incidental impurities, where the contents of C, N, S and Ti satisfy relations of Formula (1) and (2) below;
   then coiling the steel sheet at 650 °C or lower;
   subjecting the steel sheet to pickling; and
   then subjecting the steel sheet to cold rolling at a rolling reduction ratio of 60 % or more;
   in a subsequent annealing process, heating the steel sheet to a soaking temperature of 780 to 880 °C at an average heating rate from (Ac_1 - 100 °C) to Ac_1 of 15 °C/s or higher;
   holding the steel sheet at the soaking temperature for 150 seconds or less; and
   cooling the steel sheet to 350 °C or lower at an average cooling rate until at least 350 °C of 5 °C/s to 50 °C/s,

$$0.05 \leq [\%C] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (1)$$

, where

$$Ti^* = [\%Ti] - (47.9/14) \times [\%N] - (47.9/32.1) \times [\%S] \text{ ----- } (2),$$

and
[%M] indicates the content (mass %) of M element.

5. The method for manufacturing a thin steel sheet according to claim 4, wherein the composition of the steel material further includes, in mass%, Nb: 0.02 % to 0.10 %, and satisfies a relation of Formula (3) below in place of the Formula (1):

$$0.05 \leq [\%C] - (12/92.9) \times [\%Nb] - (12/47.9) \times [\%Ti^*] \leq 0.10 \text{ ----- } (3).$$

6. The method for manufacturing a thin steel sheet according to claim 4 or 5, wherein the composition of the steel material further includes, in mass%, one or more elements selected from Cr: 0.1 % to 1.0 %, Ni: 0.1 % to 1.0 %, Mo: 0.1 % to 1.0 %, Cu: 0.1 % to 2.0 % and B: 0.0005 % to 0.0030 %.

INTERNATIONAL SEARCH REPORT

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A. CLASSIFICATION OF SUBJECT MATTER
C22C38/00(2006.01)i, *B21B3/00*(2006.01)i, *C21D9/46*(2006.01)i, *C22C38/06*(2006.01)i, *C22C38/58*(2006.01)i

According to International Patent Classification (IPC) or to both national classification and IPC

B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)
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Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched

Jitsuyo Shinan Koho	1922-1996	Jitsuyo Shinan Toroku Koho	1996-2013
Kokai Jitsuyo Shinan Koho	1971-2013	Toroku Jitsuyo Shinan Koho	1994-2013

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 2006-183130 A (JFE Steel Corp.), 13 July 2006 (13.07.2006), & US 2007/0144633 A1 & EP 1731627 A1 & WO 2005/095664 A1 & CA 2546009 A & KR 10-0881047 B1 & CN 1914345 A & AU 2005227564 A & TWB 00I312810	1-6
A	JP 2008-240123 A (JFE Steel Corp.), 09 October 2008 (09.10.2008), (Family: none)	1-6
A	JP 2008-240125 A (JFE Steel Corp.), 09 October 2008 (09.10.2008), (Family: none)	1-6

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

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Date of the actual completion of the international search	Date of mailing of the international search report
23 January, 2013 (23.01.13)	05 February, 2013 (05.02.13)

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INTERNATIONAL SEARCH REPORT

International application No.
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C (Continuation). DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	JP 2011-214069 A (Sumitomo Metal Industries, Ltd.), 27 October 2011 (27.10.2011), (Family: none)	1-6
A	JP 2010-275628 A (JFE Steel Corp.), 09 December 2010 (09.12.2010), & EP 2426230 A1 & WO 2010/126161 A1 & TW 201102444 A & CA 2759913 A & KR 10-2012-0008038 A & CN 102414335 A	1-6

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REFERENCES CITED IN THE DESCRIPTION

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Patent documents cited in the description

- JP 5255804 A **[0005] [0008]**
- JP 8311541 A **[0006] [0008]**
- JP 2006183131 A **[0007] [0008]**
- JP 2005314792 A **[0007] [0008]**