



(11) **EP 1 735 474 B1**

(12) **EUROPEAN PATENT SPECIFICATION**

(45) Date of publication and mention
of the grant of the patent:
21.10.2015 Bulletin 2015/43

(51) Int Cl.:
C21D 9/46 (2006.01) **C22C 38/00** (2006.01)
C22C 38/02 (2006.01) **C22C 38/04** (2006.01)
C22C 38/06 (2006.01) **C22C 38/12** (2006.01)

(21) Application number: **05789750.6**

(86) International application number:
PCT/KR2005/000828

(22) Date of filing: **22.03.2005**

(87) International publication number:
WO 2006/001583 (05.01.2006 Gazette 2006/01)

(54) **COLD ROLLED STEEL SHEET AND HOT DIPPED STEEL SHEET WITH SUPERIOR STRENGTH AND BAKE HARDENABILITY AND METHOD FOR MANUFACTURING THE STEEL SHEETS**

KALTGEWALZTES STAHLBLECH UND FEUERVEREDELTES STAHLBLECH MIT HOHER
FESTIGKEIT UND WARMHÄRTBARKEIT UND VERFAHREN ZU HERSTELLUNG DER
STAHLBLECHE

FEUILLE D'ACIER LAMINEE A FROID ET FEUILLE D'ACIER TRAITEE A CHAUD PRESENTANT
UNE RESISTANCE ET UN DURCISSEMENT A LA CUISSON SUPERIEURS ET PROCEDE DE
FABRICATION DE CES FEUILLES D'ACIER

(84) Designated Contracting States:
DE FR

(30) Priority: **25.03.2004 KR 20040020204**
09.12.2004 KR 20040103610

(43) Date of publication of application:
27.12.2006 Bulletin 2006/52

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• **DATABASE WPI Week 200376 Thomson**
Scientific, London, GB; AN 2003-809358
XP002597317 -& KR 2003 053 802 A 2 July 2003
(2003-07-02)

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Description**Technical Field**

5 [0001] The present invention relates to a bake-hardenable cold-rolled steel sheet for automobile bodies, a hot-dipped steel sheet manufactured using the same, and a method for manufacturing the same. More particularly, the present invention relates to a bake-hardenable high-strength cold-rolled steel sheet having excellent bake hardenability, aging resistance at room temperature, and secondary work embrittlement resistance, a hot-dipped steel sheet manufactured using the same, and a method for manufacturing the same.

Background Art

10 [0002] As used for exterior plates of automobile bodies, bake-hardenable cold-rolled steel sheets typically have a tensile strength of 390 MPa, and exhibit excellent ductility due to the fact that their yield strength is approximately the same as that of mild steel upon press forming. Additionally, the bake-hardenable cold-rolled steel sheets have a characteristic in which the yield strength thereof is increased upon paint baking or coating treatment after press forming.

[0003] As for the bake-hardenable cold-rolled steel sheets, steel sheets are mainly used, which are manufactured by batch annealing low carbon, P-added, Al-killed steels, and have a bake hardening value of about 40 to 50 MPa.

20 [0004] However, when manufacturing the bake-hardenable cold-rolled steel sheets through batch annealing, although compatibility between the formability and the bake hardenability can be easily ensured, there is a problem of lowering the productivity.

[0005] In order to improve the productivity based on the batch annealing method, a method of manufacturing the bake-hardenable cold-rolled steel sheets by continuous annealing of the low carbon, P-added, Al-killed steels has been developed.

25 [0006] However, when manufacturing the bake-hardenable cold-rolled steel sheets through the continuous annealing method, although the bake hardenability can be easily ensured due to a relatively rapid cooling rate of the method, there is a problem of deterioration in the formability due to rapid heating and short period of annealing.

[0007] * Accordingly, the bake-hardenable cold-rolled steel sheets manufactured by the continuous annealing method are restrictively used for the exterior plates of the automobiles, which do not require the formability.

30 [0008] Recently, according to the rapid progress of steel manufacturing technology, it is possible to optimize the amount of solute elements in a steel.

[0009] Thus, it is possible to manufacture bake-hardenable cold-rolled steel sheets with improved formability by addition of intensive carbonitride-forming elements, such as Ti or Nb, to the Al-killed steel. As such, as the bake-hardenable cold-rolled steel sheets manufactured by the continuous annealing method are improved in the formability, the bake-hardenable cold-rolled steel sheets are increasingly utilized as exterior plates for automobiles which require dent resistance.

35 [0010] Japanese Unexamined Patent Publication No. (Sho) 61-26757 discloses an ultra low carbon bake-hardenable cold-rolled steel sheet, which comprises 0.0005 ~ 0.015 wt% of C; 0.05 wt% or less of S+N; and Ti and Nb or a compound thereof.

[0011] Japanese Unexamined Patent Publication No. (Sho) 57-89437 discloses a bake-hardenable cold-rolled steel sheet, which uses a Ti-added steel comprising 0.010 wt% or less of C, and has a bake hardening value of about 40 MPa.

[0012] The methods of the disclosures impart bake hardenability to the steel sheets while preventing deterioration of the other properties of the steel sheets by appropriately controlling the amount of solute elements in the steel through appropriate control of the added amount of Ti and Nb or the cooling rate during annealing.

45 [0013] However, according to the disclosures, since it is necessary to strictly control the amount of Ti, N, and S in the manufacturing process in order to ensure appropriate bake hardenability, there is a problem of increasing manufacturing costs.

[0014] On the other hand, U.S. Patent Nos. 5,556,485 and 5,656,102 disclose methods of manufacturing a bake-hardenable cold-rolled steel sheet using a Ti-V based ultra low carbon steel, which comprises 0.0005 - 0.1 wt% of C; 0 ~ 2.5 wt% of Mn; 0 - 0.5 wt% of Al; 0 ~ 0.04 wt% of N; 0 ~ 0.5 wt% of Ti; and 0.005 ~ 0.6 % of V.

50 [0015] In these disclosures, although V can create a carbide such as VC, since it does not sufficiently improve the formability due to its too low re-melting temperature, Ti is added in an amount of about 0.02 wt% or more in order to enhance the formability.

[0016] Thus, in the case of U.S. Patent Nos. 5,556,485 and 5,656,102, there are problems of deterioration in aging resistance caused by coarse grains as well as an increase of the manufacturing costs due to addition of large amounts of Ti.

55 [0017] Meanwhile, Japanese Unexamined Patent Publication No. (Hei) 5-93502 discloses a method for enhancing the bake hardenability through addition of Sn, and Japanese Unexamined Patent Publication No. (Hei) 9-249936 discloses a method for enhancing the ductility of the steel by relieving stress concentration on grain boundaries through addition of V and Nb.

[0018] Additionally, Japanese Unexamined Patent Publication No. (Hei) 8-49038 discloses a method for enhancing the formability through addition of Zr, and Japanese Unexamined Patent Publication No. (Hei) 7-278654 discloses a method for enhancing the formability by increasing the strength while minimizing deterioration of work hardening index (N-value) through addition of Cr.

[0019] However, these methods are designed only to enhance either the bake hardenability or the formability through addition of alloy elements, and do not simultaneously enhance the bake hardenability, the aging resistance, and the secondary work embrittlement resistance.

[0020] In the methods of the above disclosures, although boron, known as an element capable of enhancing the secondary work embrittlement resistance, is added in an amount of about 5 ppm to the steel, it is limited in its ability to enhance the secondary work embrittlement resistance due to the presence of excessive P content.

[0021] Furthermore, if B is excessively added to the steel to enhance the secondary work embrittlement resistance, deterioration will occur due to an excessive amount of B, and thus, the amount of B which can be added to the steel is limited.

[0022] Accordingly, since the ductility-brittleness transition temperature (DBTT) must be - 20°C or more in order to prevent the secondary work embrittlement, there are needs in the art to add other components to the bake-hardenable steel other than B, and to examine new manufacturing conditions.

[0023] On the other hand, it is necessary to enhance corrosion resistance in order to extend life span of automobile components.

[0024] For this purpose, the steel sheets are usually subjected to electroplating or hot dipping.

[0025] Although electroplated steel sheet has good plating characteristics and excellent corrosion resistance, the electroplated steel sheet is very expensive in comparison to hot dipped steel sheets, and thus the hot dipped steel sheets are mainly used for automobiles.

[0026] Accordingly, in order to improve the corrosion resistance of the hot-dipped steel sheets, studies on a hot-dipping process and the hot-dipped steel sheets have been continued, and usage of the hot-dipped steel sheets has also been increased

Disclosure of Invention

Technical Problem

[0027] Therefore, the present invention has been made in view of the above problems, and it is an object of the invention to provide bake-hardenable high-strength cold-rolled steel sheets having bake hardenability, aging resistance, and secondary work embrittlement resistance by adding a small amount of Nb together with Al, Mo and B, while appropriately controlling of the Nb/C ratio and the grain size, hot-dipped steel sheets manufactured using the same, and a method for manufacturing the same.

Technical Solution

[0028] In accordance with one aspect of the present invention, the above and other objects can be accomplished by the provision of a bake-hardenable cold-rolled steel sheet manufactured through hot rolling, cold rolling and continuous annealing of a steel, the steel sheet comprising: 0.0016 ~ 0.01 % of C; 0.1 % or less of Si; 0.2 ~ 1.5 % of Mn; 0.05 ~ 0.15 % of P; 0.01 % or less of S; 0.08 ~ 0.5 % of (soluble) Al; 0.0025 % or less of N; 0.003 ~ 0.1 % of Nb optionally 0.003 % of Ti; 0.01 ~ 0.4 % of Mo; 0.0005 ~ 0.005 % of B; and the balance of Fe and other unavoidable impurities, in terms of weight%, while satisfying an Nb/C ratio of 0.3 ~ 0.7, wherein the steel sheet has fine AlN precipitates formed upon hot rolling the steel, and a grain size (ASTM No.) of 9 or more, the fine AlN precipitates having a size acting as a barrier for suppressing grain growth during annealing of the steel sheet.

[0029] In accordance with another aspect of the invention, a hot-dipped steel sheet manufactured through hot rolling, cold rolling, continuous annealing, and hot dipping of a steel is provided, the steel sheet consisting of : 0.0016~0.01% of C; 0.1% or less of Si; 0.2 ~ 1.5% of Mn; 0.05 ~ 0.15 % of P; 0.01 % or less of S; 0.08 ~ 0.5% of (soluble)Al; 0.0025% or less of N; 0.003 ~ 0.1% of Nb; 0 ~ 0.003 % of Ti; 0.01 ~ 0.4% of Mo; 0.0005 ~ 0.005% of B; and the balance of Fe and other unavoidable impurities, in terms of weight%, while satisfying an Nb/C ratio of 0.3 ~ 0.7, wherein the steel sheet has fine AlN precipitates formed upon hot rolling the steel, and a grain size (ASTM No.) of 9 or more, the fine AlN precipitates having a size acting as a barrier for suppressing grain growth during annealing of the steel sheet.

[0030] In accordance with still another aspect of the present invention, a method for manufacturing a bake-hardenable high-strength cold-rolled steel sheet, the method comprising the steps of: hot-rolling a steel slab with finish rolling at or above the A_{r3} transformation temperature to provide a hot rolled steel sheet after heating the steel slab to a temperature of 1,200 °C or more, the steel slab consisting of : 0.0016 ~ 0.01 % of C; 0.1 % or less of Si; 0.2 ~ 1.5 % of Mn; 0.05 ~ 0.15 % of P; 0.01 % or less of S; 0.08 ~ 0.5 % of (soluble) Al; 0.0025 % or less of N; 0.003 ~ 0.1 % of Nb; 0 ~ 0.003 %

of Ti; 0.01 ~ 0.4 % of Mo; 0.0005 ~ 0.005 % of B; and the balance of Fe and other unavoidable impurities, in terms of weight%, while satisfying an Nb/C ratio of 0.3 ~ 0.7; coiling the hot-rolled steel sheet; cold rolling the hot-rolled steel sheet; and continuous annealing the cold-rolled steel sheet.

[0031] In accordance with yet another aspect of the invention, a method for manufacturing a hot-dipped steel sheet, comprising the steps of: hot-rolling a steel slab with finish rolling at or above an Ar_3 transformation temperature to provide a hot rolled steel sheet after heating the steel slab to a temperature of 1,200 °C or more, the steel slab comprising: 0.0016 ~ 0.01 % of C; 0.1 % or less of Si; 0.2 ~ 1.5 % of Mn; 0.05 ~ 0.15 % of P; 0.01 % or less of S; 0.08 ~ 0.5 % of (soluble) Al; 0.0025 % or less of N; 0.003 ~ 0.1 % of Nb; 0 ~ 0.003 % of Ti; 0.01 ~ 0.4 % of Mo; 0.0005 ~ 0.005 % of B; and the balance of Fe and other unavoidable impurities, in terms of weight%, while satisfying an Nb/C ratio of 0.3 ~ 0.7; coiling the hot-rolled steel sheet; cold rolling the hot-rolled steel sheet; continuous annealing the cold-rolled steel sheet; hot dipping the annealed steel sheet; and temper rolling the hot-dipped steel sheet.

Brief Description of the Drawings

[0032] The above and other objects, features and other advantages of the present invention will be more clearly understood from the following detailed description taken in conjunction with the accompanying drawings, in which:

Fig. 1 is a graph illustrating the influence of grain size upon bake hardening value and aging index;

Fig. 2 is a graph illustrating the influence of Al content upon mechanical properties;

Fig. 3 is a graph illustrating the influence of Mo content upon the bake hardening value and the aging index;

Fig. 4 is a micrograph of microstructure of Inventive steel No. 8 after annealing;

Fig. 5 is a graph illustrating the influence of drawing ratio upon DBTT for the Inventive steel No. 8, Comparative steel No. 6, and a conventional steel;

Fig. 6 is a graph illustrating stress-strain curves before and after aging the Inventive steel No. 8; and

Fig. 7 is a graph illustrating the influence of the Nb/C ratio upon the bake hardening value and the aging index for steels of the invention.

Best Mode for Carrying Out the Invention

[0033] Preferred embodiments of the invention will now be described in detail.

[0034] Requirements for a bake-hardenable cold-rolled steel sheet include a high bake hardening value, a low aging index at room temperature, and excellent secondary work embrittlement resistance.

[0035] Bake hardening or natural aging is a phenomenon occurring when interstitial elements existing as solute elements in a steel, particularly, nitrogen or carbon, are fixed to dislocations generated during deformation. When contents of solute nitrogen and carbon are increased in the steel, a bake hardening value is also increased, with accompanying natural aging due to the excessive amount of solute elements, thereby causing deterioration of formability. Thus, it is very important to optimize the amount of solute nitrogen and carbon in the steel.

[0036] Since nitrogen has a higher diffusion speed in the steel than that of carbon, nitrogen affects the natural aging rather than carbon.

[0037] Thus, the amount of solute nitrogen is reduced in order to secure excellent aging resistance.

[0038] As for a method for reducing the amount of solute nitrogen in the steel, it is well-known in the art to lower the amount of nitrogen in the steel as much as possible or to precipitate nitrogen contained in the steel into a nitride in forms of TiN and AlN as much as possible by adding Ti and Al, nitride formation elements, to the steel, thereby reducing the amount of solute nitrogen in the steel.

[0039] For the solute carbon, if the amount of solute carbon is excessively lowered, the aging resistance is enhanced, whereas the bake hardenability is deteriorated. On the other hand, if the amount of solute carbon is excessively increased, the bake hardenability is enhanced, whereas the aging resistance is deteriorated.

[0040] Thus, in order to secure excellent aging resistance and bake hardenability at the same time, it is necessary to appropriately control the amount of solute carbon in the steel.

[0041] As for a method for controlling the amount of solute carbon in the steel, it is well-known to control the amount of carbon in the steel or to precipitate carbon contained in the steel into a carbide in forms of TiC, $Ti_4C_2S_2$ and NbC as much as possible by adding Ti or Nb, carbide formation elements, to the steel.

[0042] According to the invention, in order to allow the solute carbon to be appropriately contained in the steel, it is necessary to control the Nb/C ratio.

[0043] Additionally, in the case of the solute carbon, even though the content thereof is contained in the same amount, the bake hardenability and the aging resistance may be differently exhibited according to a position of the solute carbon in the steel, that is, according to whether the solute carbon is located in grain boundaries or in grains, which will be described hereinafter.

[0044] The reason for the varying influence on the bake hardenability and the aging resistance according to the position of the solute carbon in the steel is related to the mobility of the solute carbon.

[0045] When the solute carbon is located in grain, it can move relatively freely, thereby influencing not only the bake hardenability but also the aging resistance.

[0046] On the other hand, since the solute carbon in the grain boundary is located in a relatively stable position, it has minimal effects upon low temperature aging, but during high temperature baking, it is activated and affects the bake hardenability.

[0047] As such, the solute carbon in grain influences the aging resistance and the bake hardenability simultaneously, whereas the solute carbon in the grain boundary influences only the bake hardenability.

[0048] Accordingly, appropriate control of the position of the solute carbon in the steel, that is, controlling the solute carbon to be located in the grain boundaries rather than the grains can secure the aging resistance and the bake hardenability at the same time.

[0049] Thus, it is important to control the position of the solute carbon in the steel together with an appropriate content of the solute carbon in the steel in order to secure both excellent aging resistance and bake hardenability.

[0050] One of the features of the invention is appropriate control of the amount of solute carbon and the position thereof in the steel.

[0051] That is, according to the invention, the contents of carbon and niobium are appropriately controlled in order to control the amount of solute carbon, and the grains are controlled to have a fine grain size so as to allow as much of the solute carbon in the steel as possible to be located in the grain boundaries.

[0052] As mentioned above, when the grains become fine, the area of the grain boundaries is further increased, and an increased amount of solute carbon present in the steel can be located in the grain boundaries, thereby securing the excellent aging resistance and the bake hardenability simultaneously.

[0053] Fig. 1 is a graph illustrating the relationship between variation in grain size and bake hardening (BH) value and an aging index (AI), as one of the results of studies performed by the inventors.

[0054] As shown in Fig. 1, as an ASTM No. of grain is increased, that is, as the grains become finer, the AI is remarkably reduced in comparison to the BH value, causing value of the BH value - AI to be gradually increased. As a result, it can be appreciated from Fig. 1 that as the grains become finer, the AI increases.

[0055] On the basis of the result shown in Fig. 1, the inventors attempted to refine the grains of annealed steel sheet below a predetermined value to make as much solute carbon present in the grain boundaries in the steel as possible.

[0056] That is, the inventors understood from the result of the studies that it is desirable to control the grains to have a grain size (ASTM No.) of 9 or more in order to minimize deterioration of the bake hardenability while maximizing the aging resistance.

[0057] Another feature of the invention is that the grains become finer through formation of fine AlN precipitates upon hot rolling.

[0058] That is, according to the invention, the fine AlN precipitates are formed upon hot rolling, by addition of large quantities of Al to the steel, and suppress grain growth when annealing a cold-rolled steel sheet, thereby allowing the grains to become finer.

[0059] Thus, according to the invention, the Al content is appropriately controlled so as to allow the fine AlN precipitates to be formed upon hot rolling.

[0060] Additionally, fine grains are more securely achieved by appropriately controlling the size of the AlN precipitates.

[0061] Al is known as a deoxidizer and/or a component added for fixing nitrogen by use of coarse AlN precipitates in a bake-hardenable steel.

[0062] The bake-hardenable steel, wherein Al and N are combined to form the fine AlN precipitates upon hot rolling through appropriate control of Al content, thereby refining the grains, so that the solute carbon in the steel is located in the grain boundaries rather than in the grains, thereby ensuring the excellent aging resistance and bake hardenability at the same time, is not known in the art.

[0063] Since there are demands for steels which exhibit no detectable change in aging resistance for 6 months or more together with a higher BH value, there is a need for a technology, which can enhance the bake hardenability while minimizing deterioration of the aging resistance.

[0064] In this aspect, the inventor appreciates that Al is a very effective element for providing the bake-hardenable steel.

[0065] That is, when soluble aluminum is added in the typical range of 0.02 ~ 0.06 % to a steel to which Nb has been added and to which Ti has been added, in the range of 0-30 ppm, if at all, coarse AlN precipitates are formed in the steel, and simply serve to fix the solute nitrogen. However, when the soluble aluminum is added in the range of 0.08 % or more to the steel, very fine AlN precipitates are formed, and act as a kind of barrier suppressing grain growth upon recrystallization annealing, so that the grains of the steel are further reduced in size than a conventional Nb-added steel with the soluble aluminum added according to a conventional method, thereby providing an effect of enhancing the bake hardenability without changing the Al.

[0066] Of course, in a Ti-added steel, nitrogen is precipitated into TiN precipitates, most of which are coarse, at a high

temperature of 1,300 °C or more, and thus does not significantly influence solid solution or grain refining.

[0067] In Fig. 2, an example of variation of the bake hardening (BH) value and the aging index (AI) of a hot-dipped steel sheet according to variation of the amount of soluble aluminum is shown.

[0068] It can be seen from Fig. 2 that as Al content is increased, the BH value is increased, whereas the AI varies only slightly.

[0069] In view of the bake hardenability, it can be appreciated from Fig. 2 that the soluble aluminum content is preferably in the range of 0.08 % or more.

[0070] According to one of the results obtained by analyzing enhanced effects of the bake hardenability through addition of Al as described above, BH is mainly affected by the contents of Ti, Nb and Al, as can be appreciated from Equation 1;

$$BH = 50 - (885 * Ti) - (1589 * Nb) + (62 * Al) \text{ --- (1)}$$

[0071] Additionally, according to the invention, a hot coiling temperature is appropriately selected for grain refining.

[0072] When the hot coiling temperature is too high, the grains become coarse during coiling even if the composition of the steel, that is, contents of carbon, soluble aluminum, and niobium, is controlled. As a result, coarsening of the grains occurs during recrystallization annealing so that the grains have a grain size (ASTM No.) of 9 or less, thereby increasing the AI. On the contrary, when the hot coiling temperature is lowered below a certain degree, aging resistance at room temperature may be enhanced, but grain refining becomes excessive, possibly causing an increase of yield strength and deterioration of formability, by which elongation and r-value are decreased.

[0073] Still another feature of the invention is enhancement of secondary work embrittlement resistance through addition of appropriate amounts of Mo.

[0074] Generally, components for automobiles are formed to have a desired shape through repeated pressing.

[0075] Secondary work embrittlement means that cracks are formed during a secondary working process after a primary working process.

[0076] The cracks are caused by a phenomenon in which fractures are formed along the grain boundaries due to phosphorous (P), which is present in the grain boundaries of the steel and weakens coupling force between the grains.

[0077] In order to eliminate the secondary work embrittlement, it is basically desirable that P is not added to the steel. However, P is the solute element which provides the least reduction of elongation in comparison to an increase of strength, and, in particular, has an advantage in that it is low in price.

[0078] As a result, for the purpose of strengthening the steel, although the fact is accepted that P must be basically added to the steel, investigations also have recently been progressed to strengthen the steel via addition of other solute elements instead of P in order to eliminate the secondary work embrittlement while ignoring the slight increase in manufacturing costs resulting therefrom.

[0079] However, according to results of the investigations, it is anticipated that P will continue to be used as an element for strengthening steel for the foreseeable future.

[0080] As a method for enhancing the secondary work embrittlement in a P-added steel, investigations have been progressed for preventing the secondary work embrittlement through an increase of site competition effect against P or the bonding force of the grains boundaries by adding B to the steel or by allowing the solute elements to remain in the steel, such as the bake-hardenable steel, or through minimization of grain boundary diffusion of P by lowering a coiling temperature below a predetermined temperature during hot rolling, but have not yet provided a perfect solution to these problems.

[0081] Accordingly, the present invention employs Mo in order to further enhance the secondary work embrittlement resistance.

[0082] According to the results of the investigations of the inventors, it can be found that since Mo enhances the bonding force of the grains boundaries, it is very advantageous for enhancement of the secondary work embrittlement resistance.

[0083] Further, since Mo has an affinity to the solute elements in the steel, it suppresses diffusion of the solute elements toward dislocations when the steel is maintained for a long time at room temperature, thereby enhancing the aging resistance.

[0084] Fig. 3 is a graph showing one of the results obtained by analyzing the effect of enhancing the aging resistance through addition of Mo, and it can be seen from Fig. 3 that when Mo content is increased, the BH value is not significantly changed, whereas the AI is lowered, thereby enhancing the aging resistance.

[0085] According to one of the results obtained by analyzing the effect of enhancing the aging resistance through addition of Mo, the AI is mainly affected by the contents of Ti, Nb and Mo, as can be appreciated from Equation 2;

$$AI = 44 - (423 * Ti) - (2119 * Nb) - (125 * Mo) \text{ --- (2)}$$

[0086] The inventors induce an optimal composition of the elements in order to use the properties of Mo and to prevent the deterioration of the property through excessive addition of Mo.

[0087] Additionally, according to the invention, the B content is appropriately controlled in order to enhance the secondary work embrittlement resistance.

[0088] Meanwhile, according to the invention, various well-known technologies may be additionally applied in order to enhance the secondary work embrittlement resistance, one example of which is appropriate control of coiling temperature.

[0089] Compositions of the steel according to the invention will now be described in detail.

[0090] Carbon (C): C is an element for solid solution strengthening and bake hardening. If carbon content is excessively low, the tensile strength of the steel is significantly lowered, sufficient bake hardening effect is not secured due to a low absolute amount of carbon in the steel, and the secondary work embrittlement resistance is also deteriorated since the site competition effect between the solute carbon and P is removed.

[0091] Thus, the carbon content is preferably 0.0016 % or more in order to ensure an effect of carbon addition.

[0092] On the other hand, if the carbon content is too high, since the aging resistance at room temperature cannot be secured due to excessive quantities of solute carbon remaining in the steel, stretcher strain occurs during press forming, thereby reducing the formability and ductility of the steel. Thus, the upper limit of the carbon content is set to 0.01 %, and preferably to 0.0025 %.

[0093] According to the invention, when the carbon content is in the range of 0.0016 ~ 0.0025 %, the ability to control the carbon content to be in this range is realized in a level of 95 % or more in practice.

[0094] Silicon (Si): Si is an element for increasing the strength of the steel. However, as the silicon content is increased, deterioration of ductility becomes remarkable, and since silicon deteriorates hot dipping capability, it is advantageous to minimize the quantity of silicon added to the steel.

[0095] Thus, according to the invention, in order to prevent deterioration of the properties including coating properties of the steel due to silicon, the silicon content is restricted to 0.1 % or less, and particularly to 0.02 % or less.

[0096] Manganese (Mn): Mn is an element for preventing hot embrittlement caused by formation of FeS by completely precipitating sulfur in the steel into MnS while allowing the grains to be reduced in size without deteriorating the ductility, as well as strengthening the steel.

[0097] In order to achieve sufficient effects from addition of manganese, Mn is preferably added in the range of 0.2 % or more.

[0098] Meanwhile, if the Mn content exceeds 1.5 %, the formability is deteriorated due to solid solution strengthening as the strength of the steel is rapidly increased. In particular, when manufacturing a hot-dipped steel sheet, a great amount of oxides, such as MnO, and a number of coating defects, such as a stripe pattern, are formed on the surface of the steel sheet during an annealing process, thereby deteriorating the properties of the steel including coating adherence.

[0099] Accordingly, the Mn content is preferably in the range of 0.2 ~ 1.5 %, and more particularly, in the range of 0.2 ~ 1.2 %.

[0100] Phosphorus (P): P is a substitutional alloying element, which provides the highest solid solution strengthening effect, and serves to enhance the in-plane anisotropy index while improving the strength of the steel.

[0101] With the results of the investigation of the inventors, P causes the grains of a hot-rolled steel sheet to become finer, and then serves to promote development of the microstructure having (111) plane, which is advantageous in that it enhances the average r-value (plasticity-anisotropy index), during a subsequent annealing process. In particular, it can be found that, due to the site competition effect between carbon and phosphorous in view of the bake hardenability: the greater the amount of P the higher the bake hardenability.

[0102] However, the increase of phosphorous content causes the coupling force of the grain boundaries to be weakened, thereby deteriorating the secondary work embrittlement resistance.

[0103] Accordingly, it is necessary to restrict the phosphorous content to within a range of 0.05 ~ 0.15 %, and preferably in the range of 0.05 ~ 0.11 %.

[0104] Sulfur (S): Sulfur content is preferably in the range of 0.01 % or less.

[0105] Aluminum (Al): Aluminum is an alloying element considered an important element of the present invention.

[0106] Al is generally used for deoxidizing and fixing the solute nitrogen in the steel. However, in the present invention, Al is combined with N during hot rolling to form fine AlN precipitates in the steel, thereby suppressing the grain growth and thus promoting grain refining during the annealing process, thereby enhancing the aging resistance and bake hardenability.

[0107] For an Nb-added bake-hardenable steel, grain refining is mainly achieved by NbC precipitates.

[0108] However, according to the invention, since the fine AlN precipitates suppress the grain growth together with the NbC precipitates during the annealing process, the grains can be further reduced in size, thereby providing excellent aging resistance and bake hardenability.

[0109] In the invention, it is necessary to set the Al content so as to provide the fine AlN precipitates during hot rolling.

[0110] When the Al content is below 0.08 %, since it is difficult to provide the fine AlN precipitates during hot rolling, it is necessary to set a lower limit of the Al content to 0.08 %.

[0111] Preferably, the Al content exceeds 0.1 %, and thus finer Al precipitates can be secured by addition of Al exceeding 0.1 %.

[0112] Considering formability, surface quality affected by oxidized inclusions, and manufacturing costs, an upper limit of the Al content is preferably 0.5 %, and more preferably 0.12 %.

[0113] It is necessary for the AlN precipitates to have a size, which can serve as a barrier to suppress the grain growth during annealing of the cold-rolled steel sheets, and it is preferable that the AlN precipitates have an average size of 20 μm or less.

[0114] When the AlN precipitates have the average size of 20 μm or less as described above, finer grains are obtained.

[0115] Nitrogen (N): Nitrogen exists in a solute state before or after annealing, and deteriorates the formability of the steel. Furthermore, since nitrogen has a higher aging ability than other interstitial solute elements, it is necessary to fix nitrogen by use of Ti or Al.

[0116] In the case of the invention wherein an appropriate amount of Nb is added to the steel together with a small amount of Ti, excessive addition of nitrogen causes generation of solute nitrogen in the steel.

[0117] Thus, the nitrogen content is restricted to 0.0025 % or less.

[0118] Titanium (Ti): Titanium is typically added to the steel as a carbonitride-forming element, and forms nitrides such as TiN, sulfides such as TiS or $\text{Ti}_4\text{C}_2\text{S}_2$, and carbides such as TiC, in the steel.

[0119] However, according to the invention, Ti is not added to the steel, and, if at all, the amount of Ti added to the steel is restricted to 0.003 % or less, which can fix a small amount of nitrogen under consideration that Ti is contained

[0120] Niobium (Nb): Niobium is a very important element together with Al and Mo in the present invention.

[0121] Nb is an intensive carbonitride-forming element, and fixes carbon in the steel as NbC precipitates. In particular, since the NbC precipitates are very fine in comparison to other precipitates, it acts as an intensive barrier to impede the grain growth during recrystallization annealing, thereby causing the grains to become finer.

[0122] When the grains are reduced in size as described above, the bake hardenability and the secondary work embrittlement resistance are enhanced.

[0123] However, in order to enhance the bake hardenability, not only is the grain refining effect required, but a predetermined amount of solute carbon is also required.

[0124] Accordingly, according to the invention, the grains are refined by means of the NbC precipitates, and the amount of Nb is controlled so as to secure the bake hardenability by allowing the predetermined amount of solute carbon to remain in the steel. The Nb content is preferably in the range of 0.003 ~ 0.1 %, and more preferably in the range of 0.003 ~ 0.011 %.

[0125] It is desirable that the size of NbC precipitates is 30 nm or less.

[0126] Additionally, in order to more appropriately control the bake hardenability and the aging resistance at room temperature, it is necessary to control not only the Nb content but also the Nb/C ratio at the same time, and according to the invention, it is necessary to control the Nb/C ratio to be in the range of 0.3 ~ 0.7.

[0127] In the case of an Nb/C ratio of less than 0.3, the bake hardening value is lowered due to lack of an absolute amount of solute carbon in the steel, and in the case of a Nb/C ratio of greater than 0.7, even though the bake hardening value is increased due to an increase of the amount of solute carbon in the steel, the aging index is also increased, thereby deteriorating the aging resistance at room temperature. Accordingly, it is necessary to limit the ratio of Nb/C to within the range of 0.3 ~ 0.7.

[0128] Molybdenum (Mo): Molybdenum is another very important element in the present invention.

[0129] Mo serves to enhance the strength of the steel or to form an Mo-based carbide in the solute state in the steel. In particular, Mo serves to increase the coupling force of the grain boundaries in the solute state in the steel, thereby preventing fracture of the grain boundaries caused by P, that is, enhancing the secondary work embrittlement resistance, and suppresses the diffusion of carbon by virtue of affinity with carbon, thereby enhancing the aging resistance.

[0130] However, addition of too much Mo will cause an increase of manufacturing costs.

[0131] Thus, considering the effect of Mo addition upon manufacturing costs, the Mo content is preferably in the range of 0.01 ~ 0.4 %, and more preferably in the range of 0.01 ~ 0.1 %.

[0132] Boron (B): Boron is an interstitial element, and is located within the steel. B is in the solute state in the grain boundaries or combined with nitrogen to form the nitride of BN.

[0133] Since B has a significant influence upon the properties of the steel according to an added amount, it is necessary to precisely control the amount added.

[0134] That is, when a small quantity of B is added, B is segregated in the grain boundaries, and enhances the secondary work embrittlement resistance.

[0135] However, when B is added to the steel above a predetermined amount, the steel can be remarkably deteriorated in ductility while being subjected to an increase of the strength, and thus it is necessary to add an appropriate amount of B.

[0136] Thus, according to the invention, considering these characteristics and an ability of manufacturing the steel through addition of B, the amount of B is preferably in the range of 0.0005 ~ 0.005 %, and more preferably in the range of 0.0005 ~ 0.0015 %.

[0137] The steel sheet of the invention must have a grain size (ASTM No.) of 9 or more while satisfying the compositions as described above.

[0138] As such, when the steel sheet has the grain size of ASTM number of 9 or more, the amount of solute carbon in the grain boundaries is increased, thereby allowing the bake hardenability to be further enhanced while maintaining excellent aging resistance at room temperature.

[0139] On the other hand, when the steel sheet has a grain size of ASTM number less than 9, sufficient effect of enhancing the bake hardenability cannot be desired.

[0140] The steel sheet of the invention has a greater amount of solute carbon in the grain boundaries rather than in the grains.

[0141] In the present invention, a preferable amount of solute carbon in grain is approximately 3 ~ 6 ppm.

[0142] Additionally, the size of AlN precipitates formed during hot rolling is preferably 20 μm or less.

[0143] When the size of AlN precipitates is 20 μm or less, the grains become finer, resulting in further enhancement of the bake hardenability.

[0144] The steel sheet of the invention has excellent bake hardenability, aging resistance, secondary work embrittlement resistance, and a tensile strength of 300 MPa or more.

[0145] A preferable embodiment of a method for manufacturing steel of the invention will now be described.

[0146] After manufacturing a molten steel having the composition as described above, a steel slab is manufactured through continuous casting of the molten steel, and is then heated to a temperature of 1,200 °C or more. The heated steel slab is subjected to hot-rolling with finish rolling at or above the Ar_3 transformation temperature, preferably at a temperature of 900 ~ 950 °C, which is immediately above the Ar_3 transformation temperature, thereby providing a hot rolled steel sheet

[0147] Heating of the steel slab to the temperature of 1,200 °C or more is for the purpose of providing sufficiently homogeneous austenite structure to the steel slab before hot rolling.

[0148] That is, when the steel slab is at a temperature less than 1,200 °C, the structure of the steel may have combined grains instead of homogeneous austenite grains, causing deterioration of the properties of the steel.

[0149] If the finish hot rolling temperature is too low, the top and tail portions, and edges of a hot coil become single-phase regions, respectively, thereby increasing the in-plane anisotropy while deteriorating the formability of the steel. If the finish hot rolling temperature is excessively high, remarkably coarse grains are formed in the steel, so that defects such as orange peels can be easily formed on the surface of the steel sheet after working.

[0150] Thus, the finish hot rolling temperature must be established on the basis of these viewpoints as described above.

[0151] Meanwhile, when a hot coiling temperature is excessively high, sufficient effect of grain refining cannot be obtained due to an increase in the size of grains after annealing, and the secondary work embrittlement can easily occur due to an increase of segregation of phosphorous in the grain boundaries. When the hot coiling temperature is too low, the aging resistance and secondary work embrittlement resistance are enhanced due to the grain refining effect, whereas the yield strength is excessively increased together with deterioration of the formability.

[0152] Thus, the hot coiling temperature must be established on the basis of these viewpoints as described above.

[0153] According to the invention, the hot coiling temperature is in the range of 550 ~ 650 °C, and more preferably in the range of 600 ~ 650 °C.

[0154] After the hot rolled steel sheet is subjected to acid pickling, cold rolling is performed, thereby manufacturing a cold-rolled steel sheet.

[0155] At this time, a reduction rate is preferably 70 ~ 80 %.

[0156] The reduction rate is maintained at 70% or more for the purpose of enhancing the formability of the steel sheet, in particular, the r-value, together with the aging resistance by virtue of grain refining.

[0157] Meanwhile, when the reduction rate is above 80%, grain refining occurs, and becomes excessive together with gradual reduction of the r-value due to the excessive reduction rate, causing the material to be hardened.

[0158] Then, the cold-rolled steel sheet is subjected to continuous annealing by a typical method.

[0159] Continuous annealing of the invention is performed at about a temperature of 770 ~ 830 °C.

[0160] Since the Nb-added steel has a higher recrystallization temperature than the Ti-added steel, and the present invention provides the Nb-added steel, annealing is preferably performed at a temperature of 770 °C or more.

[0161] That is, when annealing is performed at a temperature less than 770 °C, the yield strength is increased while the elongation and the r-value are deteriorated due to existence of non-recrystallized grains.

[0162] On the other hand when annealing is performed at a temperature exceeding 830 °C, the formability is enhanced. However, in this case, since the grain size (ASTM No.) is less than 9, which is the desired ASTM No. of grain, the AI is 30 MPa or more, and the aging resistance is deteriorated.

[0163] The annealed cold-rolled steel sheet is cooled to room temperature after continuous annealing, and is then subjected to hot dipping. Alternatively, while performing continuous annealing, the cold-rolled steel sheet is cooled from the continuous annealing temperature to the hot dipping temperature, and is then subjected to hot dipping.

[0164] According to the invention, hot dipping is not limited to particular conditions, and is performed by a typical method.

[0165] A hot dipping temperature is typically in the range of about 450 ~ 500 °C.

[0166] Next, the hot dipped steel sheet is subjected to temper rolling.

[0167] At this time, temper rolling is performed in order to ensure the aging resistance at room temperature together with appropriate bake hardenability. The temper rolling ratio is a little higher than typical temper rolling, and preferably in the range of 1.2 ~ 1.5 %.

[0168] That is, the reason of a little higher temper rolling ratio than the typical temper rolling is for preventing deterioration of the aging resistance due to the solute carbon in the steel.

[0169] However, when the drawing ratio of temper rolling is excessively high, even though the aging resistance is enhanced, work hardening occurs, thereby deteriorating the properties of the steel. In particular, when manufacturing the hot-dipped steel sheet using the steel sheet of the invention, excessive temper rolling results in deterioration of the coating adherence, thereby causing a coated layer to be peeled off.

[0170] Thus, the temper rolling ratio must be established in view of the aging resistance at room temperature and the coating adherence.

[0171] The invention will be described in detail with reference to examples.

Example 1

[0172] Steel slabs having compositions as shown in Table 1 were heated to a temperature of 1,220 °C, and subjected to finish hot rolling at a temperature of 920 °C in order to provide hot rolled steel sheets.

[0173] The hot rolled steel sheets were coiled under the conditions shown in Table 2, and subjected to cold rolling and annealing. Then, the annealed cold-rolled steel sheets were subjected to hot dipping at a temperature of 450 °C, followed by temper rolling at a temper rolling ratio of about 1.5 %. Next, a bake hardening (BH) value, an aging index (AI), a grain size (ASTM No.), a size of AlN precipitates, and a ductility-brittleness transition temperature (DBTT) were measured. Results thereof are shown in Table 2.

[0174] In Table 2, the DBTT was measured for evaluating the secondary work embrittlement, and measured at a drawing ratio of 2.0.

[0175] For Inventive steel No. 8, the microstructure was observed at a magnification of 200 times after annealing the steel sheet, and the results of observation are shown in Fig. 4.

[0176] For the Inventive steel No. 8, Comparative steel No. 6, and conventional steel, variation in DBTT according to variation in drawing ratio was investigated, and the results are shown in Fig. 5.

[0177] At this time, the conventional steel is a steel comprising 0.0019 wt% of C; 0.63 wt% of Mn; 0.056 wt% of P; 0.03 wt% of soluble Al; 0.005 wt% of Ti; 0.006 wt% of Nb, and 0.014 wt% of N.

[0178] Additionally, tensile strengths of the Inventive steel No. 8 before and after annealing were measured, and stress-strain curves thereof are shown in Fig. 6.

[0179] In Fig. 6, an as-received curve is a stress-strain curve for the tensile strength of the steel measured immediately after being manufactured, and a 180-days aged curve is a stress-strain curve for the tensile strength of the steel measured after transportation of the steel to Thailand and storage for 6 months.

[0180] During transportation to the tropics and storage of the Inventive steel No. 8 in Thailand, the Inventive steel No. 8 was maintained at an average temperature of 32 °C and a humidity of about 70 %.

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Table 1

Sample No.	Composition (wt%)												
		C	Mn	Si	P	S	Sol.Al	Ti	Nb	N	Mo	B	Nb/C
IS	1	0.0024	0.58	0.01	0.069	0.0082	0.087	0	0.008	0.0016	0.034	0.0005	0.43
	2	0.0022	0.63	0.009	0.068	0.0081	0.098	0	0.01	0.0024	0.048	0.0005	0.59
	3	0.0023	0.45	0.011	0.062	0.0058	0.105	0.0025	0.0082	0.0019	0.061	0.0007	0.46
	4	0.0024	0.61	0.008	0.071	0.0083	0.118	0.0015	0.0073	0.0015	0.059	0.0005	0.39
	5	0.0017	0.98	0.01	0.096	0.0070	0.105	0	0.004	0.0017	0.051	0.0008	0.3
	6	0.0019	1.01	0.015	0.10	0.0063	0.089	0	0.005	0.0020	0.062	0.0009	0.34
	7	0.0021	1.10	0.007	0.099	0.0073	0.105	0	0.007	0.0013	0.045	0.0005	0.43
	8	0.0023	0.98	0.013	0.091	0.0057	0.104	0.001	0.011	0.0013	0.098	0.0007	0.62
	9	0.0020	0.43	0.01	0.030	0.008	0.30	0.003	0.008	0.0017	0.049	0.0005	0.52
	10	0.0022	0.32	0.008	0.035	0.007	0.42	0.002	0.01	0.0021	0.021	0.0006	0.59
CS	1	0.0054	0.64	0.01	0.069	0.0071	0.082	0.001	0.011	0.0017	0.021	0.0007	0.26
	2	0.0022	0.63	0.014	0.066	0.0085	0.040	0.025	0.009	0.0015	0.015	0.0005	0.57
	3	0.0012	0.65	0.011	0.070	0.0072	0.075	0.001	0.0105	0.0019	0.059	0.0008	1.13
	4	0.0021	0.93	0.008	0.096	0.0089	0.043	0	0.022	0.0017	0.021	0.0006	1.35
	5	0.0022	0.59	0.009	0.062	0.0066	0.071	0.002	0.009	0.0022	0	0.0007	0.53
	6	0.0019	0.99	0.007	0.099	0.0078	0.041	0.001	0.008	0.0021	0	0	0.54
	7	0.0020	0.62	0.015	0.047	0.0085	0.021	0	0.010	0.0019	0	0	0.65
	8	0.0023	0.98	0.016	0.120	0.0078	0.098	0.001	0.009	0.0023	0.031	0	0.50
	9	0.0024	0.54	0.01	0.059	0.0087	0.29	0	0.005	0.0022	0	0.0007	0.48
	10	0.0069	0.43	0.009	0.056	0.010	0.36	0.003	0.010	0.0019	0.015	0.0005	0.19

Table 2

Sample No.		CT (°C)	Reduction (%)	AT (°C)	BH value (MPa)	AI (MPa)	Grain Size (ASTM No.)	DBTT (°C)	AIN Size (μm)
IS	1	620	78	805	42.7	22.8	10.5	-50	20 or less
	2	620	77	820	40.2	16.8	9.8	-40	20 or less
	3	620	78	825	41.3	17.9	9.9	-50	20 or less
	4	610	76	800	44.4	20.5	10.5	-60	20 or less
	5	640	78	820	50.2	29.1	10.0	-40	20 or less
	6	620	78	820	47.6	25.7	11.1	-50	20 or less
	7	620	75	830	45.4	23.5	11.5	-40	20 or less
	8	620	78	820	38.1	8.0	11.3	-40	20 or less
	9	620	78	820	53.2	19.7	12.5	-70	20 or less
	10	620	77	820	58.4	19.3	11.7	-60	20 or less

(continued)

Sample No.		CT (°C)	Reduction (%)	AT (°C)	BH value (MPa)	AI (MPa)	Grain Size (ASTM No.)	DBTT (°C)	AIN Size (μm)
CS	1	620	78	810	62.0	51.2	11.2	-60	20 or less
	2	640	78	820	13.1	12.5	8.1	10	20 or more
	3	620	78	810	0	0	8.2	20	20 or more
	4	630	76	800	0	0	9.1	20	20 or more
	5	620	78	810	38.3	24.1	10.9	0	20 or more
	6	630	77	820	36.4	26.6	9.2	-10	20 or more
	7	620	76	810	34.1	22.8	9.5	-10	20 or more
	8	640	78	820	40.9	20.6	9.8	0	20 or less
	9	620	77	820	60.0	33.4	11.2	-60	20 or less
	10	640	77	820	63.8	39.7	11.7	-70	20 or less
IS: Inventive steel, CS: Comparative steel, CT: Coiling temperature, AT: Annealing temperature									

[0181] As shown in Table 2, it can be appreciated that Inventive steel Nos. 1 to 10 have a grain size (ASTM No.) of 9.8 ~ 12.5, and have AIN precipitates with an average size of 20 μm or less.

[0182] It can be appreciated from Fig. 4 that the Inventive steel No. 8 has very fine grains while having a very uniform distribution of grains over the microstructure.

[0183] As shown in Table 2, the grains of the inventive steels are very fine, and this is because the inventive steels have a greater content of Al than the comparative steel so that fine AIN precipitates are formed in the inventive steels, and suppress grain growth upon recrystallization annealing together with NbC precipitates.

[0184] Due to such grain refining effect, the bake hardening value is in the range of 38.1 ~ 58.4 MPa, and the AI for evaluating the aging resistance at room temperature is in the range of 8.0 ~ 29.1 MPa. Accordingly, it can be appreciated that balance between the BH value and the aging resistance is excellent.

[0185] Meanwhile, it seems that a relatively low AI compared with a relatively high bake hardening value is based on retarding effect of solute carbon in the steel caused by addition of Mo in addition to grain refining effect caused by the AIN precipitates.

[0186] Additionally, in view of the secondary work embrittlement, it can be seen that the DBTT at the drawing ratio of 2.0 is in the range of -40 ~ -70 °C.

[0187] As shown in Fig. 5, it can be seen that the Inventive steel No. 8 has an excellent DBTT caused by an increase of coupling force between the grains due to addition of Mo, in comparison to Comparative steel No. 6 and the conventional steel.

[0188] As shown in Fig. 6, it can be seen that the Inventive steel No. 8 has an excellent aging resistance under long-term duration of 6 months or more at a high temperature

[0189] Thus, it can be appreciated that the inventive steels have excellent bake hardenability and secondary work embrittlement resistance, and are not deteriorated due to aging under long-term duration of 6 months or more at a high temperature of the tropic region.

[0190] Meanwhile, Comparative steel No. 1 has a carbon content of 0.0054 wt% higher than the carbon content of the invention in the range of 0.0016 ~ 0.0025 %. Although the Comparative steel No. 1 has an excellent DBTT and a significantly high BH value, it has an AI of 30 MPa, and thus, it can be seen that the aging resistance is significantly deteriorated. That is, although the Comparative steel No. 1 has a grain size of ASTM No. 11.2, and has very fine grains that satisfy the requirement for grains size of the invention, the high content of solute carbon in the steel causes the aging resistance of the Comparative steel No. 1 to be deteriorated.

[0191] Additionally, Comparative steel No. 2 has a soluble Al content of 0.04 wt%, which is lower than the Al content of the invention in the range of 0.08 ~ 0.12 %, and a Ti content of 0.025 %, which is higher than the Ti content of the invention. The Comparative steel No. 2 has large grains, a low BH value, and a deteriorated DBTT.

[0192] That is, for the Comparative steel No. 2, the grain refining effect by virtue of the AIN precipitates, and the effect of enhancing the BH value were not achieved. Furthermore, the high content addition of Ti causes most of the carbon added to the steel to be precipitated as TiC, so that the bake hardenability is hardly imparted to the steel, and reduction of the solute carbon in the steel also causes the site competition effect between C and P to be lowered, thereby deteri-

orating the DBTT.

[0193] Comparative steel No. 3 satisfies the composition of the invention, but has a carbon content of 0.0012 wt%, which is lower than the carbon content of the invention.

[0194] It can be seen that, due to reduction in the absolute amount of carbon in the steel, the Comparative steel No. 3 has coarse grains, low BH value and AI, and a deteriorated DBTT of 20 °C.

[0195] Comparative steel No. 4 has a soluble Al content lower than that of the invention, and an Nb content higher than that of the invention.

[0196] For the Comparative steel No. 4, the grain refining effect by virtue of Al, and the effect of enhancing the BH value were not achieved. Furthermore, although the Comparative steel No. 4 has a grain size of ASTM No. 9.1 that satisfies the requirement for grain size of the invention, it can be seen that the excessive addition of Nb causes excessive generation of NbC precipitates, and lack of the solute carbon in the steel, so that the BH value is not obtained at all, and the DBTT is significantly deteriorated.

[0197] Compared with the composition of the invention, Comparative steel No. 5 does not comprise Mo and B at all, and the effect of enhancing the secondary work embrittlement resistance by virtue of Mo and B is thus not expected in the Comparative steel No. 5.

[0198] That is, although the Comparative steel No. 5 has excellent BH and AI values, it has a significantly deteriorated DBTT due to non-addition of Mo and B.

[0199] Comparative steel No. 6 has a soluble Al content lower than that of invention, and does not comprise Mo at all. It can be seen that although the Comparative steel No. 6 has excellent BH value and AI, it has a deteriorated DBTT due to decrease of the coupling force between the grains caused by non-addition of Mo compared with a high content of P.

[0200] Comparative steel No. 7 has a soluble Al content lower than that of invention, and does not comprise Mo and M at all. The Comparative steel No. 7 has a BH of 34.1, and an AI of 22.8. It has a deteriorated DBTT due to non-addition of Mo and B.

[0201] Comparative steel No. 8 has a P content 0.12 % higher than that of the invention in the range of 0.05 ~ 0.11 %, and does not comprise B. It can be seen that, although the DBTT of the Comparative steel No. 8 is enhanced by virtue of Mo, the effect of enhancing the DBTT is limited due to high content addition of P, and in particular, the DBTT is still negative due to non-addition of B.

[0202] Comparative steel No. 9 does not comprise Mo, and has an AI of 33.4 MPa. It can be seen that the aging resistance is deteriorated.

[0203] Comparative steel No. 10 has an Nb/C ratio lower than that of the invention. Although it has excellent bake hardenability due to a high BH value, it also has deteriorated aging resistance due to a high AI.

Example 2

[0204] Samples were manufactured following the same compositions and manufacturing conditions as those of the Inventive steel No. 8 as shown in Tables 1 and 2 except that the ratio of Nb/C was changed as shown in Fig. 7. Then, the BH value and AI were measured according to variation of the ratio of Nb/C, and results thereof are shown in Fig. 7.

[0205] It can be seen from Fig. 7 that the Nb/C ratio must be in the range of 0.3 ~ 0.7 in order to secure a BH value of 30 MPa or more and an AI of 30 MPa or less.

Industrial Applicability

[0206] As apparent from the above description, according to the present invention, the bake-hardenable high-strength cold-rolled steel sheets having excellent bake hardenability, aging resistance, and secondary work embrittlement resistance, and the hot-dipped steel sheets using the same may be provided.

[0207] Although the preferred embodiments of the present invention have been disclosed for illustrative purposes, those skilled in the art will appreciate that various modifications, additions and substitutions are possible, without departing from the scope of the invention as disclosed in the accompanying claims.

Claims

1. A bake-hardenable high-strength cold-rolled steel sheet manufactured through hot rolling, cold rolling and continuous annealing of a steel, consisting of 0.0016~0.01% of C; 0.1% or less of Si; 0.2~1.5% Mn; 0.05~0.15% of P; 0.01% or less of S; 0.08~0.5% of soluble Al; 0.0025% or less of N; 0.003~0.1% of Nb; 0.01~0.4% of Mo; 0.0005~0.005% of B, optionally 0.003% or less of Ti and the balance of Fe and other unavoidable impurities, in terms of weight%, while satisfying an Nb/C atomic ratio of 0.3~0.7, wherein the steel sheet has NbC precipitates having an average size of 30nm or less, 0nm exempt, and the steel sheet has a grain size ASTM No of 9 or more, a tensile strength

of 300MPa or more, and fine AlN precipitates formed upon hot rolling the steel, the fine AlN precipitates having an average size of 20 μ m or less, 0 μ m exempt, acting as a barrier for suppressing grain growth during annealing of the steel sheet.

2. The steel sheet according to claim 1, wherein the Al content is greater than 0.1 % and equal to or less than 0.5 %.

3. The steel sheet according to claim 1 or 2, wherein the steel sheet comprises greater amounts of solute carbon in the grain boundary than in the grain, and the amount of solute carbon in the grain is in the range of 3 - 6 ppm.

4. The steel sheet according to claim 1, consisting of 0.0016~0.0025% of C; 0.02% or less of Si; 0.2~1.2% Mn; 0.05~0.11% of P; 0.01% or less of S; 0.08~0.12% of soluble Al; 0.0025% or less of N; 0.003~0.011% of Nb; 0.01~0.1% of Mo; 0.0005~0.0015% of B, optionally 0.003% or less of Ti and the balance of Fe and other unavoidable impurities, in terms of weight%, a bake hardening value BH of 30MPa or more, an aging index AI of 30MPa or less, a ductility-brittleness transition temperature DBTT of -30°C or less at a drawing ratio of 2.0, and a tensile strength of 340~390MPa.

5. The steel sheet according to claim 4, wherein the steel sheet comprises greater amounts of solute carbon in the grain boundary than in the grain, and the amount of solute carbon in the grain is in the range of 3 - 6 ppm.

6. The steel sheet according to claim 1, wherein the steel sheet is additionally processed in hot dipping method, thereby the steel sheet is a hot-dipped steel sheet.

7. The steel sheet according to claim 6, wherein the Al content is greater than 0.1 % and equal to or less than 0.5 %.

8. The steel sheet according to claim 6 or 7, wherein the steel sheet comprises greater amounts of solute carbon in the grain boundary than in the grain, and the amount of solute carbon in the grain is in the range of 3 - 6 ppm.

9. The steel sheet according to claim 6, wherein the hot-dipped steel sheet consists of:

0.0016-0.0025% of C; 0.02% or less of Si; 0.2~1.2% of Mn; 0.05~0.11% of P; 0.01% or less of S; 0.08-0.12% of soluble Al; 0.0025% or less of N; 0.003~0.011 % of Nb; 0.01~0.1 % of Mo; 0.0005~0.0015% of B, optionally 0.003% or less of Ti and the balance of Fe and other unavoidable impurities, in terms of weight%, a bake hardening value BH of 30MPa or more, an aging index AI of 30MPa or less, a ductility-brittleness transition temperature DBTT of -30°C or less at a drawing ratio of 2.0, and a tensile strength of 340~390MPa.

10. The steel sheet according to claim 9, wherein the steel sheet comprises greater amounts of solute carbon in the grain boundary than in the grain, and the amount of solute carbon in the grain is in the range of 3 - 6 ppm.

11. A method for manufacturing a bake-hardenable high-strength cold-rolled steel sheet having, a grain size ASTM No of 9 or more, a tensile strenght of 300MPa or more, the steel sheet having NbC precipitates having an average size of 30nm or less, 0nm exempt, and fine AlN precipitates having an average size of 20 μ m or less, 0 μ m exempt; the method comprising the steps of: hot-rolling a steel slab with finish rolling at or above an Ar3 transformation temperature to provide a hot rolled steel sheet after heating the steel slab to a temperature of 1,200°C or more, the steel slab consisting of 0.0016~0.01% of C, 0.1% or less of Si, 0.2~1.5% of Mn, 0.05~0.15% of P, 0.01% or less of S, 0.08~0.5% of soluble Al, 0.0025% or less of N, 0.003~0.1% of Nb, 0.01~0.4% of Mo, 0.0005~0.005% of B, optionally 0.003% or less of Ti and the balance of Fe and other un-avoidable impurities, in terms of weight%, while satisfying and Nb/C atomic ratio of 0.3~0.7; coiling the hot-rolled steel sheet at a temperature of 550°C to 650°C; cold rolling the hot-rolled steel sheet; and continuous annealing the cold-rolled steel sheet at a temperature of 770°C to 830°C.

12. The method according to claim 11, comprising the steps of:

hot-rolling a steel slab with finish rolling at a temperature of 900~950°C to provide a hot rolled steel sheet, after homogenizing the steel slab at a temperature of 1200°C or more, the steel slab consisting of 0.0016~0.0025% of C, 0.02% or less of Si, 0.2~1.2% of Mn, 0.05~0.11% of P, 0.01% or less of S, 0.08-0.12% of soluble Al, 0.0025% or less of N, 0.003~0.011% of Nb, 0.01~0.1% of Mo, 0.0005~0.0015% of B, optionally 0.003% or less of Ti and the balance of Fe and other un-avoidable impurities, in terms of weight%;
Coiling the hot-rolled steel sheet at a temperature of 600~650°C;

Cold rolling the hot-rolled steel sheet at a reduction rate of 75~80%; and Continuous annealing the cold-rolled steel sheet at a temperature of 770~830°C.

13. The method according to claim 11, further comprising the steps of:

hot dipping the annealed steel sheet; and
temper rolling the hot-dipped steel sheet.

14. The method according to claim 13, comprising the steps of:

hot-rolling a steel slab with finish rolling at a temperature of 900~950°C to provide a hot rolled steel sheet, after homogenizing the steel slab at a temperature of 1200°C or more, the steel slab consisting of 0.0016~0.0025% of C, 0.02% or less of Si, 0.2~1.2% of Mn, 0.05~0.11 % of P, 0.08~0.12% of soluble Al, 0.003~0.011 % of Nb, 0.01~0.1% of Mo, 0.0005~0.0015% of B, optionally 0.003% or less of Ti and the balance of Fe and other unavoidable impurities, in terms of weight%;,
hot-rolling a steel slab with finish rolling at a temperature of 900~950°C to provide a hot rolled steel sheet, after homogenizing the steel slab at a temperature of 1,200°C or more,
coiling the hot-rolled steel sheet at a temperature of 600~650°C;
cold rolling the hot-rolled steel sheet at a reduction rate of 75~80%; continuous annealing the cold-rolled steel sheet at a temperature of 770~830°C;
hot dipping the annealed steel sheet; and
temper rolling the hot-dipped steel sheet at a reduction rate of 1.2~1.5%.

Patentansprüche

1. Einbrennhärtbares, hochfestes kaltgewalztes Stahlblech, das durch Warmwalzen, Kaltwalzen und kontinuierliches Glühen eines Stahls hergestellt ist, bestehend aus 0,0016 ~ 0,01 % C; 0,1 % oder weniger Si; 0,2 ~ 1,5 % Mn; 0,05 ~ 0,15 % P; 0,01 % oder weniger S; 0,08 ~ 0,5 % löslichem Al; 0,0025 % oder weniger N; 0,003 ~ 0,1 % Nb; 0,01 ~ 0,4 % Mo; 0,0005 ~ 0,005 % B, gegebenenfalls 0,003 % oder weniger Ti und der Rest aus Fe und anderen unvermeidbaren Verunreinigungen, in Form von Massenanteilen, unter Erfüllung eines Nb/C-Atomverhältnisses von 0,3 ~ 0,7, wobei das Stahlblech NbC-Ausscheidungen mit einer durchschnittlichen Größe von 30 nm oder weniger aufweist, 0 nm ausgenommen, und das Stahlblech eine Korngröße von ASTM-Nr. 9 oder größer, eine Zugfestigkeit von 300 MPa oder mehr und feine AlN-Ausscheidungen aufweist, die bei dem Warmwalzen des Stahls gebildet werden, wobei die feinen AlN-Ausscheidungen eine durchschnittliche Größe von 20 µm oder weniger aufweisen, 0 µm ausgenommen, die als eine Barriere zum Unterdrücken des Kornwachstums während des Glühens des Stahlblechs wirken.
2. Stahlblech nach Anspruch 1, wobei der Al-Gehalt größer als 0,1 % und gleich oder weniger als 0,5 % ist.
3. Stahlblech nach Anspruch 1 oder 2, wobei das Stahlblech größere Mengen an gelöstem Kohlenstoff in der Korngrenze als im Korn aufweist und die Menge des gelösten Kohlenstoffs in dem Korn im Bereich von 3 - 6 ppm liegt.
4. Stahlblech nach Anspruch 1, bestehend aus 0,0016 ~ 0,0025 % C; 0,02 % oder weniger Si; 0,2 ~ 1,2 % Mn; 0,05 ~ 0,11 % P; 0,01 % oder weniger S; 0,08 ~ 0,12 % löslichem Al; 0,0025 % oder weniger N; 0,003 ~ 0,011 % Nb; 0,01 ~ 0,1 % Mo; 0,0005 ~ 0,0015 % B;,, gegebenenfalls 0,003 % oder weniger Ti und der Rest Fe und andere unvermeidbare Verunreinigungen, in Form von Massenanteilen, einem Bake-Hardening-Wert BH von 30 MPa oder mehr, einem Alterungsindex AI von 30 MPa oder weniger, einer Spröd-Duktil-Übergangstemperatur DBTT von -30 °C oder weniger bei einem Ziehverhältnis von 2,0 und einer Zugfestigkeit von 340 ~ 390 MPa.
5. Stahlblech nach Anspruch 4, wobei das Stahlblech größere Mengen an gelöstem Kohlenstoff in der Korngrenze als im Korn aufweist und die Menge an gelöstem Kohlenstoff in dem Korn im Bereich von 3 - 6 ppm liegt.
6. Stahlblech nach Anspruch 1, wobei das Stahlblech zusätzlich durch ein Schmelztauchverfahren bearbeitet ist, wodurch das Stahlblech ein feuerverzinktes Stahlblech ist.
7. Stahlblech nach Anspruch 6, wobei der Al-Gehalt größer als 0,1 % und gleich oder weniger als 0,5 % ist.

8. Stahlblech nach Anspruch 6 oder 7, wobei das Stahlblech größere Mengen an gelöstem Kohlenstoff in der Korngrenze als im Korn aufweist und die Menge an gelöstem Kohlenstoff in dem Korn im Bereich von 3 - 6 ppm liegt.

9. Stahlblech nach Anspruch 6, wobei das feuerverzinkte Stahlblech aus 0,0016 ~ 0,0025 % C; 0,02 % oder weniger Si; 0,2 ~ 1,2 % Mn; 0,05 ~ 0,11 % P; 0,01 % oder weniger S; 0,08 ~ 0,12 % löslichem Al; 0,0025 % oder weniger N; 0,003 ~ 0,011 % Nb; 0,01 ~ 0,1 % Mo; 0,0005 ~ 0,0015 % B, gegebenenfalls 0,003 % oder weniger Ti und der Rest aus Fe und anderen unvermeidbaren Verunreinigungen besteht, in Form von Massenanteilen, einem Bake-Hardening-Wert BH von 30 MPa oder mehr, einem Alterungsindex AI von 30 MPa oder weniger, einer Spröd-Duktil-Übergangstemperatur DBTT von -30 °C oder weniger bei einem Ziehverhältnis von 2,0 und einer Zugfestigkeit von 340 ~ 390 MPa.

10. Stahlblech nach Anspruch 9, wobei das Stahlblech größere Mengen an gelöstem Kohlenstoff in der Korngrenze als im Korn aufweist und die Menge an gelöstem Kohlenstoff in dem Korn im Bereich von 3 - 6 ppm liegt.

11. Verfahren zur Herstellung eines einbrennhärtbaren, hochfesten kaltgewalzten Stahlblechs, das eine Korngröße von ASTM-Nr. 9 oder größer, eine Zugfestigkeit von 300 MPa oder mehr aufweist, wobei das Stahlblech NbC-Ausscheidungen mit einer durchschnittlichen Größe von 30 nm oder weniger, 0 nm ausgenommen, und feine AlN-Ausscheidungen mit einer durchschnittlichen Größe von 20 µm oder weniger, 0 µm ausgenommen, aufweist; wobei das Verfahren die folgenden Schritte umfasst:

Warmwalzen einer Stahlbramme mit Fertigwalzen bei oder oberhalb einer Ar3-Übergangstemperatur, um ein warmgewalztes Stahlblech nach dem Erwärmen der Stahlbramme auf eine Temperatur von 1.200 °C oder mehr bereitzustellen, wobei die Stahlbramme aus 0,0016 ~ 0,01 % C; 0,1 % oder weniger Si; 0,2 ~ 1,5 % Mn; 0,05 ~ 0,15 % P; 0,01 % oder weniger S; 0,08 ~ 0,5 % löslichem Al; 0,0025 % oder weniger N; 0,003 ~ 0,1 % Nb; 0,01 ~ 0,4 % Mo; 0,0005 ~ 0,005 % B, gegebenenfalls 0,003 % oder weniger Ti und der Rest aus Fe und anderen unvermeidbaren Verunreinigungen besteht, in Form von Massenanteilen, unter Erfüllung eines Nb/C-Atomverhältnisses von 0,3 ~ 0,7;

Wickeln des warmgewalzten Stahlblechs bei einer Temperatur von 550 °C bis 650 °C;

Kaltwalzen des warmgewalzten Stahlblechs; und kontinuierliches Glühen des kaltgewalzten Stahlblechs bei einer Temperatur von 770 °C bis 830 °C.

12. Verfahren nach Anspruch 11, umfassend die folgenden Schritte:

Warmwalzen einer Stahlbramme mit Fertigwalzen bei einer Temperatur von 900 ~ 950 °C, um ein warmgewalztes Stahlblech nach dem Homogenisieren der Stahlbramme bei einer Temperatur von 1.200 °C oder mehr bereitzustellen, wobei die Stahlbramme aus 0,0016 ~ 0,0025 % C; 0,02% oder weniger Si; 0,2 ~ 1,2 % Mn; 0,05 ~ 0,11 % P; 0,01 % oder weniger S; 0,08 ~ 0,12 % löslichem Al; 0,0025 % oder weniger N; 0,003 ~ 0,011 % Nb; 0,01 ~ 0,1 % Mo; 0,0005 ~ 0,0015 % B;,, gegebenenfalls 0,003 % oder weniger Ti und der Rest aus Fe und anderen unvermeidbaren Verunreinigungen besteht, in Form von Massenanteilen;

Wickeln des warmgewalzten Stahlblechs bei einer Temperatur von 600 ~ 650 °C;

Kaltwalzen des warmgewalzten Stahlblechs mit einer Reduktionsrate von 75 ~ 80 %; und kontinuierliches Glühen des kaltgewalzten Stahlblechs bei einer Temperatur von 770 ~ 830 °C.

13. Verfahren nach Anspruch 11, ferner umfassend die folgenden Schritte:

Schmelztauchen des geglühten Stahlblechs; und

Dressierwalzen des feuerverzinkten Stahlblechs.

14. Verfahren nach Anspruch 13, umfassend die folgenden Schritte:

Warmwalzen einer Stahlbramme mit Fertigwalzen bei einer Temperatur von 900 ~ 950 °C, um ein warmgewalztes Stahlblech nach dem Homogenisieren der Stahlbramme bei einer Temperatur von 1.200 °C oder mehr bereitzustellen, wobei die Stahlbramme aus 0,0016 ~ 0,0025 % C; 0,02 % oder weniger Si; 0,2 ~ 1,2 % Mn; 0,05 ~ 0,11 % P; 0,08 ~ 0,12 % löslichem Al; 0,003 ~ 0,011 % Nb; 0,01 ~ 0,1 % Mo; 0,0005 ~ 0,0015 % B, gegebenenfalls 0,003 % oder weniger Ti und der Rest aus Fe und anderen unvermeidbaren Verunreinigungen besteht, in Form von Massenanteilen;

Warmwalzen einer Stahlbramme mit Fertigwalzen bei einer Temperatur von 900 ~ 950 °C, um ein warmgewalztes Stahlblech nach dem Homogenisieren der Stahlbramme bei einer Temperatur von 1.200 °C oder mehr

bereitzustellen,
 Wickeln des warmgewalzten Stahlblechs bei einer Temperatur von 600 ~ 650 °C;
 Kaltwalzen des warmgewalzten Stahlblechs mit einer Reduktionsrate von 75 ~ 80 %;
 kontinuierliches Glühen des kaltgewalzten Stahlblechs bei einer Temperatur von 770 ~ 830 °C;
 Schmelztauchen des geglühten Stahlblechs; und
 Dressierwalzen des feuerverzinkten Stahlblechs mit einer Reduktionsrate von 1,2 ~ 1,5 %.

Revendications

1. Tôle d'acier laminé à froid à haute résistance durcissable par cuisson fabriquée par laminage à chaud, laminage à froid et recuit continu d'un acier, composée de 0,0016-0,01 % de C ; 0,1 % ou moins de Si ; 0,2-1,5 % de Mn ; 0,05-0,15 % de P ; 0,01 % ou moins de S ; 0,08-0,5 % d'Al soluble ; 0,0025 % ou moins de N ; 0,003-0,1 % de Nb ; 0,01-0,4 % de Mo ; 0,0005-0,005 % de B ; éventuellement 0,003 % ou moins de Ti et pour le reste de Fe et d'autres inévitables impuretés, exprimés en pourcentages pondéraux, tout en satisfaisant un rapport atomique Nb/C de 0,3-0,7, la tôle d'acier ayant des précipités de NbC ayant une taille moyenne de 30 nm ou moins, à l'exclusion de 0 nm, et la tôle d'acier ayant une granulométrie ASTM de 9 ou plus, une résistance à la traction de 300 MPa ou plus, et de fins précipités d'AIN formés lors du laminage à chaud de l'acier, les fins précipités d'AIN ayant une taille moyenne de 20 µm ou moins, à l'exclusion de 0 µm, agissant comme une barrière pour supprimer la croissance des grains lors du recuit de la tôle d'acier.
2. Tôle d'acier selon la revendication 1, dans laquelle la teneur en Al est supérieure à 0,1 % et inférieure ou égale à 0,5 %.
3. Tôle d'acier selon la revendication 1 ou 2, la tôle d'acier comprenant de plus grandes quantités de carbone en solution dans le joint de grain que dans le grain, et la quantité de carbone en solution dans le grain se situant dans la fourchette de 3-6 ppm.
4. Tôle d'acier selon la revendication 1, composée de 0,0016-0,0025 % de C ; 0,02 % ou moins de Si ; 0,2-1,2 % de Mn ; 0,05-0,11 % de P ; 0,01 % ou moins de S ; 0,08-0,12 % d'Al soluble ; 0,0025 % ou moins de N ; 0,003-0,011 % de Nb ; 0,01-0,1 % de Mo ; 0,0005-0,0015 % de B ; éventuellement 0,003 % ou moins de Ti et pour le reste de Fe et d'autres inévitables impuretés, exprimés en pourcentages pondéraux, une valeur de durcissement à la cuisson BH de 30 MPa ou plus, un indice de vieillissement AI de 30 MPa ou moins, une température de transition ductilité-fragilité DBTT de -30 °C ou moins à un rapport d'étrépage de 2,0, et une résistance à la traction de 340-390 MPa.
5. Tôle d'acier selon la revendication 4, la tôle d'acier comprenant de plus grandes quantités de carbone en solution dans le joint de grain que dans le grain, et la quantité de carbone en solution dans le grain se situant dans la fourchette de 3-6 ppm.
6. Tôle d'acier selon la revendication 1, la tôle d'acier étant également traitée dans un procédé de trempage à chaud, la tôle d'acier étant ainsi une tôle d'acier trempé à chaud.
7. Tôle d'acier selon la revendication 6, dans laquelle la teneur en Al est supérieure à 0,1 % et inférieure ou égale à 0,5 %.
8. Tôle d'acier selon la revendication 6 ou 7, la tôle d'acier comprenant de plus grandes quantités de carbone en solution dans le joint de grain que dans le grain, et la quantité de carbone en solution dans le grain se situant dans la fourchette de 3-6 ppm.
9. Tôle d'acier selon la revendication 6, la tôle d'acier trempé à chaud étant composée de 0,0016-0,0025 % de C ; 0,02 % ou moins de Si ; 0,2-1,2 % de Mn ; 0,05-0,11 % de P ; 0,01 % ou moins de S ; 0,08-0,12 % d'Al soluble ; 0,0025 % ou moins de N ; 0,003-0,011 % de Nb ; 0,01-0,1 % de Mo ; 0,0005-0,0015 % de B ; éventuellement 0,003 % ou moins de Ti et pour le reste de Fe et d'autres inévitables impuretés, exprimés en pourcentages pondéraux, et ayant une valeur de durcissement à la cuisson BH de 30 MPa ou plus, un indice de vieillissement AI de 30 MPa ou moins, une température de transition ductilité-fragilité DBTT de - 30 °C ou moins à un rapport d'étrépage de 2,0, et une résistance à la traction de 340-390 MPa.
10. Tôle d'acier selon la revendication 9, la tôle d'acier comprenant de plus grandes quantités de carbone en solution dans le joint de grain que dans le grain, et la quantité de carbone en solution dans le grain se situant dans la fourchette de 3-6 ppm.

11. Procédé de fabrication d'une tôle d'acier laminé à chaud à haute résistance durcissable par cuisson ayant une granulométrie ASTM de 9 ou plus et une résistance à la traction de 300 MPa ou plus, la tôle d'acier ayant des précipités de NbC ayant une taille moyenne de 30 nm ou moins, à l'exclusion de 0 nm, et de fins précipités d'AlN ayant une taille moyenne de 20 μm ou moins, à l'exclusion de 0 μm ;

le procédé comprenant les étapes consistant à :

laminer à chaud une brame d'acier avec un laminage de finition à ou au-dessus d'une température de transformation Ar3 pour obtenir une tôle d'acier laminé à chaud après chauffage de la brame d'acier jusqu'à une température de 1200 °C ou plus, la brame d'acier étant composée de 0,0016-0,01 % de C, 0,1 % ou moins de Si, 0,2-1,5 % de Mn, 0,05-0,15 % de P, 0,01 % ou moins de S, 0,08-0,5 % d'Al soluble, 0,0025 % ou moins de N, 0,003-0,1 % de Nb, 0,01-0,4 % de Mo, 0,0005-0,005 % de B, éventuellement 0,003 % ou moins de Ti et pour le reste de Fe et d'autres inévitables impuretés, exprimés en pourcentages pondéraux, tout en satisfaisant un rapport atomique Nb/C de 0,3-0,7 ;

enrouler la tôle d'acier laminé à chaud à une température de 550 °C à 650 °C ;

laminer à froid la tôle d'acier laminé à chaud ; et

recuire en continu la tôle d'acier laminé à froid à une température de 770 °C à 830 °C.

12. Procédé selon la revendication 11, comprenant les étapes consistant à :

laminer à chaud une brame d'acier avec un laminage de finition à une température de 900-950 °C pour obtenir une tôle d'acier laminé à chaud, après homogénéisation de la brame d'acier à une température de 1200 °C ou plus, la brame d'acier étant composée de 0,0016-0,0025 % de C, 0,02 % ou moins de Si, 0,2-1,2 % de Mn, 0,05-0,11 % de P, 0,01 % ou moins de S, 0,08-0,12 % d'Al soluble, 0,0025 % ou moins de N, 0,003-0,011 % de Nb, 0,01-0,1 % de Mo, 0,0005-0,0015 % de B, éventuellement 0,003 % ou moins de Ti et pour le reste de Fe et d'autres inévitables impuretés, exprimés en pourcentages pondéraux ;

enrouler la tôle d'acier laminé à chaud à une température de 600-650 °C ;

laminer à froid la tôle d'acier laminé à chaud à un taux de réduction de 75-80 % ; et

recuire en continu la tôle d'acier laminé à froid à une température de 770-830 °C.

13. Procédé selon la revendication 11, comprenant en outre les étapes consistant à :

tremper à chaud la tôle d'acier recuit ; et

écrouir la tôle d'acier trempé à chaud.

14. Procédé selon la revendication 13, comprenant les étapes consistant à :

laminer à chaud une brame d'acier avec un laminage de finition à une température de 900-950 °C pour obtenir une tôle d'acier laminé à chaud, après homogénéisation de la brame d'acier à une température de 1200 °C ou plus, la brame d'acier étant composée de 0,0016-0,0025 % de C, 0,02 % ou moins de Si, 0,2-1,2 % de Mn, 0,05-0,11 % de P, 0,08-0,12 % d'Al soluble, 0,003-0,011 % de Nb, 0,01-0,1 % de Mo, 0,0005-0,0015 % de B, éventuellement 0,003 % ou moins de Ti et pour le reste de Fe et d'autres inévitables impuretés, exprimés en pourcentages pondéraux ;

laminer à chaud une brame d'acier avec un laminage de finition à une température de 900-950 °C pour obtenir une tôle d'acier laminé à chaud, après homogénéisation de la brame d'acier à une température de 1200 °C ou plus,

enrouler la tôle d'acier laminé à chaud à une température de 600-650 °C ;

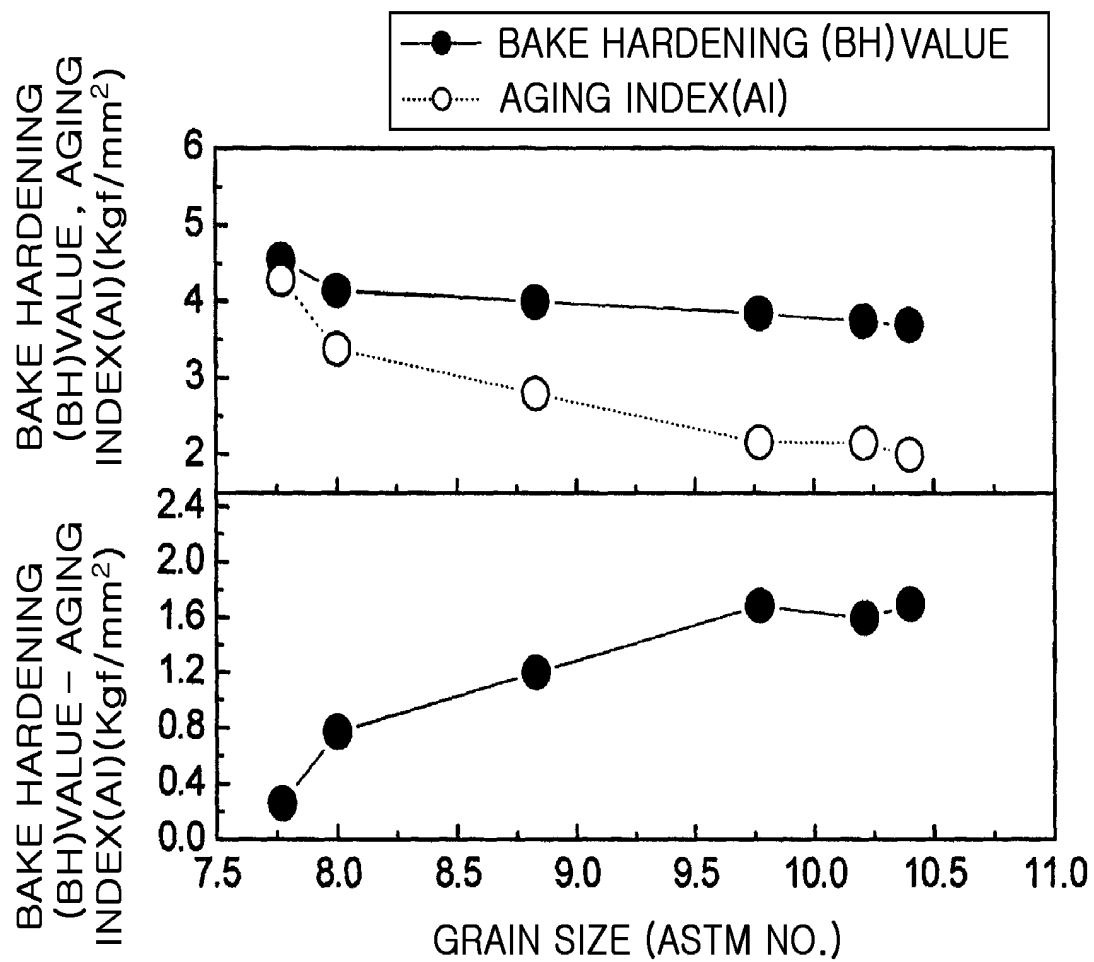
laminer à froid la tôle d'acier laminé à chaud à un taux de réduction de 75-80 % ;

recuire en continu la tôle d'acier laminé à froid à une température de 770-830 °C ;

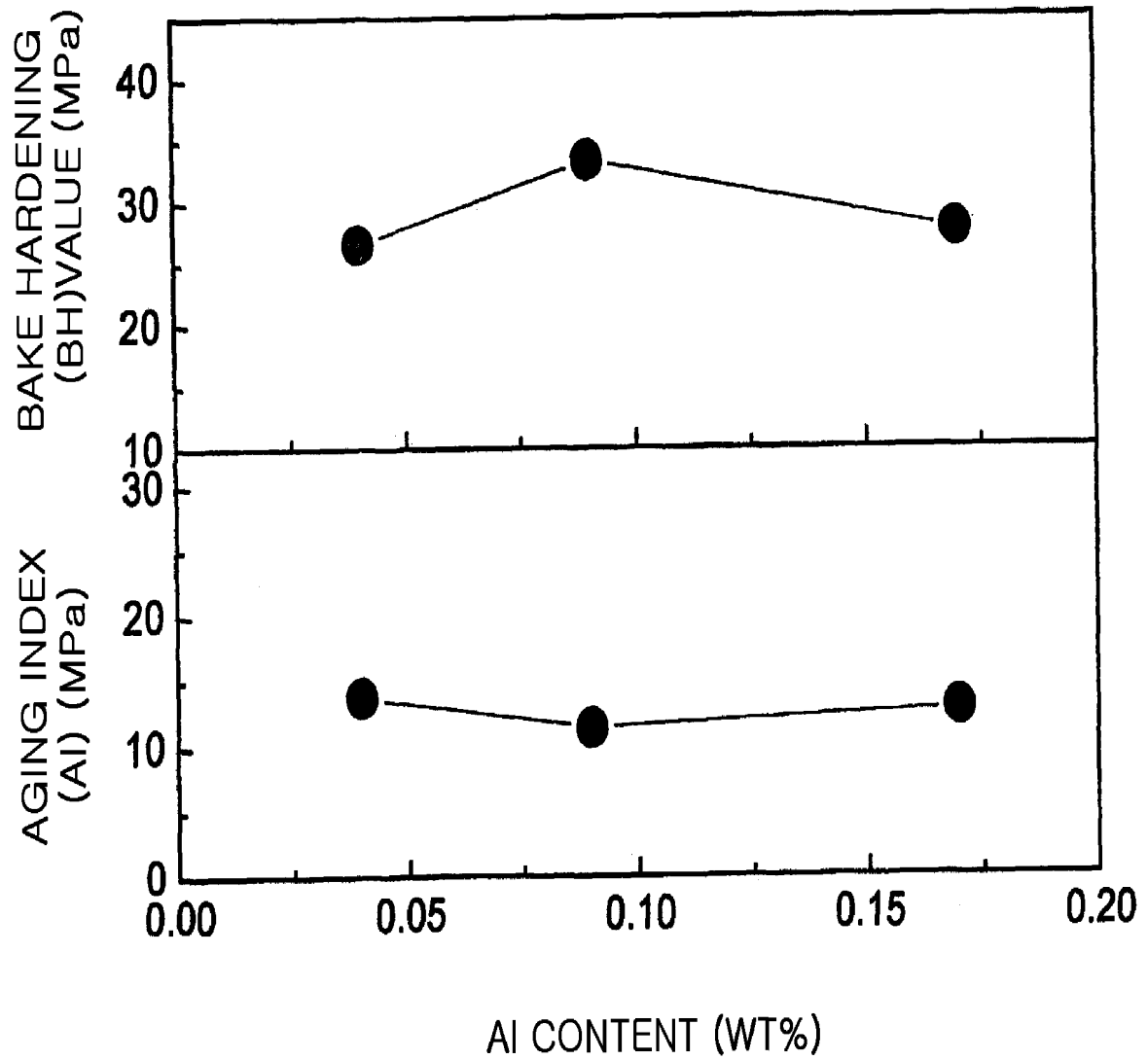
tremper à chaud la tôle d'acier recuit ; et

écrouir la tôle d'acier trempé à chaud à un taux de réduction de 1,2-1,5 %.

[Fig. 1]



[Fig. 2]



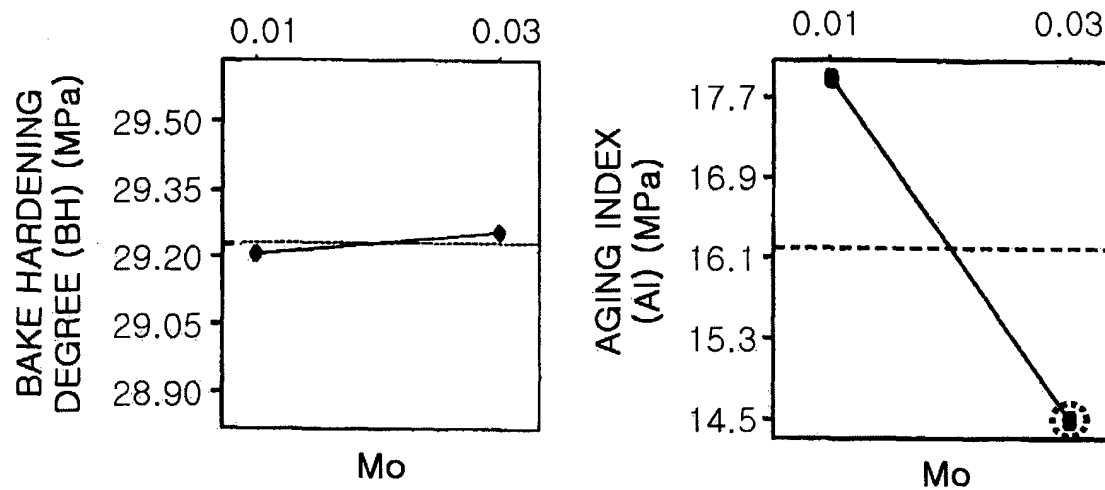


FIG. 3

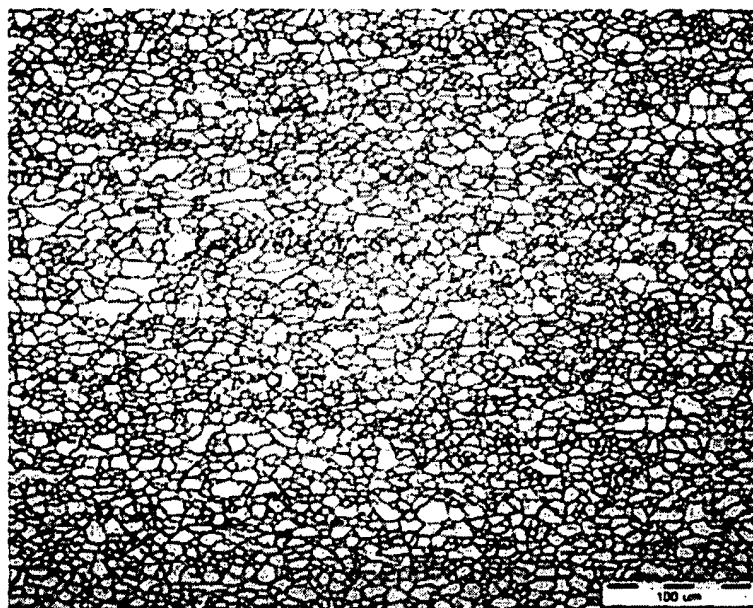
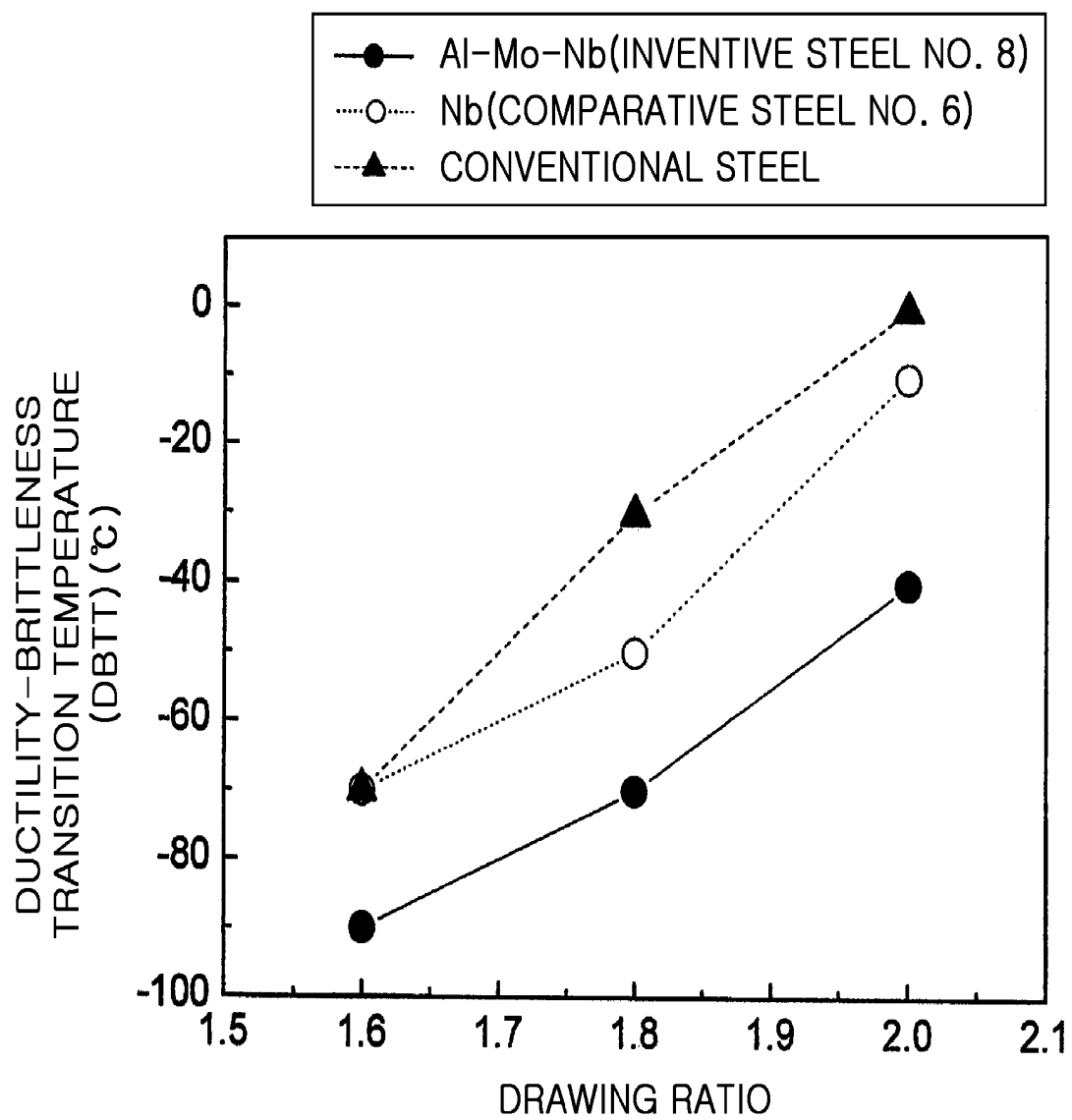
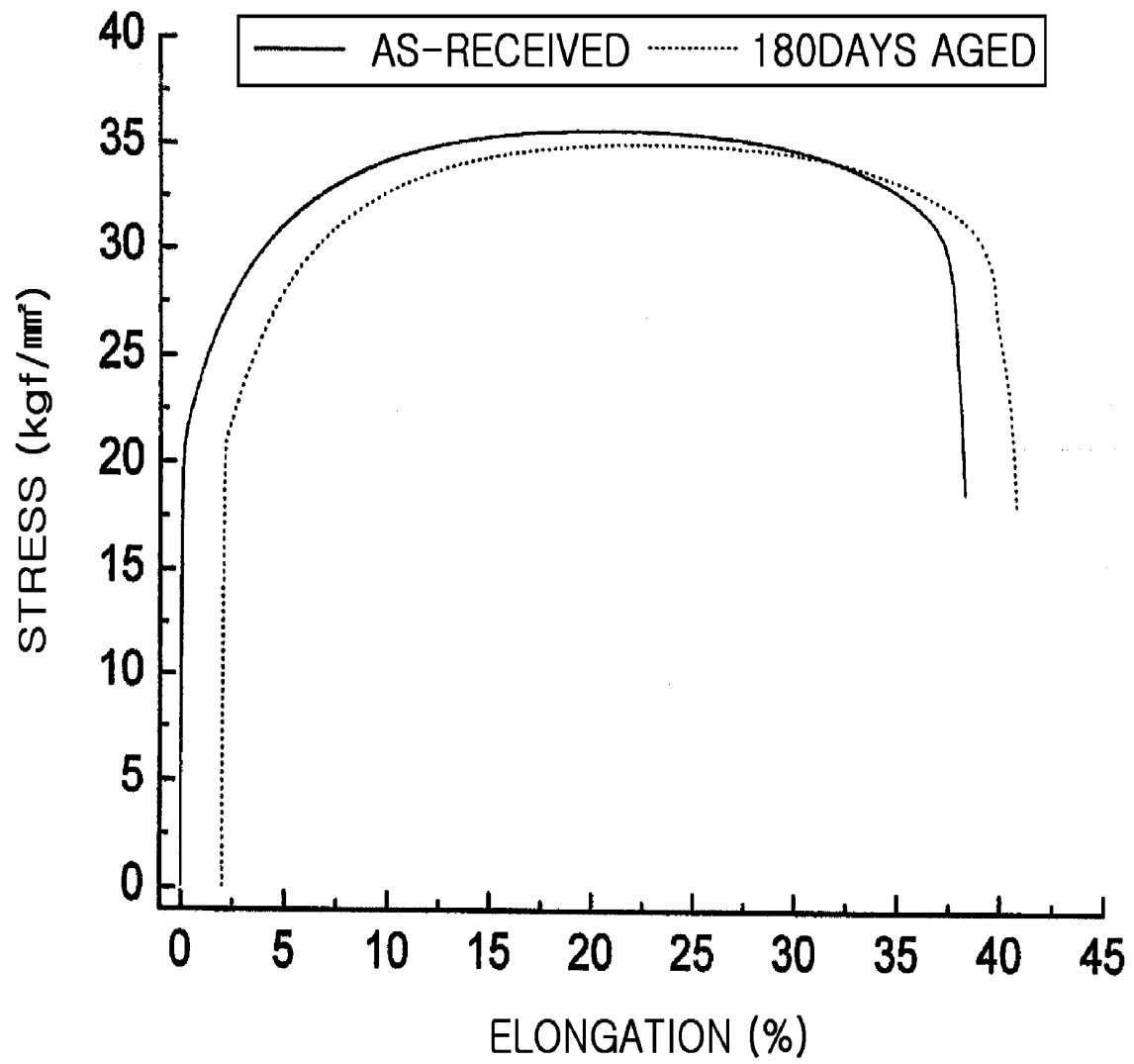


FIG. 4

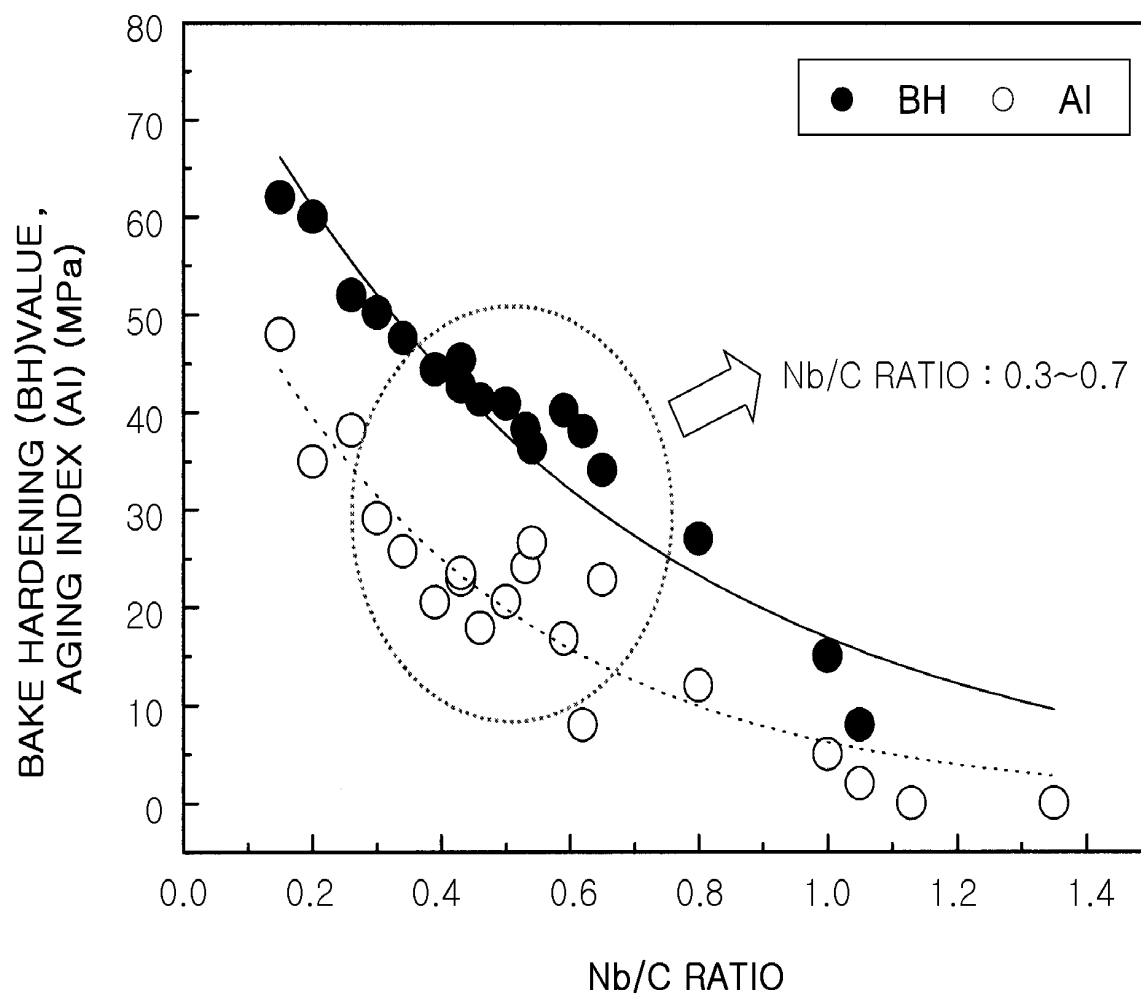
[Fig. 5]



[Fig. 6]



[Fig. 7]



REFERENCES CITED IN THE DESCRIPTION

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