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

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(57) **ABSTRACT**

There is provided a high-stiffness high-strength thin steel sheet having a tensile strength of not less than 590 MPa and a Young's modulus of not less than 225 GPa, which comprises C: 0.02-0.15%, Si: not more than 1.5%, Mn: 1.5-4.0%, P: not more than 0.05%, S: not more than 0.01%, Al: not more than 1.5%, N: not more than 0.01% and Nb: 0.02-0.40% as mass %, provided that C, N and Nb contents satisfy $0.01 \leqq C+(12/14)\times N-(12/92.9)\times Nb \leqq 0.06$ and $N \leqq (14/92.9)\times(Nb-0.01)$ and the remainder being substantially iron and inevitable impurities, and has a texture comprising a ferrite phase as a main phase and having a martensite phase at an area ratio of not less than 1%.

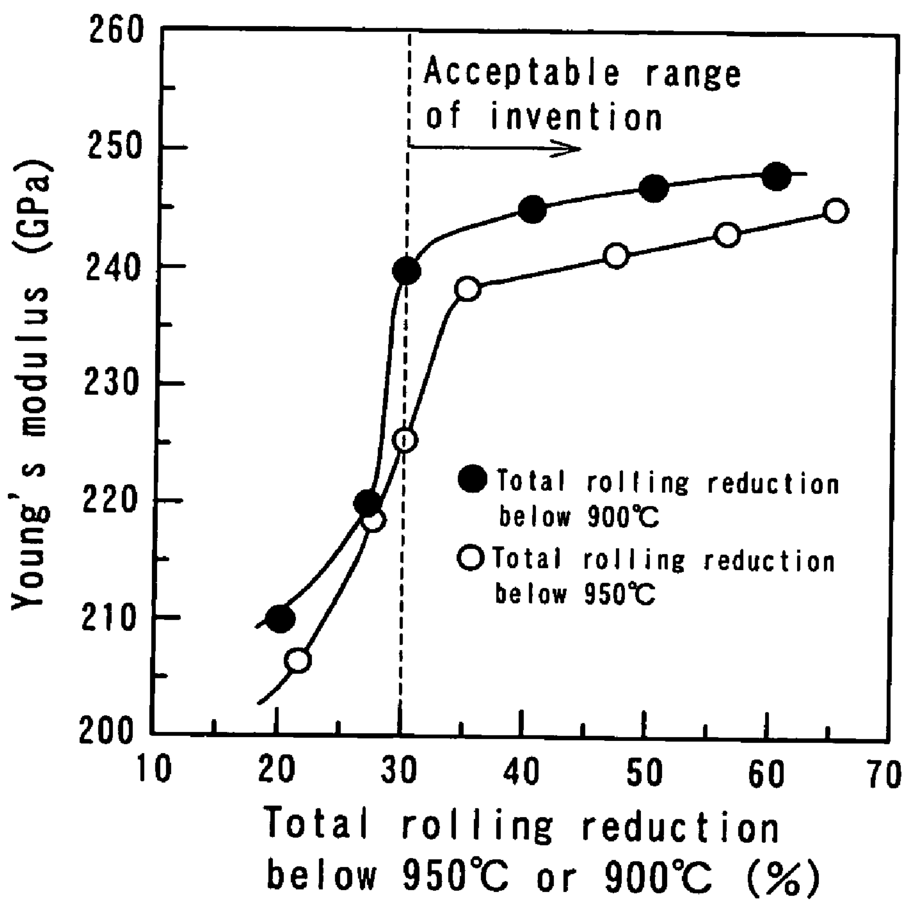


FIG. 1

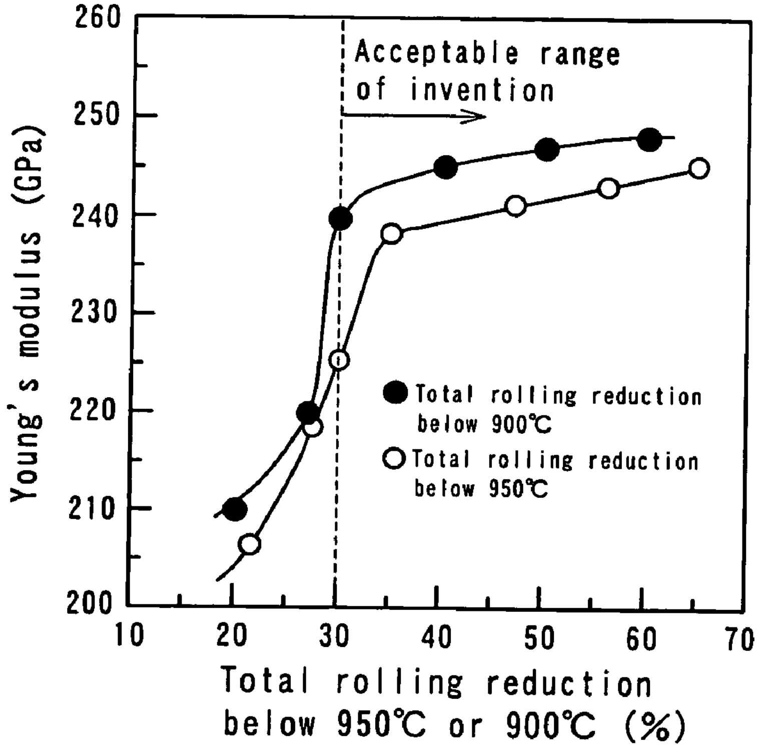

260
250
240
230
220
210
200
Young's modulus (GPa)
10
20
30
40
50
60
70
Total rolling reduction below 950℃ or 900℃ (%)
Acceptable range of invention
● Total rolling reduction below 900℃
○ Total rolling reduction below 950℃

FIG. 2

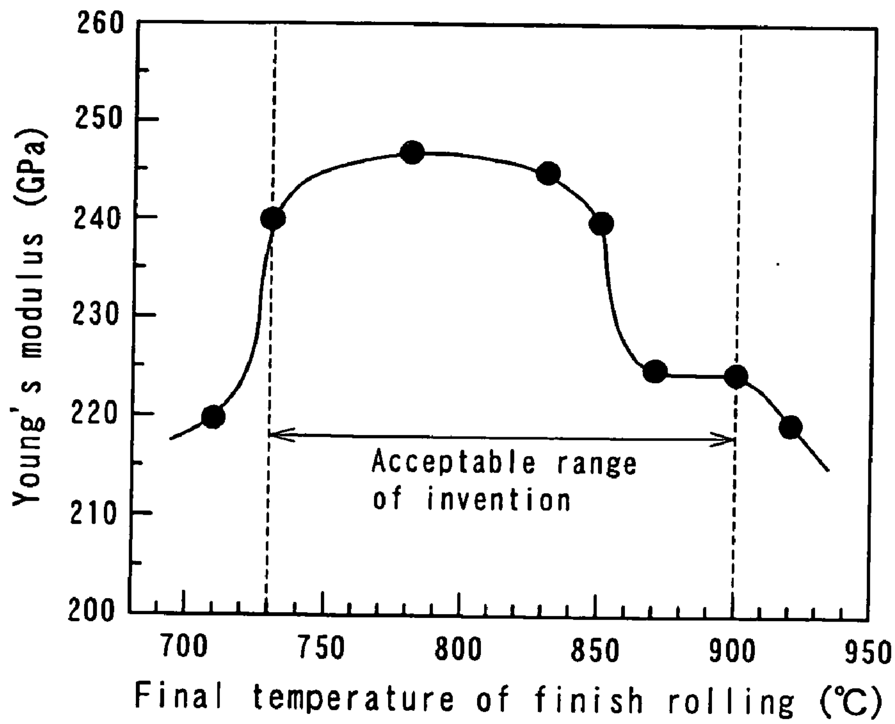

260
250
240
230
220
210
200
Young's modulus (GPa)
700
750
800
850
900
950
Final temperature of finish rolling (℃)
Acceptable range of invention

FIG. 3

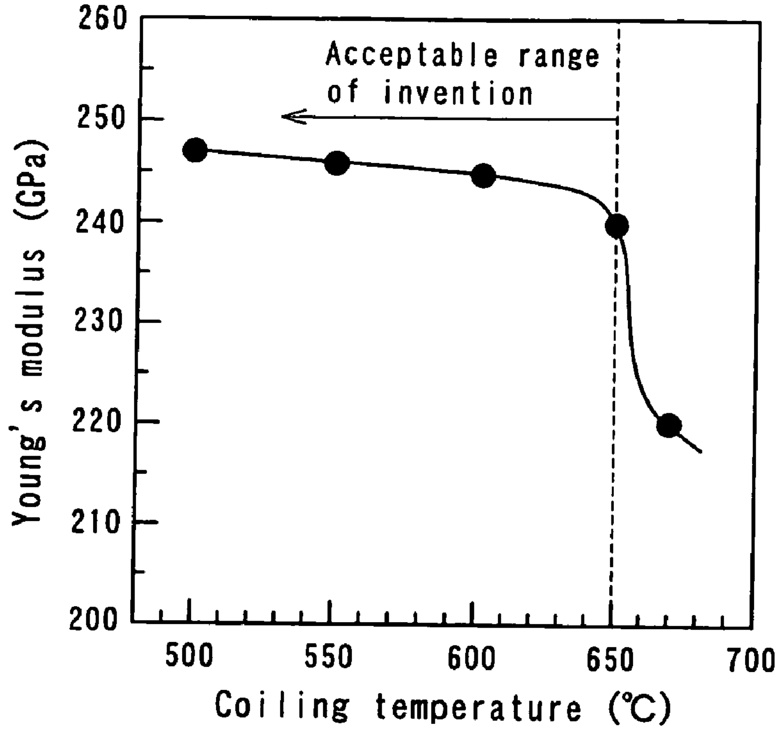

Acceptable range
of invention
Young's modulus (GPa)
260
250
240
230
220
210
200
500
550
600
650
700
Coiling temperature (℃)

FIG. 4

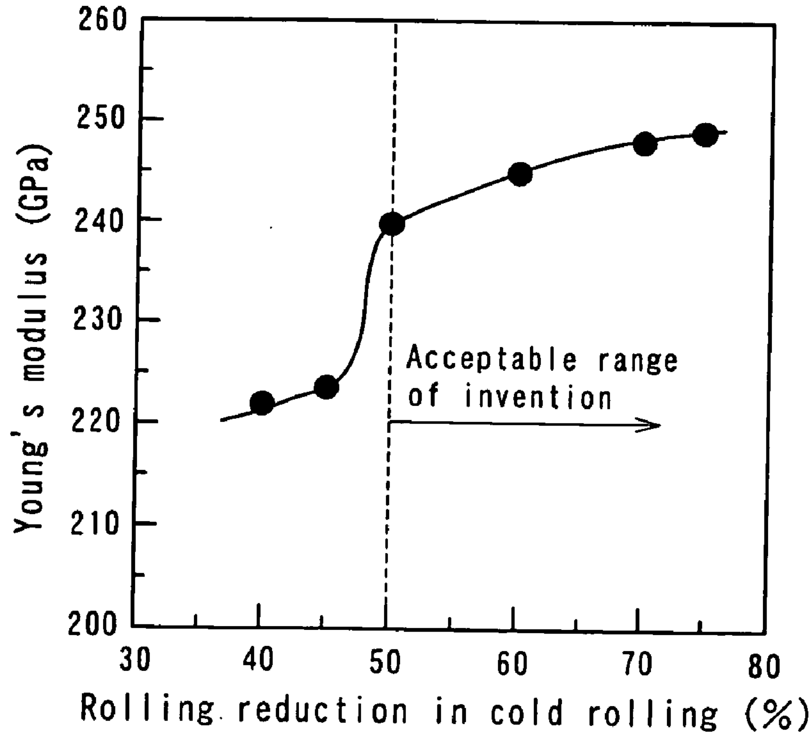
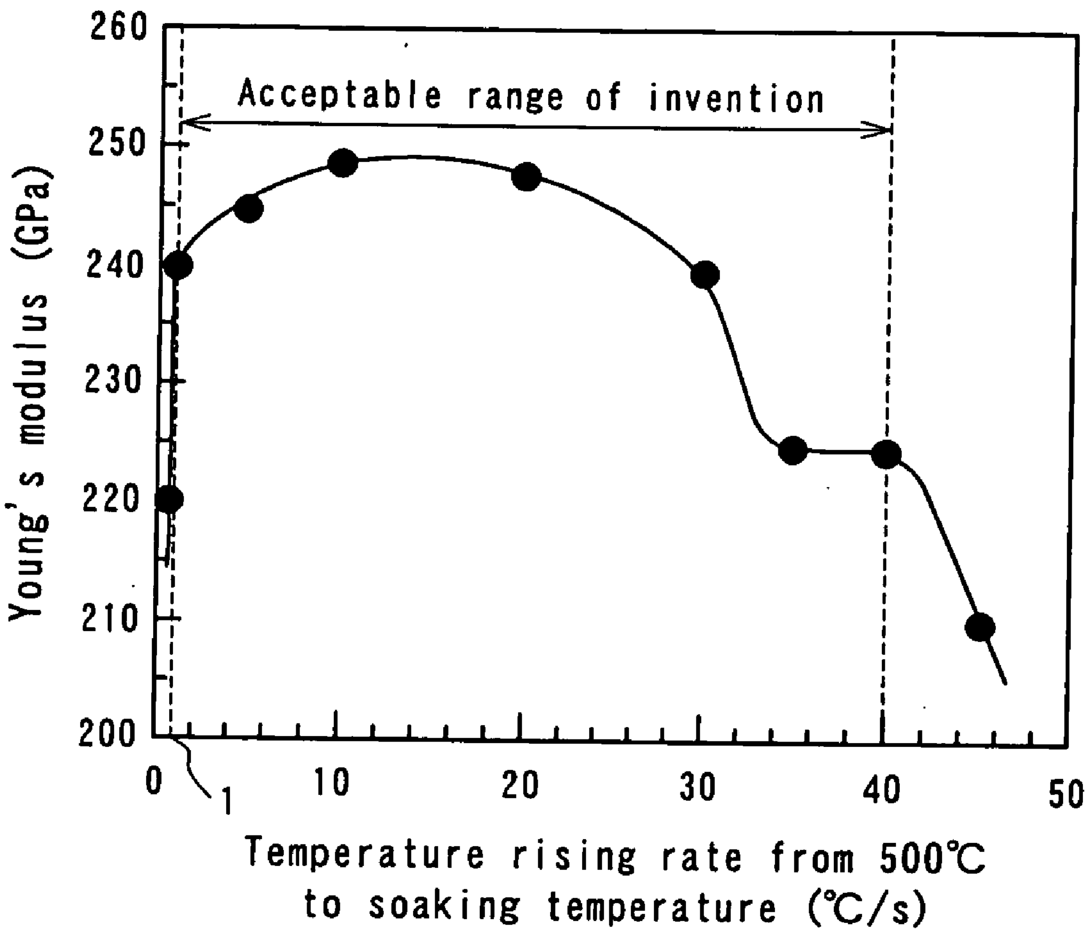

Acceptable range
of invention
Young's modulus (GPa)
260
250
240
230
220
210
200
30
40
50
60
70
80
Rolling reduction in cold rolling (%)

FIG. 5


260
250
240
230
220
210
200
Young's modulus (GPa)
Acceptable range of invention
0
1
10
20
30
40
50
Temperature rising rate from 500℃
to soaking temperature (℃/s)

HIGH-STIFFNESS HIGH-STRENGTH THIN STEEL SHEET AND METHOD FOR PRODUCING THE SAME

TECHNICAL FIELD

[0001] This invention relates to a high-stiffness high-strength thin steel sheet suitable mainly as a vehicle body for automobiles and a method for producing the same. Moreover, the high-stiffness high-strength thin steel sheet according to the invention is a column-shaped structural member having a thickness susceptibility index of the stiffness near to 1 such as a center pillar, locker, side flame, cross member or the like of the automobile and is widely suitable for applications requiring a stiffness.

RELATED ART

[0002] As a result of recent heightened interest in global environment problems, the exhaust emission control is conducted even in the automobiles, and hence the weight reduction of the vehicle body in the automobile is a very important matter. For this end, it is effective to attain the weight reduction of the vehicle body by increasing the strength of the steel sheet to reduce the thickness thereof.

[0003] Recently, the increase of the strength in the steel sheet is considerably advanced, and hence the use of thin steel sheets having a thickness of less than 2.0 mm is increasing. In order to further reduce the weight by the increase of the strength, it is indispensable to simultaneously control the deterioration of the stiffness in parts through the thinning of the thickness. Such a problem of deteriorating the stiffness of the parts through the thinning of the thickness in the steel sheet is actualized in steel sheets having a tensile strength of not less than 590 MPa, and particularly this problem is serious in steel sheets having a tensile strength of not less than 700 MPa.

[0004] In general, in order to increase the stiffness of the parts, it is effective to change the shape of the parts, or to increase the number of welding points or change the welding condition such as changeover to laser welding or the like in the spot-welded parts. However, when these parts are used in the automobile, there are problems that it is not easy to change the shape of the parts in a limited space inside the automobile, and the change of the welding conditions causes the increase of the cost and the like.

[0005] Consequently, in order to increase the stiffness of the parts without changing the shape of the parts or the welding conditions, it becomes effective to increase the Young's modulus of the material used in the parts.

[0006] In general, the stiffness of the parts under the same shape of parts and welding conditions is represented by a product of Young's modulus of the material and geometrical moment of inertia of the part. Further, the geometrical moment of inertia can be expressed so as to be approximately proportionate to t^{λ} when the thickness of the material is t. In this case, λ is a thickness susceptibility index and is a value of 1-3 in accordance with the shape of the parts. For example, in case of one plate shape such as panel parts for the automobile, λ is a value near to 3, while in case of column-shape such as structural parts, λ is a value near to 1.

[0007] When λ of the parts is 3, if the thickness is made small by 10% while equivalently maintaining the stiffness of the parts, it is required to increase the Young's modulus of the material by 37%, while when λ of the parts is 1, if the thickness is made small by 10%, it may be enough to increase the Young's modulus by 11%.

[0008] That is, in case of the parts having λ near to 1 such as column-shaped parts, it is very effective to increase the Young's modulus of the steel sheet itself for the weight reduction. Particularly, in case of steel sheets having a high strength and a small thickness, it is strongly demanded to highly increase the Young's modulus of the steel sheet.

[0009] In general, the Young's modulus is largely dependent upon the texture and is known to become high in a closest direction of atom. Therefore, it is effective to develop {112}<110> in order to develop an orientation advantageous for the Young's modulus of steel being a body-centered cubic lattice in a steel making process comprising the rolling through rolls and the heat treatment, whereby the Young's modulus can be increased in a direction perpendicular to the rolling direction.

[0010] There have hitherto been variously examined steel sheets by controlling the texture to increase the Young's modulus.

[0011] For example, the patent article 1 discloses a technique wherein a steel obtained by adding Nb or Ti to an extremely low carbon steel is hot-rolled at a rolling reduction at Ar_3-Ar_3+150° C.) of not less than 85% to promote transformation from non-crystallized austenite to ferrite to thereby render the texture of ferrite at the stage of the hot-rolled sheet into {311}<011> and {332}<113>, which is an initial orientation and is subjected to a cold rolling and a recrystallization annealing to render {211}<011> into a main orientation to thereby increase the Young's modulus in a direction perpendicular to the rolling direction.

[0012] Also, the patent article 2 discloses a method for producing a hot rolled steel sheet having an increased Young's modulus in which Nb, Mo and B are added to a low carbon steel having a C content of 0.02-0.15% and the rolling reduction at Ar_3-950° C. is made to not less than 50% to develop [211]<011>.

[0013] Further, the patent article 3 discloses a method for producing. a hot rolled steel sheet having a high stiffness in which Nb is added to a low carbon steel having a C content of not more than 0.05% and a finish rolling start temperature is made to not higher than 950° C. and a finish rolling end temperature is made to (Ar_3−50° C.)-(Ar_3+100° C.) to control the development of {100} decreasing the Young's modulus.

[0014] Moreover, the patent article 4 discloses a method for producing a hot rolled steel sheet in which Si and Al are added to a low carbon steel having a C content of not more than 0.05% to enhance Ar_3 transformation point and the rolling reduction below Ar_3 transformation point in the hot rolling is made to not less than 60% to increase Young's modulus in a direction perpendicular to the rolling direction.

[0015] Patent article 1: JP-A-H05-255804

[0016] Patent article 2: JP-A-H08-311541

[0017] Patent Article 3: JP-A-H05-247530

[0018] Patent article 4: JP-A-H09-53118

DISCLOSURE OF THE INVENTION

Problems to be Solved in the Invention

[0019] However, the aforementioned techniques have the following problems.

[0020] In the technique disclosed in the patent article 1, the Young's modulus of the steel sheet is increased by using the extremely low carbon steel having a C content of not more than 0.01% to control the texture, but the tensile strength is low as about 450 MPa at most, so that there is a problem in the increase of the strength by applying this technique.

[0021] In the technique disclosed in the patent article 2, since the C content is as high as 0.02-0.15%, it is possible to increase the strength, but as the target steel sheet is the hot rolled steel sheet, the control of the texture through cold working can not be utilized, and hence there are problems that it is difficult to further increase the Young's modulus but also it is difficult to stably produce high-strength steel sheets having a thickness of less than 2.0 mm through low-temperature finish rolling.

[0022] Also, the technique disclosed in the patent article 3 is the production of the hot rolled steel sheet, so that it has the same problems as mentioned above.

[0023] Further, in the technique disclosed in the patent article 4, the crystal grains are coarsened by conducting the rolling at the ferrite zone, so that there is a problem that the workability is considerably deteriorated.

[0024] Thus, the increase of the Young's modulus in the steel sheet by the conventional techniques is targeted to hot rolled steel sheets having a thick thickness or soft steel sheets, so that it is difficult to increase the Young's modulus of high-strength thin steel sheet having a thickness of not more than 2.0 mm by using the above conventional techniques.

[0025] As a strengthening mechanism for increasing the tensile strength of the steel sheet to not less than 590 MPa, there are mainly a precipitation strengthening mechanism and a transformation texture strengthening mechanism.

[0026] When the precipitation strengthening mechanism is used as the strengthening mechanism, it is possible to increase the strength while suppressing the lowering of the Young's modulus of the steel sheet as far as possible, but the following difficulty is accompanied. That is, when utilizing the precipitation strengthening mechanism for finely precipitating, for example, a carbonitride of Ti, Nb or the like, in the hot rolled steel sheet, the increase of the strength is attained by conducting the fine precipitation in the coiling after the hot rolling, but in the cold rolled steel sheet, the coarsening of the precipitate can not be avoided at the step of recrystallization annealing after the cold rolling and it is difficult to increase the strength through the precipitation strengthening.

[0027] When utilizing the transformation texture strengthening mechanism as the strengthening mechanism, there is a problem that the Young's modulus of the steel sheet lowers due to strain included in a low-temperature transformation phase such as bainite phase, martensite phase or the like.

[0028] It is, therefore, an object of the invention to solve the above problems and to provide a high-stiffness high-strength thin steel sheet having a tensile strength of not less than 590 MPa, preferably not less than 700 MPa, a Young's modulus of not less than 225 GPa, preferably not less than 230 GPa, more preferably not less than 240 GPa and a thickness of not more than 2.0 mm as well as an advantageous method for producing the same.

Means for Solving Problems

[0029] In order to achieve the above object, the gist and construction of the invention are as follows.

[0030] (I) A high-stiffness high-strength thin steel sheet comprising C: 0.02-0.15%, Si: not more than 1.5%, Mn: 1.5-4.0%, P: not more than 0.05%, S: not more than 0.01%, Al: not more than 1.5%, N: not more than 0.01% and Nb: 0.02-0.40% as mass %, provided that C, N and Nb contents satisfy the relationships of the following equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N-(12/92.9)\times Nb \leqq 0.06 \tag{1}$$

$$N \leqq (14/92.9)\times(Nb-0.01) \tag{2}$$

and the remainder being substantially iron and inevitable impurities, and having a texture comprising a ferrite phase as a main phase and having a martensite phase at an area ratio of not less than 1%, and having a tensile strength of not less than 590 MPa and a Young's modulus of not less than 225 GPa.

[0031] (II) A high-stiffness high-strength thin steel sheet according to the item (I), which further contains one or two of Ti: 0.01-0.50% and V: 0.01-0.50% as mass % in addition to the above composition and satisfy the relationships of the following equations (3) and (4) instead of the equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N^*-(12/92.9)\times Nb-(12/47.9)\times Ti^*-(12/50.9)\times V \leqq 0.06 \tag{3}$$

$$N^* \leqq (14/92.9)\times(Nb-0.01) \tag{4}$$

provided that N^* in the equations (3) and (4) is $N^*=N-(14/47.9)\times Ti$ at $N-(14/47.9)\times Ti>0$ and $N^*=0$ at $N-(14/47.9)\times Ti \leqq 0$, and Ti^* in the equation (3) is $Ti^*=Ti-(47.9/14)\times N-(47.9/32.1)\times S$ at $Ti-(47.9/14)\times N-(47.9/32.1)\times S>0$ and $Ti^*=0$ at $Ti-(47.9/14)\times N-(47.9/32.1)\times S \leqq 0$.

[0032] (III) A high-stiffness high-strength thin steel sheet according to the item (I) or (II), which further contains one or more of Cr: 0.1-1.0%, Ni: 0.1-1.0%, Mo: 0.1-1.0%, Cu: 0.1-2.0% and B: 0.0005-0.0030% as mass % in addition to the above composition.

[0033] (IV) A method for producing a high-stiffness high-strength thin steel sheet comprising subjecting a starting material of steel comprising C: 0.02-0.15%, Si: not more than 1.5%, Mn: 1.5-4.0%, P: not more than 0.05%, S: not more than 0.01%, Al: not more than 1.5%, N: not more than 0.01% and Nb: 0.02-0.40% as mass %, provided that C, N and Nb contents satisfy the relationships of the following equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N-(12/92.9)\times Nb \leqq 0.06 \tag{1}$$

$$N \leqq (14/92.9)\times(Nb-0.01) \tag{2}$$

to a hot rolling step under conditions that a total rolling reduction below 950° C. is not less than 30% and a finish rolling is terminated at Ar_3-900° C., coiling the hot rolled sheet below 650° C., pickling, subjecting to a cold rolling at a rolling reduction of not less than 50%, raising a temperature to 780-900° C. at a temperature rising rate from 500° C. of 1-40° C./s to conduct soaking, and then cooling at a cooling rate up to 500° C. of not less than 5° C./s to conduct annealing.

[0034] (V) A method for producing a high-stiffness high-strength thin steel sheet according to the item (IV), wherein the starting material of steel further contains one or two of Ti: 0.01-0.50% and V: 0.01-0.50% as mass % in addition to the above composition and satisfies the relationships of the following equations (3) and (4) instead of the equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N^{*}-(12/92.9)\times Nb-(12/47.9)\times Ti^{*}-(12/50.9)\times V \leqq 0.06 \tag{3}$$

$$N^{*} \leqq (14/92.9)\times(Nb-0.01) \tag{4}$$

provided that N* in the equations (3) and (4) is $N^{*}=N-(14/47.9)\times Ti$ at $N-(14/47.9)\times Ti>0$ and $N^{*}=0$ at $N-(14/47.9)\times Ti \leqq 0$, and Ti* in the equation (3) is $Ti^{*}=Ti-(47.9/14)\times N-(47.9/32.1)\times S$ at $Ti-(47.9/14)\times N-(47.9/32.1)\times S>0$ and $Ti^{*}=0$ at $Ti-(47.9/14)\times N-(47.9/32.1)\times S \leqq 0$.

[0035] (VI) A method for producing a high-stiffness high-strength thin steel sheet according to the item (IV) or (V), wherein the staring material of steel further contains one or more of Cr: 0.1-1.0%, Ni: 0.1-1.0%, Mo: 0.1-1.0%, Cu: 0.1-2.0% and B: 0.0005-0.0030% as mass % in addition to the above composition.

Effect of the Invention

[0036] According to the invention, it is possible to provide a high-stiffness high-strength thin steel sheet having a tensile strength of not less than 590 MPa, preferably not less than 700 MPa and a Young's modulus of not less than 225 GPa, preferably not less than 230 GPa, more preferably not less than 240 GPa.

[0037] That is, the starting material of low carbon steel added with Mn and Nb is roll-reduced below 950° C., preferably below 900° C. (strictly speaking, just above Ar_3 point) in the hot rolling to promote the transformation from non-recrystallized austenite to ferrite and then cold rolled to develop a crystal orientation useful for the improvement of Young's modulus and thereafter a low-temperature transformation phase suppressing the lowering of the Young's modulus is produced and a greater amount of ferrite phase useful for the improvement of the Young's modulus is retained in the cooling stage by the control of the heating rate in the annealing step and the soaking at two-phase region, whereby the thin steel sheet satisfying higher strength and higher Young's modulus can be produced, which develops an effective effect in industry.

[0038] Further explaining in detail, the starting material of low carbon steel added with Mn and Nb is roll-reduced just above Ar_3 transformation point in the hot rolling to increase the non-recrystallized austenite texture having a crystal orientation of {112}<111>, and subsequently the transformation from the non-recrystallized austenite of {112}<111> to ferrite is promoted in the cooling stage to develop ferrite orientation of {113}<110>.

[0039] In the cold rolling after the coiling and pickling, the rolling is carried out at a rolling reduction of not less than 50% to turn the crystal orientation of {113}<110> to {112}<110> useful for the improvement of the Young's modulus, and in the temperature rising stage at the subsequent annealing step, the temperature is raised from 500° C. to the soaking temperature at a heating rate of 1-40° C./s to promote the recrystallization of ferrite having an orientation of {112}<110> and provide a two-phase region at a state of partly retaining the non-recrystallized grains of {112}<110>, whereby the transformation from the non-recrystallized ferrite of {112}<110> to austenite can be promoted.

[0040] Further, in the transformation from austenite phase to ferrite phase at the cooling after the soaking, ferrite grains having an orientation of {112}<110> is grown to enhance the Young's modulus, while the steel enhancing the hardenability by the addition of Mn is cooled at a rate of not less than 5° C./s to produce the low-temperature transformation phase, whereby it is attempted to increase the strength.

[0041] Moreover, the low-temperature transformation phase is produced by retransforming the austenite phase transformed from ferrite having an orientation of {112}<110> during the cooling, so that {112}<110> can be also developed even in the crystal orientation of the low-temperature transformation phase.

[0042] Thus, the Young's modulus is enhanced by developing {112}<110> of ferrite phase, and particularly {112}<110> is increased in the orientation of the low-temperature transformation phase largely exerting on the lowering of the Young's modulus, whereby the strength can be increased by the formation of the low-temperature transformation phase and the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase can be largely suppressed.

BRIEF DESCRIPTION OF THE DRAWINGS

[0043] FIG. **1** is a graph showing an influence of a total rolling reduction below 950° C. or below 900° C. on Young's modulus;

[0044] FIG. **2** is a graph showing an influence of a final temperature in hot finish rolling on Young's modulus;

[0045] FIG. **3** is a graph showing an influence of a coiling temperature on Young's modulus;

[0046] FIG. **4** is a graph showing an influence of a rolling reduction in cold rolling on Young's modulus; and

[0047] FIG. **5** is a graph showing an influence of an average temperature rising rate from 500° C. to soaking temperature in annealing on Young's modulus.

BEST MODE FOR CARRYING OUT THE INVENTION

[0048] The high-stiffness high-strength thin steel sheet according to the invention is a steel sheet having a tensile strength of not less than 590 MPa, preferably not less than 700 MPa, a Young's modulus of not less than 225 GPa, preferably not less than 230 GPa, more preferably not less than 240 GPa, and a thickness of not more than 2.0 mm. Moreover, the steel sheet to be targeted in the invention includes steel sheets subjected to a surface treatment such as galvanization inclusive of alloying, zinc electroplating or the like in addition to the cold rolled steel sheet.

[0049] The reason of limiting the chemical composition in the steel sheet of the invention will-be described below. Moreover, the unit for the content of each element in the chemical composition of the steel sheet is "% by mass", but it is simply shown by "%" unless otherwise specified.

[0050] C: 0.02-0.15%

[0051] C is an element stabilizing austenite and can largely contribute to increase the strength by enhancing the hardenability at the cooling stage in the annealing after the cold rolling to largely promote the formation of the low-temperature transformation phase. Further, the Ar_3 transformation point is lowered in the hot rolling and it is possible to conduct the rolling at a lower temperature region when the rolling is conducted just above Ar_3, whereby the transformation from the non-recrystallized austenite to ferrite can be promoted to develop {113}<110>, and the Young's modulus can be improved at the subsequent cold rolling and annealing steps. Moreover, C can contribute to increase the Young's modulus by promoting the transformation of ferrite grains having {112}<110> from the non-recrystallized ferrite to austenite after the cold rolling.

[0052] In order to obtain such effects, the C content is required to be not less than 0.02%, preferably not less than 0.05%, more preferably not less than 0.06%. On the other hand, when the C content exceeds 0.15%, the fraction of hard low-temperature transformation phase becomes large, and the strength of the steel is extremely increased but also the workability is deteriorated. Also, the greater amount of C suppresses the recrystallization of the orientation useful for the increase of the Young's modulus at the annealing step after the cold rolling. Further, the greater amount of C brings about the deterioration of the weldability.

[0053] Therefore, the C content is required to be not more than 0.15%, preferably not more than 0.10%.

[0054] Si: not more than 1.5%

[0055] Si raises the Ar_3 transformation point in the hot rolling, so that when the rolling is carried out just above Ar_3, the recrystallization of worked austenite is promoted. Therefore, when Si is contained in an amount exceeding 1.5%, the crystal orientation required for the increase of the Young's modulus can not be obtained. Also, the greater amount of Si deteriorates the weldability of the steel sheet but also promotes the formation of fayalite on a surface of a slab in the heating at the hot rolling step to accelerate the occurrence of surface pattern so-called as a red scale. Furthermore, in case of using as a cold rolled steel sheet, Si oxide produced on the surface deteriorates the chemical conversion processability, while in case of using as a galvanized steel sheet, Si oxide produced on the surface induces non-plating. Therefore, the Si content is required to be not more than 1.5%. Moreover, in case of steel sheets requiring the surface properties or the galvanized steel sheet, the Si content is preferable to be not more than 0.5%.

[0056] Also, Si is an element stabilizing ferrite and promotes the ferrite transformation at the cooling stage after the soaking of two-phase region in the annealing step after the cold rolling to enrich C in austenite, whereby austenite can be stabilized to promote the formation of the low-temperature transformation phase. For this end, the strength of steel can be increased, if necessary. In order to obtain such an effect, the Si content is desirable to be not less than 0.2%.

[0057] Mn: 1.5-4.0%

[0058] Mn is one of important elements in the invention. Mn is an element suppressing the recrystallization of worked austenite in the hot rolling and stabilizing austenite, and since Mn lowers the Ar_3 transformation point, when the rolling is carried out just above Ar_3, it is possible to conduct the rolling at a lower temperature region, and further Mn has an action of suppressing the recrystallization of the worked austenite. Moreover, Mn can promote the transformation from the non-recrystallized austenite to ferrite to develop {113}<110> and improve the Young's modulus in the subsequent cold rolling and annealing steps.

[0059] Furthermore, Mn as an austenite stabilizing element lowers Ac_1 transformation point in the temperature rising stage at the annealing step after the cold rolling to promote the transformation from the non-recrystallized ferrite to austenite, and can develop the orientation useful for the improvement of the Young's modulus to control the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase with respect to the orientation of the low-temperature transformation phase produced in the cooling stage after the soaking.

[0060] Also, Mn enhances the hardenability in the cooling stage after the soaking and annealing at the annealing step to largely promote the formation of the low-temperature transformation phase, which can largely contribute to the increase of the strength. Further, Mn acts as a solid-solution strengthening element, which can contribute to the increase of the strength in steel. In order to obtain such an effect, the Mn content is required to be not less than 1.5%.

[0061] On the other hand, when the Mn content exceeds 4.0%, Ac_3 transformation point is excessively lowered in the temperature rising stage at the annealing step after the cold rolling, so that the recrystallization of ferrite phase at the two-phase region is difficult and it is required to raise the temperature up to an austenite single-phase region above Ac_3 transformation point. As a result, ferrite of {112}<110> orientation useful for the increase of the Young's modulus obtained by the recrystallization of worked ferrite can not be developed to bring about the lowering of the Young's modulus. Further, the greater amount of Mn deteriorates the weldability of the steel sheet. Therefore, the Mn content is not more than 4.0%, preferably not more than 3.5%.

[0062] P: not more than 0.05%

[0063] Since P segregates in the grain boundary, if the P content exceeds 0.05%, the ductility and toughness of the steel sheet lower but also the weldability is deteriorated. In case of using the alloyed galvanized steel sheet, the alloying rate is delayed by P. Therefore, the P content is required to be not more than 0.05%. On the other hand, P is an element effective for the increase of the strength as a solid-solution strengthening element and has an action of promoting the enrichment of C in austenite as a ferrite stabilizing element. In the steel added with Si, it has also an action of suppressing the occurrence of red scale. In order to obtain these actions, the P content is preferable to be not less than 0.01%.

[0064] S: not more than 0.01%

[0065] S considerably lowers the hot ductility to induce hot tearing and considerably deteriorate the surface properties. Further, S hardly contributes to the strength but also forms coarse MnS as an impurity element to lower the ductility and drill-spreading property. These problems become remarkable when the S content exceeds 0.01%, so that it is desirable to reduce the S content as far as possible. Therefore, the S content is not more than 0.01%. From a viewpoint of improving the drill-spreading property, it is preferable to be not more than 0.005%.

[0066] Al: not more than 1.5%

[0067] It is an element useful for deoxidizing steel to improve the cleanness of the steel. However, Al is a ferrite stabilizing element, and largely raises the Ar_3 transformation of the steel, so that when the rolling is carried out just above Ar_3, the recrystallization of worked austenite is promoted to suppress the development of the crystal orientation required for the increase of the Young's modulus. Further, when the Al content exceeds 1.5%, the austenite single-phase region disappears and it is difficult to terminate the rolling at austenite region in the hot rolling step. Therefore, the Al content is required to be not more than 1.5%. From this viewpoint, Al is preferable to be made lower, and further preferable to be limited to not more than 0.1%. On the other hand, Al as a ferrite forming element promotes the formation of ferrite in the cooling stage after the soaking at the two-phase region in the annealing step after the cold rolling to enrich C in austenite, whereby austenite can be stabilized to promote the formation of the low-temperature transformation phase. As a result, the strength of the steel can be enhanced, if necessary. In order to obtain such an effect, the Al content is desirable to be not less than 0.2%.

[0068] N: not more than 0.01%

[0069] N is a harmful element because slab breakage is accompanied in the hot rolling to cause surface defect. When the N content exceeds 0.01%, the occurrence of slab breakage and surface defect becomes remarkable. Therefore, the N content is required to be not more than 0.01%.

[0070] Nb: 0.02-0.40%

[0071] Nb is a most important element in the invention. That is, Nb suppresses the recrystallization of worked austenite at the finish rolling step in the hot rolling to promote the transformation from the non-recrystallized austenite to ferrite and develop {113}<110> and can increase the Young's modulus at the subsequent cold rolling and annealing steps. Also, the recrystallization of worked ferrite is suppressed at the temperature rising stage in the annealing step after the cold rolling to promote the transformation from the non-recrystallized ferrite to austenite. As to the orientation of the low-temperature transformation phase produced in the cooling stage after the soaking, the orientation useful for the increase of the Young's modulus can be developed to suppress the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase. Also, a fine carbonitride of Nb can contribute to the increase of the strength. In order to obtain such an action, the Nb content is required to be not less than 0.02%, preferably not less than 0.05%.

[0072] On the other hand, when the Nb content exceeds 0.40%, the all carbonitride can not be solid-soluted in the re-heating at the usual hot rolling step and hence coarse carbonitride remains, so that the effect of suppressing the recrystallization of worked austenite in the hot rolling step and the effect of suppressing the recrystallization of worked ferrite in the annealing step after the cold rolling can not be obtained. Also, even if the hot rolling of the slab after the continuous casting is started as it is without conducting the re-heating after the continuously cast slab is cooled, when Nb is included in an amount exceeding 0.40%, the improvement of the effect of suppressing the recrystallization is not recognized and the increase of the alloy cost is caused. Therefore, the Nb content is 0.02-0.40%, preferably 0.05-0.40%.

[0073] In the invention, the contents of C, N and Nb are required to satisfy the relationship of the following equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N-(12/92.9)\times Nb \leqq 0.06 \tag{1}$$

$$N \leqq (14/92.9)\times(Nb-0.01) \tag{2}$$

[0074] If C not fixed as a carbonitride is existent in an amount exceeding 0.06%, the introduction of strain in the cold rolling becomes non-uniform and further the recrystallization of the orientation useful for the increase of the Young's modulus is suppressed, so that the C amount not fixed as the carbonitride calculated by (C+(12/14)×N−(12/92.9)×Nb) is required to be not more than 0.06%, preferably not more than 0.05%. At this moment, N is preferentially fixed and precipitated as compared with C, so that the C amount not fixed as the carbonitride can be calculated by (C+(12/14)×N−(12/92.9)×Nb). On the other hand, when the C amount not fixed as the carbonitride is less than 0.01%, the C content in austenite decreases in the annealing at the two-phase region after the cold rolling and the formation of martensite phase after the cooling is suppressed, so that it is difficult to increase the strength of the steel. Therefore, the amount of (C+(12/14)×N−(12/92.9)×Nb), which is the C amount not fixed as the carbonitride, is 0.01-0.06%, preferably 0.01-0.05%. Further, N coarsely precipitates a nitride of Nb at a high temperature, and hence the effect of suppressing the recrystallization by Nb is reduced. In order to control this action, the N content is required to be limited to $N \leqq (14/92.9)\times(Nb-0.01)$ in relation with the Nb content, preferably $N \leqq (14/92.9) \leqq (Nb-0.02)$.

[0075] Moreover, the term "the remainder being substantially iron and inevitable impurities" used herein means that steels containing slight amounts of other elements without damaging the action and effect of the invention are included within the scope of the invention. In case of further increasing the strength, one or two of Ti and V and one or more of Cr, Ni, Mo, Cu and B may be added, if necessary, in addition to the above definition of the chemical composition.

[0076] Ti: 0.01-0.50%

[0077] Ti is an element contributing to the increase of the strength by forming a fine carbonitride. Also, it is an element contributing to the increase of the Young's modulus by suppressing the recrystallization of worked austenite in the finish rolling step of the hot rolling to promote the transformation from the non-recrystallized austenite to ferrite. Since Ti has the above actions, the content is preferable to be not less than 0.01%. On the other hand, when the Ti content exceeds 0.50%, all the carbonitride can not be solid-soluted in the re-heating at the usual hot rolling step and a coarse carbonitride remains, and hence the effect of increasing the strength and the effect of suppressing the recrystallization can not be obtained. Also, even if the hot rolling of the slab after the continuous casting is started as it is without conducting the re-heating after the continuously cast slab is cooled, the Ti content exceeding 0.50% is small in the contribution to the effect of increasing the strength and the effect of suppressing the recrystallization and also the increase of the alloy cost is caused. Therefore, the Ti content is preferably not more than 0.50%, more preferably not more than 0.20%.

[0078] V: 0.01-0.50%

[0079] V is an element contributing to the increase of the strength by forming a fine carbonitride. Since V has such an action, the V content is preferable to be not less than 0.01%. On the other hand, when the V content exceeds 0.50%, the effect of increasing the strength by the amount exceeding 0.50% is small and the increase of the alloy cost is caused. Therefore, the V content is preferably not more than 0.50%, more preferably not more than 0.20%.

[0080] In the invention, when Ti and/or V are included in addition to Nb, the contents of C, N, S, Nb, Ti and V are required to satisfy the relationship of the following equations (3) and (4) instead of the equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N^*-(12/92.9)\times Nb-(12/47.9)\times Ti^*-(12/50.9)\times V \leqq 0.06 \tag{3}$$

$$N^* \leqq (14/92.9)\times(Nb-0.01) \tag{4}$$

provided that N* in the equations (3) and (4) is N*=N−(14/47.9)×Ti at N−(14/47.9)×Ti>0 and N*=0 at N−(14/47.9)×

$Ti \leqq 0$, and Ti* in the equation (3) is $Ti^* = Ti - (47.9/14) \times N - (47.9/32.1) \times S$ at $Ti - (47.9/14) \times N - (47.9/32.1) \times S > 0$ and $Ti^* = 0$ at $Ti - (47.9/14) \times N - (47.9/32.1) \times S \leqq 0$.

[0081] Further, N coarsely precipitates the nitride of Nb at a high temperature as previously mentioned, so that the effect of suppressing the recrystallization through Nb is decreased. In case of Ti-containing steel, N is preferentially fixed as a nitride of Ti, N* as a N amount not fixed as a nitride of Ti is required to be limited to N* s $(14/92.9) \times (Nb - 0.01)$, preferably $N^* \leqq (14/92.9) \times ((Nb - 0.02)$.

[0082] Ti and V form the carbonitride to decrease the C content not fixed as the carbonitride. Further, Ti is fixed by the formation of a sulfide, so that the value of $C + (12/14) \times N^* - (12/92.9) \times Nb - (12/47.9) \times Ti^* - (12/50.9) \times V$ is required to be 0.01-0.06%, preferably 0.01-0.05% when Ti and/or V are added in order that the C content not fixed as the carbonitride is made to 0.01-0.06%.

[0083] Cr: 0.1-1.0%

[0084] Cr is an element enhancing the hardenability by suppressing the formation of cementite and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage after the soaking at the annealing step. Further, the recrystallization of worked austenite is suppressed in the hot rolling step to promote the transformation from non-recrystallized austenite to ferrite and develop {113}<110>, and the Young's modulus can be increased at the subsequent cold rolling and annealing steps. In order to obtain such an effect, Cr is preferable to be included in an amount of not less than 0.1%. On the other hand, when the Cr content exceeds 1.0%, the above effect is saturated and the alloy cost increases, so that Cr is preferable to be included in an amount of not more than 1.0%. Moreover, when the thin steel sheet of the invention is used as a galvanized steel sheet, the oxide of Cr produced on the surface induces the non-plating, so that Cr is preferable to be included in an amount of not more than 0.5%.

[0085] Ni: 0.1-1.0%

[0086] Ni is an element stabilizing austenite to enhance the hardenability, and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage after the soaking at the annealing step. Further, Ni as an austenite stabilizing element lowers Ac_1 transformation point in the temperature rising stage at the annealing step after the cold rolling to promote the transformation from the non-recrystallized ferrite to austenite, and develops the orientation useful for the increase of the Young's modulus with respect to the orientation of the low-temperature transformation phase produced in the cooling stage after the soaking, whereby the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase can be suppressed. Since Ni is an element suppressing the recrystallization of worked austenite in the hot rolling and stabilizing austenite, when Ar_3 transformation point is lowered to conduct the rolling just above Ar_3, it is possible to conduct the rolling at a lower temperature region to further suppress the recrystallization of worked austenite, and also the transformation from the non-recrystallized austenite to ferrite is promoted to develop {113}<110>, whereby the Young's modulus can be increased at the subsequent cold rolling and annealing steps. In case of adding Cu, the surface defect is induced by cracking accompanied with the lowering of the hot ductility in the hot rolling, but the occurrence of the surface defect can be controlled by composite addition of Ni. In order to obtain such an action, Ni is preferable to be included in an amount of not less than 0.1%.

[0087] On the other hand, when the Ni content exceeds 1.0%, Ac_3 transformation point is extremely lowered in the temperature rising stage at the annealing step after the cold rolling and the recrystallization of ferrite phase at the two-phase region is difficult, and hence it is required to raise the temperature up to austenite single phase region above Ac_3 transformation point. As a result, ferrite of orientation obtained by the recrystallization of worked ferrite and useful for the increase of the Young's modulus can not be developed to bring about the decrease of the Young's modulus. And also, the alloy cost increases. Therefore, Ni is preferable to be included in an amount of not more than 1.0%.

[0088] Mo: 0.1-1.0%

[0089] Mo is an element enhancing the hardenability by making small the mobility of the interface, and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage at the annealing step after the cold rolling. Further, the recrystallization of worked austenite can be suppressed, and the transformation from the non-recrystallized austenite to ferrite is promoted to develop {113}<110> and the Young's modulus can be increased at the subsequent cold rolling and annealing steps. In order to obtain such an action, Mo is preferable to be included in an amount of not less than 0.1%. On the other hand, when the Mo content exceeds 1.0%, the above effect is saturated and the alloy cost increases, so that Mo is preferable to be included in an amount of not more than 1.0%.

[0090] B: 0.0005-0.0030%

[0091] B is an element suppressing the transformation from austenite phase to ferrite phase to enhance the hardenability, and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage at the annealing step after the cold rolling. Further, the recrystallization of worked austenite can be suppressed, and the transformation from the non-recrystallized austenite to ferrite is promoted to develop {113}<110> and the Young's modulus can be increased at the subsequent cold rolling and annealing steps. In order to obtain such an effect, B is preferable to be included in an amount of not less than 0.0005%. On the other hand, when the B content exceeds 0.0030%, the above effect is saturated, so that B is preferable to be included in an amount of not more than 0.0030%.

[0092] Cu: 0.1-2.0%

[0093] Cu is an element enhancing the hardenability, and can largely contribute to the increase of the strength by largely promoting the formation of the low-temperature transformation phase in the cooling stage at the annealing step after the cold rolling. In order to obtain such an effect, Cu is preferable to be included in an amount of not less than 0.1%. On the other hand, when the Cu content exceeds 2.0%, the hot ductility is lowered and the surface defect accompanied with the cracking in the hot rolling is induced and the hardening effect by Cu is saturated, so that Cu is preferable to be included in an amount of not more than 2.0%.

[0094] The reason on the limitation of the texture according to the invention will be described below.

[0095] In the thin steel sheet of the invention, it is required to have a texture comprising a ferrite phase as a main phase and having a martensite phase at an area ratio of not less than 1%.

[0096] The term "ferrite phase as a main phase" used herein means that the area ratio of the ferrite phase is not less than 50%.

[0097] Since the ferrite phase is less in the strain, useful for the increase of the Young's modulus, excellent in the ductility and good in the workability, the texture is required to be the ferrite phase as a main phase.

[0098] Also, in order to render the tensile strength of the steel sheet into not less than 590 MPa, it is required that the low-temperature transformation phase as a hard phase is formed in a portion other than the ferrite phase as a main phase or a so-called second phase to provide a composite phase. At this moment, the feature that a hard martensite phase among the low-temperature transformation phases is particularly existent in the texture is advantageous because the fraction of the second phase for obtaining the target tensile strength level is made small and the fraction of ferrite phase is made large, whereby the increase of the Young's modulus is attained and further the workability can be improved. For this end, the martensite phase is required to be not less than 1% as an area ratio to the whole of the texture. In order to obtain the strength of lot less than 700 MPa, the area ratio of the martensite phase is preferable to be not less than 16%.

[0099] The texture of the steel sheet according to the invention is preferable to be a texture comprising ferrite phase and martensite phase, but there is no problem that phases other than the ferrite phase and martensite phase such as bainite phase, residual austenite phase, pearlite phase, cementite phase and the like are existent at the area ratio of not more than 10%, preferably not more than 5%. That is, the sum of area ratios of ferrite phase and martensite phase is preferably not less than 90%, more preferably not less than 95%.

[0100] Next, the reason on the production conditions limited for obtaining the high-stiffness high-strength thin steel sheet according to the invention and preferable production conditions will be explained.

[0101] The composition of the starting material of steel used in the production method of the invention is the same as the composition of the aforementioned steel sheet, so that the description of the reason on the limitation of the starting material of steel is omitted.

[0102] The thin steel sheet according to the invention can be produced by successively conducting a hot rolling step of subjecting the starting material of steel having the same composition as the composition of the steel sheet to a hot rolling to obtain a hot rolled sheet, a cold rolling step of subjecting the hot rolled sheet after pickling to a cold rolling to obtain a cold rolled sheet, and an annealing step of attaining the recrystallization and composite texture in the cold rolled sheet.

[0103] (Hot Rolling Step)

[0104] Finish rolling: total rolling reduction below 950° C. is not less than 30%, and the rolling is terminated at Ar_3-900° C.

[0105] In the final rolling at the hot rolling step, the rolling is conducted just above Ar_3 transformation point to develop a non-recrystallized austenite texture having a crystal orientation of {112}<111>, and the {112}<111> non-recrystallized austenite can be transformed to ferrite in the subsequent cooling stage to develop ferrite orientation of {113}<110>. This orientation advantageously acts to the improvement of the Young's modulus in the formation of the texture at the subsequent cold rolling and annealing steps. In order to obtain such an action, it is required that the total rolling reduction below 950° C. (total rolling reduction) is not less than 30%, more preferably the total rolling reduction below 900° C. is not less than 30%, and the finish rolling is terminated at a temperature region of Ar_3-900° C., preferably Ar_3-850° C.

[0106] Coiling temperature: not higher than 650° C.

[0107] When the coiling temperature after the finish rolling exceeds 650° C., the carbonitride of Nb is coarsened and the effect of suppressing the recrystallization of ferrite becomes small in the temperature rising stage at the annealing step after the cold rolling and it is difficult to transform the non-recrystallized ferrite into austenite. As a result, the orientation of the low-temperature transformation phase transformed in the cooling stage after the soaking can not be controlled, and the Young's modulus is largely lowered by the low-temperature transformation phase having such a strain. Therefore, the coiling temperature after the finish rolling is required to be not higher than 650° C.

[0108] Moreover, when the coiling temperature is too low, a great amount of the hard low-temperature transformation phase is produced and the subsequent cold rolling becomes difficult, so that it is preferable to be not lower than 400° C.

[0109] (Cold Rolling Step)

[0110] Cold rolling is carried out at a rolling reduction of not less than 50% after the pickling.

[0111] After the hot rolling step, the pickling is carried out for removing scale formed on the surface of the steel sheet. The pickling may be conducted according to the usual manner. Thereafter, the cold rolling is conducted. By the cold rolling at a rolling reduction of not less than 50% can be turned the orientation of {113}<110> developed on the hot rolled steel sheet to an orientation of {112}<110> effective for the increase of the Young's modulus. Thus, as the orientation of {112}<110> is developed by the cold rolling, the orientation of {112}<110> in ferrite is enhanced in the texture after the subsequent annealing step and further the orientation of {112}<110> is developed in the low-temperature transformation phase, whereby the Young's modulus can be increased. In order to obtain such an effect, the rolling reduction in the cold rolling is required to be not less than 50%.

[0112] (Annealing Step)

[0113] Temperature rising rate from 500° C. to soaking temperature: 1-40° C./s, Soaking temperature: 780-900° C.

[0114] The temperature rising rate at the annealing step is an important process condition in the invention. In the course of raising the temperature to a soaking temperature of two-phase region or a soaking temperature of 780-900° C. at the annealing step, the recrystallization of ferrite having an orientation of {112}<110> is promoted, while a part of ferrite grains having an orientation of {112}<110> is arrived to a two-phase region at a non-recrystallized state, whereby the transformation from the non-recrystallized ferrite having an orientation of {112}<110> can be promoted. Therefore, the Young's modulus can be increased by promoting the growth of ferrite grains having an orientation of {112}<110> when austenite is transformed into ferrite in the cooling after the soaking. Further, when the strength is increased by producing the low-temperature transformation phase, austenite phase transformed from ferrite having an orientation of {112}<110> is re-transformed in the cooling, so that {112}<110> can be also developed with respect to the crystal orientation of the low-temperature transformation phase. By developing {112}<110> of ferrite phase is increased the Young's modulus, while {112}<110> is particularly developed in the orientation of the low-temperature transformation phase largely influencing the lowering of the Young's modulus, whereby the lowering of the Young's modulus accompanied with the formation of the low-temperature transformation phase can be suppressed while forming the low-temperature transformation phase. When austenite is transformed from the non-recrystallized ferrite while promoting the recrystallization of ferrite in the temperature rising stage, an average temperature rising rate largely exerting on the recrystallization behavior from 500° C. to 780-900° C. as a soaking temperature is required to be 1-40° C./s, preferably 1-30° C./s.

[0115] In this case, the reason why the soaking temperature is 780-900° C. is due to the fact that when it is lower than 780° C., the recrystallization is not completed, while when it exceeds 900° C., the fraction of austenite becomes large and ferrite having an orientation of {112}<110> reduces or disappears. Moreover, the soaking time is not particularly limited, but it is preferable to be not less than 30 seconds for forming austenite, while it is preferable to be not more than about 300 seconds because the production efficiency is deteriorated as the time is too long.

[0116] Cooling rate to 500° C. after soaking: not less than 5° C./s

[0117] In the cooling stage after the soaking, it is required to form the low-temperature transformation phase containing martensite for increasing the strength. Therefore, an average cooling rate to 500° C. after the soaking is required to be not less than 5° C./s.

[0118] In the invention, steel having a chemical composition in accordance with the target strength level is first melted. As the melting method can be properly applied a usual converter process, an electric furnace process and the like. The molten steel is cast into a slab, which is subjected to a hot rolling as it is or after the cooling and heating. After the finish rolling under the aforementioned finish conditions in the hot rolling, the steel sheet is coiled at the afore-mentioned coiling temperature and then subjected to usual pickling and cold rolling. As to the annealing, the temperature is raised under the aforementioned condition, and in the cooling after the soaking, the cooling rate can be increased within a range of obtaining a target low-temperature transformation phase. Thereafter, the cold rolled steel sheet may be subjected to an overaging treatment, or may be passed through a hot dip zinc in case of producing as a galvanized steel sheet, or further in case of producing as an alloyed galvanized steel sheet, a re-heating may be conducted up to a temperature above 500° C. for the alloying treatment.

EXAMPLES

[0119] The following examples are given in illustration of the invention and are not intended as limitations thereof.

[0120] At first, a steel A having a chemical composition shown in Table 1 is melted in a vacuum melting furnace of a laboratory and cooled to room temperature to prepare a steel ingot (steel raw material).

TABLE 1

Kind	Chemical composition										
of steel	C	Si	Mn	P	S	Al	N	Nb	X value	Y value	Remarks
A	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	0.03	0.011	Acceptable example

[0121] Thereafter, the hot rolling, pickling, cold rolling and annealing are successively conducted in the laboratory. The basic production conditions are as follows. After the steel ingot is heated at 1250° C. for 1 hour, the hot rolling is conducted under conditions that the total rolling reduction below 900° C., i.e. total rolling reduction ratio below 900° C. is 40% and the final rolling temperature (corresponding to a final temperature of finish rolling) is 830° C. to obtain a hot rolled sheet having a thickness of 4.0 mm. Thereafter, the coiling condition (corresponding to a coiling temperature of 600° C.) is simulated by leaving the hot rolled sheet up to 600° C. and keeping in a furnace of 600° C. for 1 hour and then cooling in the furnace. The thus obtained hot rolled sheet is pickled and cold-rolled at a rolling reduction of 60% to a thickness of 1.6 mm. Then, the temperature of the cold rolled sheet is raised at 10° C./s on average up to 500° C. and further from 500° C. to a soaking temperature of 820° C. at 5° C./s on average. Next, the soaking is carried out at 820° C. for **180** seconds, and thereafter the cooling is carried out at an average cooling rate of 10° C./s up to 500° C., and further the temperature of 500° C. is kept for 80 seconds, and then the sheet is cooled in air. Moreover, Ar_3 transformation point of this steel under the above production conditions is 730° C.

[0122] In this experiment, the following conditions are further individually changed under the above production conditions as a basic condition. That is, the experiment is carried out under the basic condition except for the individual changed conditions that the total rolling reduction below 950° C. or total rolling reduction below 900° C. is 20-65% and the final temperature of the hot finish rolling is 710-920° C. and the coiling temperature is 500-670° C. and the rolling reduction of the cold rolling is 40-75% (thickness: 2.4-1.0 mm) and the average temperature rising rate from 500° C. to the soaking temperature (820° C.) in the annealing is 0.5-45° C./s.

[0123] From the sample after the annealing is cut out a test specimen of 10 mm×120 mm in a direction perpendicular to the rolling direction as a longitudinal direction, which is finished to a thickness of 0.8 mm by a mechanical polishing and a chemical polishing for removing strain, and thereafter a resonance frequency of the sample is measured by using a lateral vibration type internal friction measuring device to calculate a Young's modulus therefrom. With respect to the sheet subjected to a temper rolling of 0.5%, a tensile test specimen of JIS No. 5 is cut out in the direction perpendicular to the rolling direction and subjected to a tensile test. Further, the sectional texture is observed by a scanning type electron microscope (SEM) after the corrosion with Nital to judge the kind of the texture, while three photographs are shot at a visual region of 30 μm×30 μm and then area ratios of ferrite phase and martensite phase are measured by an image processing to determine an average value of each phase as an area ratio (fraction) of each phase.

[0124] As a result, the values of the mechanical characteristics under the basic condition in the experiment according to the production method of the invention are Young's modulus E: 245 GPa, TS: 800 MPa, E1: 20%, fraction of ferrite phase: 70% and fraction of martensite phase: 25%, from which it is clear that the thin steel sheet has an excellent balance of strength-ductility and a high Young's modulus. Moreover, the remainder of the texture other than ferrite phase and martensite phase is either of bainite phase, residual austenite phase, pearlite phase and cementite phase.

[0125] Then, the relationship between the production conditions and Young's modulus is explained based on the above test results with reference to the drawings. Even in any experimental conditions, the tensile strength is 750-850 MPa, and the fraction of ferrite phase is 80-60%, the fraction of martensite phase is 17-40%, and the remainder of the texture of the second phase other than martensite phase is either of bainite phase, residual austenite phase, pearlite phase and cementite phase.

[0126] In FIG. **1** is shown influences of the total rolling reduction below 950° C. and the total rolling reduction below 900° C. upon Young's modulus, respectively. When the total rolling reduction below 950° C. is not less than 30% being the acceptable range of the invention, the Young's modulus indicates an excellent value of not less than 225 GPa, and further when the total rolling reduction below 900° C. is not less than 30%, the Young's modulus indicates a more excellent value of not less than 240 GPa.

[0127] In FIG. **2** is shown an influence of the final temperature of the hot finish rolling upon the Young's modulus. When the final temperature is Ar_3-900° C. being the acceptable range of the invention, the Young's modulus indicates an excellent value of not less than 225 GPa, and further when the final temperature is Ar_3-850° C., the Young's modulus indicates a more excellent value of not less than 240 GPa.

[0128] In FIG. **3** is shown an influence of the coiling temperature upon the Young's modulus. When the coiling temperature is not higher than 650° C. being the acceptable range of the invention, the Young's modulus indicates an excellent value of not less than 225 GPa.

[0129] In FIG. **4** is shown an influence of the rolling reduction of the cold rolling upon the Young's modulus. When the rolling reduction is not less than 50% being the acceptable range of the invention, the Young's modulus indicates an excellent value of not less than 225 GPa.

[0130] In FIG. **5** is shown an influence of the average temperature rising rate from 500° C. to the soaking temperature of 820° C. in the annealing upon the Young's modulus. When the temperature rising rate is 1-40° C./s being the acceptable range of the invention, the Young's modulus indicates an excellent value of not less than 225 GPa, and further when the temperature rising rate is 1-30° C./s, the Young's modulus indicates a more excellent value of not less than 240 GPa.

[0131] Furthermore, steels B-Z and AA-BF having a chemical composition as shown in Tables 2 and 3 are melted in a vacuum melting furnace of a laboratory and then successively subjected to the hot rolling, pickling, cold rolling and annealing under the above basic condition, respectively. In Tables 4 and 5 are shown characteristics obtained by the aforementioned tests. Moreover, the Ar_3 transformation point in the steels B-Z and AA-BF under the above production conditions is 650-760° C. Also, the residual texture other than ferrite phase and martensite phase in the tables is either of bainite phase, residual austenite phase, pearlite phase and cementite phase.

TABLE 2

Kind of steel	Chemical composition (mass %)												
	C	Si	Mn	P	S	Al	N	Nb	other components	X value	N*	Y value	Remarks
B	0.02	0.2	2.5	0.02	0.001	0.03	0.002	0.07	—	0.01	—	0.009	Acceptable Steel
C	0.02	0.2	2.5	0.02	0.001	0.03	0.002	0.14	—	0.00	—	0.020	Comparative Steel
D	0.06	0.2	2.5	0.02	0.001	0.03	0.002	0.12	—	0.05	—	0.017	Acceptable Steel
E	0.07	0.2	2.5	0.02	0.001	0.03	0.002	0.08	—	0.06	—	0.011	Acceptable Steel
F	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.25	—	0.01	—	0.036	Acceptable Steel
G	0.06	0.2	2.5	0.02	0.001	0.03	0.002	0.35	—	0.02	—	0.051	Acceptable Steel
H	0.05	0.2	2.5	0.02	0.001	0.03	0.002	0.05	—	0.05	—	0.006	Acceptable Steel
I	0.05	0.2	2.5	0.02	0.001	0.03	0.002	0.04	—	0.05	—	0.005	Acceptable Steel
J	0.11	0.2	2.5	0.02	0.001	0.03	0.002	0.30	—	0.07	—	0.044	Comparative Steel
K	0.04	0.2	1.4	0.02	0.001	0.03	0.002	0.08	—	0.03	—	0.011	Comparative Steel
L	0.04	0.2	1.5	0.02	0.001	0.03	0.002	0.08	—	0.03	—	0.011	Acceptable Steel

TABLE 2-continued

Kind of steel	C	Si	Mn	P	S	Al	N	Nb	other components	X value	N*	Y value	Remarks
	Chemical composition (mass %)												
M	0.04	0.2	2.0	0.02	0.001	0.03	0.002	0.08	—	0.03	—	0.011	Acceptable Steel
N	0.04	0.2	3.5	0.02	0.001	0.03	0.002	0.08	—	0.03	—	0.011	Acceptable Steel
O	0.04	0.2	3.7	0.02	0.001	0.03	0.002	0.08	—	0.03	—	0.011	Acceptable Steel
P	0.02	0.01	2.5	0.01	0.001	0.03	0.002	0.07	—	0.01	—	0.009	Acceptable Steel
Q	0.02	1.5	2.5	0.01	0.001	0.03	0.002	0.07	—	0.01	—	0.009	Acceptable Steel
R	0.02	0.2	2.5	0.01	0.001	0.5	0.002	0.07	—	0.01	—	0.009	Acceptable Steel
S	0.02	0.2	2.5	0.01	0.001	1.0	0.002	0.07	—	0.01	—	0.009	Acceptable Steel
T	0.02	0.2	2.5	0.01	0.001	1.5	0.002	0.07	—	0.01	—	0.009	Acceptable Steel
U	0.02	1.5	2.5	0.01	0.001	1.0	0.002	0.07	—	0.01	—	0.009	Acceptable Steel
V	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Ti: 0.01	0.03	0.000	0.011	Acceptable Steel
W	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Ti: 0.05	0.02	0.000	0.011	Acceptable Steel
X	0.07	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Ti: 0.18	0.02	0.000	0.011	Acceptable Steel
Y	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	V: 0.05	0.02	0.002	0.011	Acceptable Steel
Z	0.08	0.2	2.5	0.02	0.001	0.03	0.002	0.08	V: 0.20	0.02	0.002	0.011	Acceptable Steel

Note)
In case of adding no Ti or V, X value = C + (12/14) × N − (12/92.9) × Nb
In case of adding Ti or V, X value = C + (12/14) × N* − (12/92.9) × Nb − (12/47.9) × Ti* − (12/50.9) × V
Y value = (14/92.9) × (Nb − 0.01) provided that N* = N − (14/47.9) × Ti at N − (14/47.9) × Ti > 0, N* = 0 at N − (14/47.9) × Ti $\leqq$ 0,
Ti* = Ti − (47.9/14) × N − (47.9/32.1) × S at Ti − (47.9/14) × N − (47.9/32.1) × S > 0,
Ti* = 0 at Ti − (47.9/14) × N − (47.9/32.1) × S $\leqq$ 0.

TABLE 3

Kind of steel	C	Si	Mn	P	S	Al	N	Nb	other components	X value	N*	Y value	Remarks
	Chemical composition (mass %)												
AA	0.07	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Ti: 0.10, V: 0.10	0.01	0.000	0.011	Acceptable Steel
AB	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Cr: 0.1	0.03	—	0.011	Acceptable Steel
AC	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Cr: 1.0	0.03	—	0.011	Acceptable Steel
AD	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Ni: 0.2	0.03	—	0.011	Acceptable Steel
AE	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Ni: 1.0	0.03	—	0.011	Acceptable Steel
AF	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Mo: 0.2	0.03	—	0.011	Acceptable Steel
AG	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Mo: 1.0	0.03	—	0.011	Acceptable Steel
AH	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Cu: 0.3	0.03	—	0.011	Acceptable Steel
AI	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Cu: 2.0	0.03	—	0.011	Acceptable Steel
AJ	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	B: 0.0010	0.03	—	0.011	Acceptable Steel
AK	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	B: 0.0030	0.03	—	0.011	Acceptable Steel
AL	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Cr: 0.1, Ni: 0.1	0.03	—	0.011	Acceptable Steel

TABLE 3-continued

Kind of steel	C	Si	Mn	P	S	Al	N	Nb	other components	X value	N*	Y value	Remarks
	Chemical composition (mass %)												
AM	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Cr: 0.1, Mo: 0.1	0.03	—	0.011	Acceptable Steel
AN	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Cr: 0.1, B: 0.0010	0.03	—	0.011	Acceptable Steel
AO	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Cr: 0.1, Ni: 0.1, Mo: 0.1, Cu: 0.1, B: 0.0010	0.03	—	0.011	Acceptable Steel
AP	0.06	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Ti: 0.1, V: 0.05, Cr: 0.2, B: 0.0010	0.01	0.000	0.011	Acceptable Steel
AQ	0.04	0.2	2.5	0.02	0.001	0.03	0.002	0.08	Ti: 0.05, V: 0.02, Cr: 0.2, Ni: 0.2, Mo: 0.1, Cu: 0.1, B.0005	0.01	0.000	0.011	Acceptable Steel
AR	0.13	0.01	2.0	0.02	0.001	0.03	0.003	0.06	Ti: 0.15, V: 0.10	0.06	0.000	0.008	Acceptable Steel
AS	0.15	0.1	2.1	0.03	0.002	0.02	0.002	0.03	Ti: 0.24, V: 0.10	0.06	0.000	0.003	Acceptable Steel
AT	0.16	0.1	2.2	0.01	0.001	0.03	0.001	0.09	Ti: 0.31	0.07	0.000	0.012	Comparative Steel
AX	0.06	0.2	3.9	0.01	0.001	0.03	0.002	0.02	Ti: 0.02	0.05	0.000	0.002	Acceptable Steel
AY	0.07	0.01	4.2	0.02	0.002	0.05	0.001	0.05	Ti: 0.01	0.06	0.000	0.006	Comparative Steel
AZ	0.05	0.01	2.3	0.03	0.001	0.04	0.002	—	—	0.05	—	−0.002	Comparative Steel
BA	0.06	0.01	1.8	0.01	0.001	0.03	0.001	0.01	—	0.06	—	0.000	Comparative Steel
BB	0.04	0.01	1.5	0.01	0.001	0.02	0.003	0.02	Ti: 0.02	0.04	0.000	0.002	Acceptable Steel
BC	0.08	0.1	1.9	0.01	0.001	0.01	0.002	0.05	Ti: 0.25	0.01	0.000	0.006	Acceptable Steel
BD	0.14	0.1	1.8	0.02	0.002	0.02	0.002	0.05	Ti: 0.45	0.02	0.000	0.006	Acceptable Steel
BE	0.10	0.1	1.9	0.01	0.001	0.03	0.002	0.05	V: 0.20	0.05	0.000	0.006	Acceptable Steel
BF	0.12	0.01	1.8	0.02	0.002	0.04	0.001	0.05	V: 0.40	0.02	0.000	0.006	Acceptable Steel

Note)
In case of adding no Ti or V, X value = C + (12/14) × N − (12/92.9) × Nb
In case of adding Ti or V, X value = C + (12/14) × N* − (12/92.9) × Nb − (12/47.9) × Ti* − (12/50.9) × V
Y value = (14/92.9) × (Nb − 0.01) provided that N* = N − (14/47.9) × Ti at N − (14/47.9) × Ti > 0, N* = 0 at N − (14/47.9) × Ti ≦ 0,
Ti* = Ti − (47.9/14) × N − (47.9/32.1) × S at Ti − (47.9/14) × N − (47.9/32.1) × S > 0,
Ti* = 0 at Ti − (47.9/14) × N − (47.9/32.1) × S ≦ 0.

TABLE 4

Kind of steel	Steel texture: Fraction of ferrite phase (%)	Fraction of martensite phase (%)	Mechanical properties: TS (MPa)	E1 (%)	E (GPa)	Remarks
B	95	4	600	30	252	Invention Example
C	100	0	530	32	255	Comparative Example
D	50	46	990	15	240	Invention Example

TABLE 4-continued

Kind of steel	Steel texture: Fraction of ferrite phase (%)	Fraction of martensite phase (%)	Mechanical properties: TS (MPa)	E1 (%)	E (GPa)	Remarks
E	50	55	1060	12	235	Invention Example
F	75	20	860	18	243	Invention Example
G	80	19	890	16	242	Invention Example
H	70	26	750	21	240	Invention Example

TABLE 4-continued

Kind of steel	Steel texture: Fraction of ferrite phase (%)	Steel texture: Fraction of martensite phase (%)	Mechanical properties: TS (MPa)	Mechanical properties: E1 (%)	Mechanical properties: E (GPa)	Remarks
I	70	25	750	22	235	Invention Example
J	30	68	1180	10	220	Comparative Example
K	90	8	570	30	231	Comparative Example
L	85	12	590	29	241	Invention Example
M	80	17	650	28	242	Invention Example
N	60	35	860	17	242	Invention Example
O	50	50	890	16	235	Invention Example
P	98	1	590	30	253	Invention Example
Q	90	7	630	30	248	Invention Example
R	94	3	620	29	242	Invention Example
S	94	3	630	29	241	Invention Example
T	93	4	640	28	240	Invention Example
U	92	4	650	27	240	Invention Example
V	70	25	810	20	246	Invention Example
w	75	23	780	21	247	Invention Example
X	73	24	810	19	245	Invention Example
Y	72	22	800	20	246	Invention Example
Z	68	28	890	15	243	Invention Example

TABLE 5

Kind of steel	Steel texture: Fraction of ferrite phase (%)	Steel texture: Fraction of martensite phase (%)	Mechanical properties: TS (MPa)	Mechanical properties: E1 (%)	Mechanical properties: E (GPa)	Remarks
AA	85	13	780	20	248	Invention Example
AB	65	30	810	19	245	Invention Example
AC	60	36	850	17	242	Invention Example
AD	64	30	810	19	245	Invention Example
AE	58	37	860	17	241	Invention Example
AF	65	31	820	18	243	Invention Example
AG	59	37	870	17	241	Invention Example
AH	67	29	810	20	243	Invention Example
AI	60	33	840	17	242	Invention Example
AJ	60	34	850	17	243	Invention Example
AK	50	43	900	15	241	Invention Example
AL	63	34	820	18	242	Invention Example
AM	61	34	820	18	241	Invention Example
AN	59	37	860	17	242	Invention Example
AO	57	38	870	16	240	Invention Example
AP	82	15	760	22	248	Invention Example
AQ	84	13	800	20	247	Invention Example
AR	70	25	900	15	235	Invention Example
AS	68	30	950	13	226	Invention Example
AT	45	55	920	14	213	Comparative Example
AX	60	35	850	16	225	Invention Example
AY	40	60	1000	13	208	Comparative Example
AZ	75	20	760	21	210	Comparative Example
BA	73	21	770	20	215	Comparative Example
BB	80	18	700	24	226	Invention Example
BC	70	28	820	17	245	Invention Example
BD	73	25	920	14	240	Invention Example

TABLE 5-continued

Kind of steel	Steel texture: Fraction of ferrite phase (%)	Steel texture: Fraction of martensite phase (%)	Mechanical properties: TS (MPa)	Mechanical properties: E1 (%)	Mechanical properties: E (GPa)	Remarks
BE	80	20	800	20	241	Invention Example
BF	70	28	890	17	243	Invention Example

[0132] In the steel C, the C content (X-value) not fixed as a carbonitride is as small as 0.00%, and the ferrite phase is 100%, and the fraction of the second phase is 0%, and TS is smaller than the acceptable range of the invention. In the steel J, the X-value is as high as 0.07%, and the Young's modulus is smaller than the acceptable range of the invention. In the steel K, the Mn content is as low as 1.4%, and TS is smaller than the acceptable range of the invention. In the steel AT, the C content is as high as 0.16%, and the X-value is as high as 0.07, and the Young's modulus is smaller than the acceptable range of the invention. In the steel AZ, the Mn content is as large as 4.2%, and the Young's modulus is smaller than the acceptable range of the invention. In the steel AZ, Nb is not contained, while in the steel BA, the Mb content is as small as 0.01%, so that the Young's modulus is smaller than the acceptable range of the invention.

[0133] With respect to the other steels, all items are within the acceptable range of the invention, and TS and Young's modulus satisfy the acceptable range of the invention.

INDUSTRIAL APPLICABILITY

[0134] According to the invention, it is possible to provide high-stiffness high-strength thin steel sheets having a tensile strength of not less than 590 MPa and a Young's modulus of not less than 225 GPa.

1. A high-stiffness high-strength thin steel sheet comprising C: 0.02-0.15%, Si: not more than 1.5%, Mn: 1.5-4.0%, P: not more than 0.05%, S: not more than 0.01%, Al: not more than 1.5%, N: not more than 0.01% and Nb: 0.02-0.40% as mass %, provided that C, N and Nb contents satisfy the relationships of the following equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N-(12/92.9)\times Nb \leqq 0.06 \tag{1}$$

$$N \leqq (14/92.9)\times(Nb-0.01) \tag{2}$$

and the remainder being substantially iron and inevitable impurities, and having a texture comprising a ferrite phase as a main phase and having a martensite phase at an area ratio of not less than 1%, and having a tensile strength of not less than 590 MPa and a Young's modulus of not less than 225 GPa.

2. A high-stiffness high-strength thin steel sheet according to claim **1**, which further contains one or two of Ti: 0.01-0.50% and V: 0.01-0.50% as mass % in addition to the above composition and satisfy the relationships of the following equations (3) and (4) instead of the equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N^*-(12/92.9)\times Nb-(12/47.9)\times Ti^*-(12/50.9)\times V \leqq 0.06 \tag{3}$$

$$N^* \leqq (14/92.9)\times(Nb-0.01) \tag{4}$$

provided that N* in the equations (3) and (4) is $N^*=N-(14/47.9)\times Ti$ at $N-(14/47.9)\times Ti>0$ and $N^*=0$ at $N-(14/47.9)\times$

$Ti \leqq 0$, and Ti* in the equation (3) is $Ti^*=Ti-(47.9/14)\times N-(47.9/32.1)\times S$ at $Ti-(47.9/14\times N-(47.9/32.1)\times S>0$ and $Ti^*=0$ at $Ti-(47.9/14)\times N-(47.9/32.1)\times S \leqq 0$.

3. A high-stiffness high-strength thin steel sheet according to claim **1** or **2**, which further contains one or more of Cr: 0.1-1.0%, Ni: 0.1-1.0%, Mo: 0.1-1.0%, Cu: 0.1-2.0% and B: 0.0005-0.0030% as mass % in addition to the above composition.

4. A method for producing a high-stiffness high-strength thin steel sheet comprising subjecting a starting material of steel comprising C: 0.02-0.15%, Si: not more than 1.5%, Mn: 1.5-4.0%, P: not more than 0.05%, S: not more than 0.01%, Al: not more than 1.5%, N: not more than 0.01% and Nb: 0.02-0.40% as mass %, provided that C, N and Nb contents satisfy the relationships of the following equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N-(12/92.9)\times Nb \leqq 0.06 \quad (1)$$

$$N \leqq (14/92.9)\times(Nb-0.01) \quad (2)$$

to a hot rolling step under conditions that a total rolling reduction below 950° C. is not less than 30% and a finish rolling is terminated at Ar_3-900° C., coiling the hot rolled sheet below 650° C., pickling, subjecting to a cold rolling at a rolling reduction of not less than 50%, raising a temperature to 780-900° C. at a temperature rising rate from 500° C. of 1-40° C./s to conduct soaking, and then cooling at a cooling rate up to 500° C. of not less than 5° C./s to conduct annealing.

5. A method for producing a high-stiffness high-strength thin steel sheet according to claim **4**, wherein the starting material of steel further contains one or two of Ti: 0.01-0.50% and V: 0.01-0.50% as mass % in addition to the above composition and satisfies the relationships of the following equations (3) and (4) instead of the equations (1) and (2):

$$0.01 \leqq C+(12/14)\times N^*-(12/92.9)\times Nb-(12/47.9)\times Ti^*-(12/50.9)\times V \leqq 0.06 \quad (3)$$

$$N^* \leqq (14/92.9)\times(Nb-0.01) \quad (4)$$

provided that N* in the equations (3) and (4) is $N^*=N-(14/47.9)\times Ti$ at $N-(14/47.9)\times Ti>0$ and $N^*=0$ at $N-(14/47.9)\times Ti \leqq 0$, and Ti* in the equation (3) is $Ti^*=Ti-(47.9/14)\times N-(47.9/32.1)\times S$ at $Ti-(47.9/14)\times N-(47.9/32.1)\times S>0$ and $Ti^*=0$ at $Ti-(47.9/14)\times N-(47.9/32.1)\times S \leqq 0$.

6. A method for producing a high-stiffness high-strength thin steel sheet according to claim **4** or **5**, wherein the staring material of steel further contains one or more of Cr: 0.1-1.0%, Ni: 0.1-1.0%, Mo: 0.1-1.0%, Cu: 0.1-2.0% and B: 0.0005-0.0030% as mass % in addition to the above composition.

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