Cold-rolled steel sheet and galvanized steel sheet having excellent strain age hardenability and method of producing the same
Kaltgewalztes Stahlblech und Zinkblech mit Reckalterungseigenschaften und Verfahren zur dessen Herstellung
Tôle d'acier laminée à froid, galvanisée ayant excellent aptitude au durcissement au vieillissement par ecruissage et son procédé de fabrication

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Description

Technical Field

[0001] The present invention relates to a cold-rolled steel sheet, an electro-galvanized steel sheet, a hot-dip galvanized steel sheet, and an alloyed hot-dip galvanized steel sheet, which are suitable as raw material steel sheets for molded products such as building members, mechanical structural parts, automobile structural parts, etc., which are used at positions required to have structural strength, particularly, strength and/or stiffness in deformation, and which are subjected to heat treatment for increasing strength after processing such as pressing or the like, and a method of producing these steel sheets.

1. In the present invention, "excellent strain age hardenability" means that in aging under conditions of holding at a temperature of 170°C for 20 min. after pre-deformation with a tensile strain of 5%, the increment in deformation stress (represented by the amount of BH – yield stress after aging - pre-deformation stress before aging) after aging is 80 MPa or more, and the increment in tensile strength (represented by ATS = tensile strength after aging - tensile strength before pre-deformation) after strain aging (pre-deformation + aging) is 40 MPa or more.

Background Art

[0002] In producing a press-molded product of a thin steel sheet, a process of coating and baking at lower than 200°C is used as a method in which a material having low deformation stress before press forming to facilitate press forming, and then hardened after press forming to increase the strength of a part. As a steel sheet for such coating and baking, a BH steel sheet has been developed.

[0003] For example, Japanese Unexamined Patent Application Publication No. 55-141526 discloses a method in which Nb is added according to the contents of C, N and Al of steel, Nb/(dissolved C + dissolved N) by at% is limited in a specified range, and the cooling rate after annealing is controlled to adjust dissolved C and dissolved N in a steel sheet. Also, Japanese Examined Patent Application Publication No. 61-45689 discloses a method in which baking hardenability is improved by adding Ti and Nb.

[0004] However, in order to improve deep drawability, strength of the raw material sheets of the above-described steel sheets is decreased, and thus the steel sheets are not always sufficient as structural materials.


[0006] In the above-described conventional techniques, strength is increased by bake-hardening due to the functions of small amounts of dissolved C and dissolved N in a steel sheet, and it is well known that a BH(Bake-Hardening) steel sheet is used for increasing only the yield strength of a material, not for increasing tensile strength. Therefore, the conventional techniques have only the effect of increasing the deformation start stress of a part, and the effect of increasing stress (tensile strength after forming) required for deformation over the entire deformation region from the deformation start to the deformation end is not said to be sufficient.

[0007] As a cold-rolled steel sheet having tensile strength increased after forming, for example, Japanese Unexamined Patent Application Publication No. 10-310847 discloses an alloying hot-dip galvanized steel sheet having tensile strength increased by 60 MPa or more by heat treatment in the temperature region of 200 to 450°C.

[0008] This steel sheet contains, by mass%, 0.01 to 0.08% of C, and 0.01 to 3.0% of Mn, and at least one of W, Cr, and Mo in a total of 0.05 to 3.0%, and further contains at least one of 0.005 to 0.1% of Ti, 0.005 to 0.1% of Nb and 0.005 to 0.1% of V according to demand, and the microstructure of the steel is composed of ferrite or mainly composed of ferrite.

[0009] However, this technique comprises forming a fine carbide in the steel sheet by heat treatment after forming to effectively propagate a dislocation of stress applied during pressing, thereby increasing the amount of strain. Therefore, heat treatment must be performed in the temperature range of 220 to 370°C. There is thus the problem of a necessary heat treatment temperature higher than general bake-hardening temperatures.

[0010] Furthermore, it is a very important problem that the body weight of an automobile is decreased in relation to the recent regulation of exhaust gases due to global environmental problems. In order to decrease the body weight of an automobile, it is effective to increase the strength of the used steel sheet, i.e., use a high-tensile-strength steel sheet, thinning the steel sheet used.

[0011] An automobile part using a high-tensile-strength thin steel sheet must exhibit a sufficient property according to its function. The property depends upon the part, and examples of the property include dent resistance, static strength against deformation such as bending, twisting, or the like, fatigue resistance, impact resistance, etc. Namely, the high-tensile-strength steel sheet used for an automobile part is required to be excellent in such a property after forming. The properties are related to the strength of a steel sheet after forming, and thus the lower limit of strength of the high-tensile-strength steel sheet used must be set for achieving thinning.
On the other hand, in the process for forming an automobile part, a steel sheet is press-molded. If the steel sheet has excessively high strength in press forming, the steel sheet causes the following problems: (1) deteriorating shape fixability; (2) deteriorating ductility to cause cracking, necking, or the like during forming; and (3) deteriorating dent resistance (resistance to a dent produced by a local compressive load) when the sheet thickness is decreased. These problems thus inhibit the extension of application of the high-tensile-strength steel sheet to automobile bodies.

As a means for overcoming the problems, a steel sheet composed of ultra-low-carbon steel is known as a raw material, for example, for a cold-rolled steel sheet for an external sheet panel, in which the content of C finally remaining in a solid solution state is controlled to an appropriate range. This type of steel sheet is kept soft during press forming to ensure shape fixability and ductility, and its yield stress is increased by utilizing the strain aging phenomenon which occurs in the step of coating and baking at 170°C for about 20 minutes after press forming, to ensure dent resistance. This steel sheet is soft during press forming because C is dissolved in steel, while dissolved C is fixed to a dislocation introduced in press forming in the coating and baking step after press forming to increase the yield stress.

However, in this type of steel sheet, the increase in yield stress due to strain age hardening is kept down from the viewpoint of prevention of the occurrence of stretcher strain causing a surface defect. This causes the fault that the steel sheet actually less contributes to a reduction in weight of a part.

On the other hand, a steel sheet composed of dissolved N to further increase the amount of bake-hardening, and a steel sheet provided with a composite structure composed of ferrite and martensite to further improve baking hardenability have been proposed for applications in which the appearance is not so important.

For example, Japanese Unexamined Patent Application Publication No. 60-52528 discloses a method of producing a high-strength steel thin sheet having good ductility and spot weldability, in which steel containing 0.02 to 0.15% of C, 0.8 to 3.5% of Mn, 0.02 to 0.15% of P, 0.10% or less of Al, and 0.005 to 0.025% of N is hot-rolled by coiling at a temperature of 550°C or less, cold-rolled, and then annealed by controlled cooling and heat treatment. A steel sheet produced by the technique disclosed in Japanese Unexamined Patent Application Publication No. 60-52528 has a mixed structure comprising a low-temperature transformation product phase mainly composed of ferrite and martensite, and having excellent ductility, and high strength is achieved by utilizing strain aging due to positively added N during coating baking.

Although the technique disclosed in Japanese Unexamined Patent Application Publication No. 60-52528 greatly increases yield stress YS due to strain age hardening, the technique less increases tensile strength TS. Also, this technique causes large variations in the increment in yield stress YS to cause large variations in mechanical properties, and thus it cannot be expected that a steel sheet can be sufficiently thinned for contributing to a reduction in weight of an automobile part, which is currently demanded.

Japanese Examined Patent Application Publication No. 5-24979 discloses a method of producing a high-strength cold-rolled steel thin sheet having baking hardenability which has a composition comprising 0.08 to 0.20% of C, 1.5 to 3.5% of Mn, and the balance composed of Fe and inevitable impurities, and a structure composed of homogeneous bainite containing 5% or less of ferrite, or bainite partially containing martensite. The cold-rolled steel sheet disclosed in Japanese Examined Patent Application Publication No. 5-24979 is produced by quenching in the temperature range of 200 to 400°C in the cooling process after continuous annealing, and then slowly cooling to obtain a structure mainly composed of bainite and having a large amount of bake-hardening which is not obtained by a conventional method.

However, in the steel sheet disclosed in Japanese Examined Patent Application Publication No. 5-24979, yield strength is increased after coating and baking to obtain a large amount of bake-hardening which is not obtained a conventional method, while tensile strength cannot be increased. Therefore, in application to a strength member, improvements in fatigue resistance and impact resistance after forming cannot be expected. Therefore, there is a problem in which the steel sheet cannot be used for applications greatly required to have fatigue resistance and impact resistance, etc.

Also, Japanese Examined Patent Application Publication No. 61-12008 discloses a method of producing a high-tensile-strength steel sheet having a high r value. This method is characterized by annealing ultra-low-C steel used as a raw material in a ferrite-austenite coexistence region after cold rolling. However, the resultant steel sheet has a high r value and a high degree of baking hardenability (BH property), but the obtained BH amount is about 60 MPa at most. Also, the yield point of the steel sheet is increased after strain aging, but TS is not increased, thereby causing the problem of limiting application to parts.

Furthermore, the above-described steel sheet exhibits excellent strength after coating and baking in a simple tensile test, but produces large variations in strength during plastic deformation under actual pressing conditions. Therefore, it cannot be said that the steel sheet is sufficiently applied to parts required to have reliability.

With respect of a hot-rolled steel sheet among coating baked steel sheets for press molded products, for example, Japanese Examined Patent Application Publication No. 8-23048 discloses a method of producing a hot-rolled steel sheet which is soft during processing, and has tensile strength increased by coating and baking after processing to be effective to improve fatigue resistance.

In this technique, steel contains 0.02 to 0.13 mass % of C, and 0.0080 to 0.0250 mass % of N, and the finisher...
deliver temperature and the coiling temperature are controlled to leave a large amount of dissolved N in the steel, thereby forming a composite structure as a metal structure mainly composed of ferrite and martensite. Therefore, an increase of 100 MPa or more in tensile strength is achieved at the heat treatment temperature of 170°C after forming.

Japanese Unexamined Patent Application Publication No. 10-183301 discloses a hot-rolled steel sheet having excellent baking hardenability and natural aging resistance, in which the C and N contents are limited to 0.01 to 0.12 mass % and 0.0001 to 0.01 mass %, respectively, and the average crystal grain diameter is controlled to 8 μm or less to ensure a BH amount of as high as 80 MPa or more, and suppress the Al amount to 45 MPa or less.

However, this steel sheet is a hot-rolled sheet, and is thus difficult to obtain a high r value because the ferrite aggregation texture is made random due to austenite-ferrite transformation. Therefore, the steel, sheet cannot be said to have sufficient deep drawability.

Furthermore, even if the hot-rolled steel sheet obtained by this technique is used as a starting material for cold rolling and recrystallization annealing, the increase in tensile strength obtained after forming and heat treatment is not always equivalent to a hot-rolled steel sheet, and a BH amount of as high as 80 MPa or more cannot be always obtained. This is because the microstructure of the cold-rolled steel becomes different from that of hot-rolled one due to cold rolling and recrystallization annealing, and strain greatly accumulates during cold rolling to easily form a carbide, a nitride or a carbonitride, thereby changing the states of dissolved C and dissolved N.

In consideration of the above-described present conditions, an object of the present invention is to provide a cold-rolled steel sheet and a hot-dip galvanized steel sheet (including an alloyed steel sheet) for deep drawing, which have excellent deep drawability, TS x r value ≥ 750 MPa, and excellent strain aging hardenability (BH ≥ 80 MPa and ΔTS ≥ 40 MPa), and an effective method of producing these steel sheets.

Disclosure of Invention

In order to achieve the objects, the inventors produced various steel sheets having different compositions under various production conditions, and experimentally evaluated various material properties. As a result, it was found that both moldability and hardenability after forming can be improved by using as a strengthening element N, which has not be positively used before in a field requiring high processability, and effectively using the great strain age hardening phenomenon manifested by the action of the strengthening element.

The inventors also found that in order to advantageously use the strain age hardening phenomenon due to N, the strain age hardening phenomenon due to N must be advantageously combined with a condition for coating and baking an automobile, or further positively combined with a heat treatment condition after forming. It was thus found to be effective to appropriately control the hot rolling condition, the cold rolling and the cold rolling annealing condition to control the microstructure of a steel sheet and the amount of dissolved N in certain ranges. It was also found that in order to stably manifest the strain age hardening phenomenon due to N, it is important to control the Al content of the composition according to the N content.

A steel sheet of the present invention exhibits higher strength after coating and baking in a simple tensile test, as compared with a conventional steel sheet, and exhibits small variations in strength in plastic deformation under actual pressing conditions and stable part strength, thereby enabling application to parts required to have reliability. For example, a portion where large strain is applied to decrease the thickness has higher hardenability than other portions, and is considered homogeneous when being evaluated based on a surcharge load ability of (thickness) x (strength), thereby stabilizing strength as a part.

As a result of further intensive research for achieving the objects, the inventors found the following:

1) In order to increase tensile strength after forming and heat treatment, a new dislocation must be introduced for progressing tensile deformation. The movement of the dislocation introduced by pre-deformation must be prevented by interaction between the dislocation introduced by forming and an interstitial element or a precipitate even when upper yield stress is attained.
2) In order to obtain the above interaction by forming a carbide, a nitride or a carbonitride of W, Cr, Mo, Ti, Nb, Al or the like, the heat treatment temperature after forming must be increased to 200°C or more. Therefore, it is more advantageous to positively use the interstitial element or a Fe carbide or Fe nitride because the heat treatment temperature after forming is decreased.
3) Of interstitial elements, dissolved N has the higher interaction with a dislocation introduced by forming than dissolved C even when the heat treatment temperature after forming is decreased, and thus a dislocation introduced by pre-deformation less moves when upper yield stress is attained.
4) Although dissolved N is present in crystal grains and crystal grain boundaries in steel, the increase in strength after forming and heat treatment increases as the area of the crystal grain boundaries increases. Namely, the smaller crystal grain diameter is advantageous.
5) In order to increase the crystal grain boundary area, it is advantageous to add a combination of Nb and B and...
The temperature of the final pass was 910°C higher than the Ar$_3$ transformation point, and then cooled with water for 0.1 second. Then, the sheet bar was subjected to heat treatment corresponding to coiling at 500°C for 1 hour.

Experiment 1

A sheet bar (thickness: 30 mm) having a composition containing, by mass %, 0.0010% of C, 0.0150% of B, 0.015 of Si, 0.5% of Mn, 0.03% of P, 0.08% of S and 0.011% of N, 0.005 to 0.05% of Nb and 0.005 to 0.03% of Al, and the balance composed of Fe and inevitable impurities was uniformly heated at 1150°C, hot-rolled by three passes so that the temperature of the final pass was 900°C higher than the Ar$_3$ transformation point, and then cooled with water for 0.1 second. Then, the sheet bar was subjected to heat treatment corresponding to coiling at 500°C for 1 hour.

The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealed at 800°C for 40 seconds, and then temper-rolled with a rolling reduction ratio of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and tensile strength was measured with a strain rate of 0.02/s by using a general tensile testing machine. Also, tensile strain of 10% was applied to a tensile test specimen of JIS No. 5 separately obtained from the cold-rolled sheet in the rolling direction, and then the specimen was subjected to a normal tensile test after heat treatment at 120°C for 20 minutes. The difference between the tensile strength of the specimen obtained from the cold-rolled sheet and the tensile strength of the specimen heat treated at 120°C for 20 minutes after application of 10% tensile strain was considered as the increase in tensile strength after forming ($\Delta$TS).

The figure indicates that $\Delta$TS becomes 60 MPa or more when the value of (N% - 14/93•Nb% - 14/27•Al%) - 14/11•B%) satisfies 0.0015 mass %.

Experiment 2

A sheet bar (thickness: 30 mm) having a composition containing, by mass %, 0.0010% of C, 0.02 of Si, 0.6% of Mn, 0.01% of P, 0.009% of S and 0.012% of N, 0.01% of Al, 0.015% of Nb, 0.00005 to 0.0025% of B, and the balance composed of Fe and inevitable impurities was uniformly heated at 1100°C, hot-rolled by three passes so that the temperature of the final pass was 920°C higher than the Ar$_3$ transformation point, and then cooled with water for 0.1 second. Then, the sheet bar was subjected to heat treatment corresponding to coiling at 450°C for 1 hour.

The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealed at 820°C for 40 seconds, and then temper-rolled with a rolling reduction ratio of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and tensile strength was measured with a strain rate of 0.02/s by using a general tensile testing machine. Also, tensile strain of 10% was applied to a tensile test specimen of JIS No. 5 separately obtained from the cold-rolled sheet in the rolling direction, and then the specimen was subjected to a normal tensile test after heat treatment at 120°C for 20 minutes.

As a result of observation of the microstructure, it was also found that by adding a combination of Nb and B to make fine crystal grains, a high $\Delta$TS can be obtained.

As a result of observation of the microstructure, it was also found that by adding a combination of Nb and B to make fine crystal grains, a high $\Delta$TS can be obtained.

Namely, with a B content of less than 0.0005 mass %, the effect of making fine crystal grains by adding a combination with Nb is small. On the other hand, with a B content of over 0.0015 mass %, the amount of B segregated in the grain boundaries and the vicinities thereof is increased to decrease the amount of effective dissolved N because of the strong interaction between B atoms and N atoms, thereby possibly decreasing $\Delta$TS.

Experiment 3

A sheet bar (thickness: 30 mm) of each of steel A having a composition containing, by mass %, 0.0010% of C, 0.012% of B, 0.0010% of Si, 0.011% of Mn, 0.03% of P, 0.008% of S, 0.014% of Nb, 0.01% of Al, and the balance composed of Fe and inevitable impurities, and steel B having a composition containing, by mass %, 0.010% of C, 0.0012% of N, 0.0010% of B, 0.01% of Si, 0.5% of Mn, 0.03% of P, 0.008% of S, 0.014% of Nb, 0.01% of Al, and the balance composed of Fe and inevitable impurities was uniformly heated at 1150°C, hot-rolled by three passes so that the temperature of the final pass was 910°C higher than the Ar$_3$ transformation point, and then cooled with a gas for 0.1 second. Then, each of the sheet bars was subjected to heat treatment corresponding to coiling at 600°C for 1 hour.
[0043] Each of the thus-obtained hot-rolled sheets having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealed at 880°C for 40 seconds, and then temper-rolled with a rolling reduction ratio of 0.8%.

[0044] Then, a tensile test specimen of JIS No. 5 was obtained from each of the resultant cold-rolled sheets in the rolling direction, and tensile strength was measured with a strain rate of 0.02/s by using a general tensile testing machine. Also, tensile strain of 10% was applied to a tensile test specimen of JIS No. 5 separately obtained from each of the cold-rolled sheets in the rolling direction, and then the specimen was subjected to a normal tensile test after heat treatment at various temperatures for 20 minutes.

[0045] Fig. 3 shows the results of measurement of the influence of the heat treatment temperature after forming on ΔTS. This figure indicates that in the relatively low temperature region of heat treatment temperatures of 200°C or less after forming, the ultra-low carbon steel A having a high N content exhibits higher ΔTS than the semi-ultra low carbon steel B having a low N content, and while in the high temperature region, both steel materials exhibit substantially the same ΔTS. There experimental results reveal that in order to ensure ΔTS in the low temperature region, it is effective to use dissolved N.

[0046] A sheet bar of steel containing 0.0015% of C, 0.30 of Si, 0.8% of Mn, 0.03% of P, 0.005% of S and 0.012% of N, and 0.02 to 0.08% of Al was uniformly heated at 1050°C, hot-rolled by seven passes so that the temperature of the final pass was 670°C, and then recrystallized and annealed at 700°C for 5 hours. The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealed at 875°C for 40 seconds, and then temper-rolled with a rolling reduction of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and TS x r value and ΔTS were measured with a strain rate of 3 x 10⁻³/s by using a general tensile testing machine. The results are shown in Fig. 5. In this experiment, steel containing N/Al ≥ 0.3 is satisfied, TS x r value ≥ 750 and ΔTS ≥ 40 MPa are achieved. It was also confirmed that when N/Al ≥ 0.3, BH ≥ 80 MPa is attained.

Experiment 4

[0048] A sheet bar of steel containing 0.0015% of C, 0.30 of Si, 0.8% of Mn, 0.03% of P, 0.008% of S and 0.011% of N, 0.005 to 0.05% of Nb, and 0.005 to 0.03% of Al was uniformly heated at 1000°C, hot-rolled by seven passes so that the temperature of the final pass was 670°C, and then recrystallized and annealed at 700°C for 5 hours. The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealed at 875°C for 40 seconds, and then temper-rolled with a rolling reduction of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and TS x r value and ΔTS were measured with a strain rate of 3 x 10⁻³/s by using a general tensile testing machine. The results are shown in Fig. 5. In this experiment, steel containing N/Al ≥ 0.3 is satisfied, TS x r value ≥ 750 and ΔTS ≥ 40 MPa are achieved. It was also confirmed that when N/Al ≥ 0.3, BH ≥ 80 MPa is attained.

Experiment 5

[0049] A sheet bar of steel containing 0.0015% of C, 0.011% of P, 0.30 of Si, 0.6% of Mn, 0.03% of P, 0.008% of S and 0.011% of N, 0.005 to 0.05% of Nb, and 0.005 to 0.03% of Al was uniformly heated at 1000°C, hot-rolled by seven passes so that the temperature of the final pass was 670°C, and then recrystallized and annealed at 800°C for 60 seconds. The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealing at 875°C for 40 seconds, and then temper-rolled with a rolling reduction of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and TS x r value, BH and ΔTS were measured with a strain rate of 3 x 10⁻³/s by using a general tensile testing machine. The relations between the measured values and N/(Al+Nb+B) are shown in Fig. 5. In this experiment, steel containing 0.005 to 0.05% of Nb and 0.0010% of B was used. This figure indicates that in the range of N/(Al+Nb+B) ≥ 0.30, BH ≥ 80 MPa, ΔTS ≥ 60 MPa, and TS x r value ≥ 850 are achieved.

Experiment 6

[0050] A sheet bar of steel containing 0.0010% of C, 0.02 of Si, 0.6% of Mn, 0.01% of P, 0.009% of S and 0.015% of N, 0.01% of Al, 0.015% of Nb and 0.0001 to 0.0025% of B was uniformly heated at 1050°C, hot-rolled by seven passes so that the temperature of the final pass was 680°C, and then recrystallized and annealed at 850°C for 5 hours. The thus-obtained hot-rolled sheet having a thickness of 4 mm was cold-rolled with a rolling reduction ratio of 82.5%, recrystallized and annealed at 875°C for 40 seconds, and then temper-rolled with a rolling reduction of 0.8%. Then, a tensile test specimen of JIS No. 5 was obtained from the resultant cold-rolled sheet in the rolling direction, and TS x r value, BH and ΔTS were measured with a strain rate of 3 x 10⁻³/s by using a general tensile testing machine. The relations between the measured values and the B content are shown in Fig. 6.
This figure indicates that in the B content range of 0.0003 to 0.0015%, BH ≥ 80 MPa, ΔTS ≥ 60 MPa, which is higher than the case of B < 0.0003%, and TS x r value ≥ 850 are achieved. As a result of observation of the microstructure, it was also confirmed that in this B range, crystal grains are significantly made fine.

The results shown in Figs. 5 and 6 indicate that in the range of N/(A1+Nb+B) ≥ 0.30 wherein B ≥ 0.0003%, the crystal grains are further made fine by combining Nb, and ΔTS and the level of TS x r value are further improved. When B < 0.0003%, the effect of making fine crystal grains by combining Nb is not exhibited. On the other hand, when B > 0.0015%, properties further deteriorate. This is possibly due to the fact that the amount of B segregated in the grain boundaries and the vicinities thereof is increased to decrease the amount of effective dissolved N due to the strong interaction between B and N atoms. The same research as described above was carried out for the case in which Ti and V were added in place of Nb, and it was confirmed that the same effect as Nb could be obtained. The present invention has been achieved based on the above-described findings, and the gist of the invention was follows.

In a first aspect of the present invention, a cold-rolled steel sheet having excellent strain age hardenability comprises the compositions in the following ranges. Namely, a cold-rolled steel sheet having excellent strain age hardenability comprises a composition, by mass %, comprising:

- C: less than 0.01%
- Si: 0.005 to 1.0%
- Mn: 0.01 to 1.5%
- P: 0.1% or less
- S: 0.01% or less
- Al: 0.005 to 0.030%; and
- N: 0.005 to 0.040%;

wherein N/Al is 0.30 or more, the amount of dissolved N is 0.0010% or more, and the balance is composed of Fe and inevitable impurities.

In the first aspect of the present invention, the composition, by mass %, further comprises:

- B: 0.0001 to 0.0030%; and
- Nb: 0.005 to 0.050%;

wherein the ranges of B and Nb satisfy the following equations (1) and (2):

\[ N\% \geq 0.0015 + 14/93 \cdot Nb\% + 14/27 \cdot Al\% + 14/11 \cdot B\% \ldots (1) \]

\[ C\% \leq 0.5 \cdot (12/93) \cdot Nb\% \ldots (2) \]

In the first aspect of the present invention, the above composition, by mass %, optionally further comprises at least one of Cu, Ni and Mo in a total of 1.0% or less according to demand.

In the first aspect of the present invention, the steel sheet preferably has a crystal grain diameter of 20 μm or less.

In the first aspect of the present invention, strength after forming is preferably increased by 60 MPa or more by heat treatment in the low temperature region of 120 to 200°C.

In the first aspect of the present invention, the surface of the cold-rolled steel sheet may be coated by electro-galvanization, hot-dip galvanization, or alloying hot-dip galvanization.

In a second aspect of the present invention, a method of producing a cold-rolled steel sheet having excellent strain age hardenability comprises hot-rolling a steel slab under conditions in which the slab is cooled immediately after the end of finish rolling and coiled at a coiling temperature of 400 to 800°C, cold-rolling the hot-rolled sheet with a rolling reduction ratio of 60 to 95%, and then performing recrystallization annealing at a temperature of 650 to 900°C, wherein the steel slab has a composition, by mass %, comprising:

- C: less than 0.01%
- Si: 0.005 to 1.0%
- Mn: 0.01 to 1.5%
- P: 0.1% or less
- S: 0.01% or less
Al: 0.005 to 0.030%; and
N: 0.005 to 0.040%;
wherein N/Al is 0.30 or more, and the balance is substantially composed of Fe.

In the second aspect of the present invention, the composition, by mass %, further comprises:

B: 0.0001 to 0.0030%; and
Nb: 0.005 to 0.050%;
wherein the ranges of B and Nb satisfy the following equations (1) and (2):

\[
N\% \geq 0.0015 + 14/93 \cdot Nb\% + 14/27 \cdot Al\% + 14/11 \cdot B\% \ldots \ (1)
\]

\[
C\% \leq 0.5 \cdot (12/93) \cdot Nb\% \ldots \ (2)
\]

In the second aspect of the present invention, in the heating-up step in recrystallization annealing, the temperature is preferably increased at a rate of 1 to 20°C/s in the temperature region from 500°C to the recrystallization temperature.

In the second aspect of the present invention, hot-dip galvanization and heat alloying may be performed after the recrystallization annealing.

Brief Description of the Drawings

Fig. 1 shows the relation between steel compositions (N% - 14/93•Nb% - 14/27•Al% - 14/11•B%) and the increase in tensile strength (ΔTS) after forming.

Fig. 2 shows the relation between the B content and ΔTS of steel containing a combination Nb and B.

Fig. 3 shows comparison of the difference in increase in tensile strength by heat treatment after forming in a low temperature region between steel B (conventional steel) containing a large amount of dissolved C and steel A (steel of this invention) containing a large amount of dissolved N.

Fig. 4 shows the influence of the crystal grain diameter d and steel compositions (N% - 14/93•Nb% - 14/27•Al% - 14/11•B%) on the decrease in elongation (ΔEI) due to natural aging and the increase in tensile strength (ΔTS) after forming.

Fig. 5 shows the relations between Ts x r value, BH, ΔTS and N/(Al+Nb+B).

Fig. 6 shows the relations between Ts x r value, BH, ΔTS and the B amount.

Best Mode for Carrying Out the Invention

Description will now be made of the reasons for limiting compositions to the ranges below in accordance with a first embodiment of the present invention.

C: less than 0.01 mass %

From the viewpoint of excellent deep drawability and press moldability, C is advantageously as small as possible. Also, redissolution of NbC proceeds in the annealing step after cold rolling to increase the amount of dissolved C in crystal grains, thereby easily causing deterioration in natural aging resistance. Therefore, the C amount is preferably suppressed to less than 0.01 mass %, more preferably 0.0050 mass % or less, and most preferably 0.0030 mass % or less.

Si: 0.005 to 1.0 mass %

Si is a useful composition for suppressing a decrease in elongation, and improving strength. However, with a Si content of less than 0.005 mass %, the effect of addition of Si is insufficient, while with a Si content of over 1.0 mass %, surface properties deteriorate to deteriorate ductility. Therefore, the Si content is limited to the range of 0.005 to 1.0
mass %, and preferably the range of 0.01 to 0.75 mass %.

Mn: 0.01 to 1.5 mass %

[0067] Mn not only is useful as a strengthening composition for steel, but also has the function to suppress embrittlement with S due to the formation of MnS. However, with a Mn content of less than 0.01 mass %, the effect of addition of Mn is insufficient, while with a Mn content of over 1.5 mass %, surface properties deteriorate to deteriorate ductility. Therefore, the Mn content is limited to the range of 0.01 to 1.5 mass %, and preferably the range of 1.10 to 0.75 mass %.

P: 0.01 mass % or less

[0068] P is a solid solution strengthening element which effectively contributes to reinforcement of steel. However, with a P content of over 0.01 mass %, deep drawability deteriorates due to the formation of phosphide such as (FeNb),P or the like. Therefore, P is limited to 0.10 mass % or less.

S: 0.01 mass % or less

[0069] With a high S content, the amount of inclusions is increased to deteriorate ductility. Therefore, contamination with S is preferably prevented as much as possible, but an S content up to 0.01 mass % is allowable.

Al: 0.005 to 0.030 mass %

[0070] Al is added as a deoxidizer for improving the yield of carbonitride forming components. However; with an Al content of less than 0.005 mass %, the effect is insufficient, while with an Al content of over 0.030 mass %, the amount of N to be added to steel is increased to easily cause slab defects during steel making. Therefore, Al is contained in the range of 0.005 to 0.030 mass %.

N: 0.005 to 0.040 mass %

[0071] In the present invention, N is an important element which plays the role of imparting strain age hardenability to a steel sheet. However, with an N content of less than 0.005 mass %, a sufficient strain age hardenability cannot be obtained, while with an N content of as high as over 0.040 mass %, press moldability deteriorates. Therefore, N is contained in the range of 0.005 to 0.040 mass %, and preferably in the range of 0.008 to 0.015 mass %.

B: 0.0001 to 0.003 mass %

[0072] B is added in a combination with Nb to exhibit the function to effectively make fine the hot-rolled structure and the cold-rolled recrystallized annealed structure, and to improve cold-work embrittlement resistance. However, with a B content of less than 0.0001 mass %, the sufficient effect of making fine the structures cannot be obtained, while with a B content of over 0.003 mass %, the amount of BN precipitate is increased, and dissolution in the slab heating step is hindered. Therefore, B is contained in the range of 0.0001 to 0.003 mass %, preferably in the range of 0.0001 to 0.0015 mass %, and more preferably in the range of 0.0007 to 0.0012 mass %.

Nb: 0.005 to 0.050 mass %

[0073] Nb is added in a combination with B to contribute to refinement of the hot-rolled structure and the cold-rolled recrystallized annealed structure, and have the function to fix dissolved C as NbC. Furthermore, Nb forms a nitride NbN to contribute to refinement of the cold-rolled recrystallized annealed structure. However, with a Nb content of less than 0.005 mass %, not only it becomes difficult to precipitate and fix dissolved C, but also the hot-rolled structure and the cold-rolled, recrystallized, annealed structure are not sufficiently made fine, while with a Nb content of over 0.050 mass %, ductility deteriorates. Therefore, Nb is contained in the range of 0.005 to 0.050 mass %, and preferably 0.010 to 0.030 mass %.

[0074] As described above, Nb has the function to fix dissolved C as NbC, and forms a nitride NbN. Similarly, Al and B form AlN and BN, respectively. Therefore, in order to ensure the sufficient amount of dissolved N and sufficiently decrease the amount of dissolved C, it is important to satisfy the following relations (1) and (2):
In the present invention, in order to obtain a high strain aging property and prevent aging deterioration, the crystal grain diameter is preferably decreased.

Namely, as described above with reference to Fig. 4, even when \( (N\% - 14/93\cdot Nb\% - 14/27\cdot Al\% - 14/11\cdot B\%) \geq 0.0015 \text{ mass \%}, \text{i.e., when a relatively large amount of dissolved N is contained}, \Delta E_l \text{ can be suppressed to 2.0\% or less by decreasing the crystal grain diameter d to 20 \mu m or less. The crystal grain diameter d is more preferably decreased to 15 \mu m or less. This is because, as shown in Fig. 4, } \Delta E_l \text{ can be suppressed to 2.0\% or less by decreasing the crystal grain diameter d to 15 \mu m or less.}

The production conditions according to a second embodiment of the present invention will be described. Steel having the above-described suitable composition is melted by a known melting method such as a converter or the like, and a steel slab is formed by an ingot making method or a continuous casting method.

Then, the steel slab is heated and soaked, and then hot-rolled to form a hot-rolled sheet. In the present invention, the heating temperature of hot rolling is not specified, but the heating temperature of hot rolling is preferably set to 1300°C or less. This is because it is advantageous to fix and precipitate dissolved C as a carbide in order to improve deep drawability. In order to further improve processability, the heating temperature is preferably set to 1150°C or less. However, with a heating temperature of less than 900°C, improvement in processability is saturated to conversely increase the rolling load in hot rolling, thereby increasing the danger of causing a rolling trouble. Therefore, the lower limit of the heating temperature is preferably 900°C.

The total rolling reduction ratio of hot rolling is preferably 70\% or more. This is because with a toal rolling reduction ratio of less than 70\%, the crystal grains of the hot-rolled sheet are not sufficiently made fine.

During hot rolling, finish rolling is preferably finished in the temperature region of 650 to 960°C, and the finishing temperature of hot-rolling may be in the \( \gamma \) region above the \( Ar_3 \) transformation point, or the \( \alpha \) region below the \( Ar_3 \) transformation point. With the finishing temperature in hot-rolling process over 960°C, the crystal grains of the hot-rolled sheet are coarsened to deteriorate deep drawability after cold rolling and annealing. On the other hand, with a temperature of less than 650°C, deformation resistance is increased to increase the hot-rolling load, causing difficulties in rolling.

Preferably, cooling is started immediately after the end of final rolling in hot-rolling process to prevent normal grain growth and suppress AlN precipitation in the cooling step.

Although the cooling condition is not limited, the starting time of the cooling step is preferably within 1.5 seconds, more preferably 1.0 second, and most preferably 0.5 second, after the end of finish rolling. This is because when cooling is performed immediately after the end of rolling, a large amount of ferrite nuclei is produced due to an increase in the degree of over cooling with accumulated strain to promote ferrite transformation and suppress the diffusion of dissolved N in the \( \gamma \) phase into the ferrite grains, thereby increasing the amount of dissolved N present in the ferrite grain boundaries.

The cooling rate is preferably 10°C/s or more in order to ensure dissolved N. Particularly, when the finishing temperature of hot-rolling is the \( Ar_3 \) transformation point or more, the cooling rate is preferably 50°C/s or more in order to ensure dissolved N.

Then, the hot-rolled sheet is coiled. In order to coarsen a carbide, the coiling temperature is advantageously as high as possible. However, with a coiling temperature of over 800°C, the scale formed on the surface of the hot-rolled sheet is thickened to increase the load of the work of removing the scale, and progress the formation of a nitride, causing a change in the amount of dissolved N in the coil length direction. On the other hand, with a coiling temperature of less than 400°C, the coiling work becomes difficult. Therefore, the coiling temperature of the hot-rolled sheet must be in the range of 400 to 800°C.

Then, the hot-rolled sheet is cold-rolled, but the rolling reduction ratio of cold rolling must be 60 to 95\%. This is because with a rolling reduction ratio of cold rolling of less than 60\%, a high \( r \) value cannot be expected, while with a rolling reduction ratio of over 95\%, the \( r \) value is decreased.

The cold-rolled sheet subjected to cold rolling is then recrystallized and annealed. Although the annealing method may be either continuous annealing or batch annealing, continuous annealing is advantageous. The continuous annealing may be performed either in a normal continuous annealing line or in a continuous hot-dip galvanization line.

The preferable annealing conditions include 650°C or more for 5 seconds or more. This is because with an annealing temperature of less than 650°C, and an annealing condition of less than 5 seconds, recrystallization is not completed to decrease deep drawability. In order to improve deep drawability, annealing is preferably performed in the ferrite single phase region at 800°C or more for 5 seconds or more.
Annealing in the high-temperature \( \alpha + \gamma \) two-phase region partially produces \( \alpha \rightarrow \gamma \) transformation to improve the r value due to the development of the \{111\} aggregation structure. However, when \( \alpha \rightarrow \gamma \) transformation completely proceeds, the aggregation structure is made random to decrease the r value, thereby deteriorating deep drawability.

The upper limit of the annealing temperature is preferably 900°C. This is because with an annealing temperature of over 900°C, redissolution of a carbide proceeds to excessively increase the amount of dissolved C, thereby deteriorating the natural aging property. When \( \alpha \rightarrow \gamma \) transformation occurs, the aggregation structure is made random to decrease the r value, deteriorating deep drawability.

Furthermore, in the heating-up step in recrystallization annealing, slow heating is performed in the temperature region from 500°C to the recrystallization temperature to sufficiently precipitate AlN, and the like, thereby effectively decreasing the crystal grain diameter of the steel sheet.

The temperature region in which controlled heating must be performed is 500°C, at which AlN or the like starts to precipitate, to the recrystallization temperature. The heating rate is preferably in the range of 1 to 20°C/s because with a heating rate of over 20°C/s, the sufficient amount of precipitates cannot be obtained, while with a heating rate of less than 1°C/s, precipitates are coarsened to weaken the effect of suppressing grain growth.

After the recrystallization annealing, temper rolling of 10% or less may be performed for correcting the shape and controlling surface roughness.

The cooling rate after soaking in recrystallization annealing is preferably 10 to 50°C/s. This is because with a cooling rate of 10°C/s or less, grains are grown during cooling to coarsen the crystal grains, thereby deteriorating the strain aging property and natural aging property. While with a cooling rate of 50°C/s or more, dissolved N does not sufficiently diffuse into the grain boundaries, deteriorating the natural aging property. The cooling rate is preferably 10 to 30°C/s.

After the recrystallization annealing, hot-dip galvanization and alloying by heating are performed to form a galvannealed steel sheet as occasion demands.

With a steel sheet subjected to surface treatment generally used for steel thin sheets, such as a steel sheet (a dull-finished steel sheet, a bright-finished steel sheet, or a steel sheet having a specified roughness pattern formed on the surface thereof), which is produced by temper-rolling the alloyed hot-dip galvanized steel sheet, for improving processability and the appearance after processing, a steel sheet having an oil film layer of antirust oil or lubricating oil formed on the surface thereof, or the like, the effect of the present invention can be sufficiently exhibited in the composition range of the prevent invention.

Therefore, a cold-rolled steel sheet and a galvannealed steel sheet can be obtained, which have excellent deep drawability and excellent strain aging hardenability, that tensile strength increased by press forming and heat treatment.

As described above, in the present invention, "excellent strain aging hardenability" means that in aging under conditions of holding at a temperature of 170°C for 20 min. after pre-deformation with a tensile strain of 5%, the increment in deformation stress (represented by the amount of BH = yield stress after aging - pre-deformation stress before aging) after aging is 80 MPa or more, and the increment in tensile strength (represented by \( \Delta TS = \) tensile strength after aging - tensile strength without strain aging) after strain aging (pre-deformation + aging) is 40 MPa or more.

In defining the strain aging hardenability, the amount of pre-strain (pre-deformation) is an important factor. As a result of research of the influence of the amount of pre-strain on strain aging hardenability, the inventors found that (1) the deformation stress in the above-described deformation system can be referred to as an amount of approximately uniaxial strain (tensile strain) except the case of excessive deep drawing, (2) the amount of uniaxial strain of an actual part exceeds 5%, and (3) the strength of a part sufficiently corresponds to the strength (YS and TS) obtained after strain aging with a pre-strain of 5%. In the present invention, based on these findings, the pre-deformation of strain aging is defined to a tensile strain of 5%.

Conventional coating and baking conditions include 170°C and 20 min as standards. When a strain of 5% is applied to the steel sheet of the present invention, which contains a large amount of dissolved N, hardening can be achieved even by aging at low temperature. In other words, the range of aging conditions can be widened. In order to attain a sufficient amount of hardening, generally, retention at a higher temperature for a longer time is advantageous as long as softening does not occurs by over aging.

Specifically, in the steel sheet of the present invention, the lower limit of the heating temperature at which hardening significantly takes place after pre-deformation is about 100°C. On the other hand, with the heating temperature of over 300°C, hardening peaks, thereby causing the tendency to soften and significantly causing thermal strain and temper color. With the retention time of about 30 seconds or more, hardening can be sufficiently achieved at a heating temperature of about 200°C. In order to obtain more stable hardening, the retention time is preferably 60 seconds or more. However, retention for over 20 minutes is practically disadvantageous because further hardening cannot be expected,
and the production efficiency significantly deteriorates.

[0104] Therefore, in the present invention, the conventional coating and baking conditions, i.e., the heating temperature of 170°C and the retention time of 20 minutes, are set as the aging conditions. With the steel sheet of the present invention, hardening can be stably achieved even under the aging conditions of a low heating temperature and a short retention time, which fail to achieve sufficient hardening in a conventional bake-hardening steel sheet. The heating method is not limited, and atmospheric heating with a furnace, which is generally used for coating and baking, and other methods such as induction heating, heating with a nonoxidation flame, a laser, plasma, or the like, etc. can be preferably used.

[0105] The strength of an automobile part must be sufficient to resist an external complicated stress load, and thus not only strength in a low strain region but also strength in a high strain region are important for a raw material steel sheet. In consideration of this point, in the steel sheet of the present invention used as a raw material for automobile parts, BH is 80 MPa or more, and ΔTS is 40 MPa or more. More preferably, BH is 100 MPa or more, and ΔTS is 50 MPa or more. In order to further increase BH and TS, the heating temperature in aging may be set to a higher temperature, and/or the retention time may be set to a longer time.

[0106] The steel sheet of the present invention has the advantage that when the steel sheet is allowed to stand at room temperature for about one week without heating after forming, an increase in strength of about 40% of that at the time of complete aging can be expected.

[0107] The steel sheet of the present invention also has the advantage that even when it is allowed in an unmolded state at room temperature for a long time, aging deterioration (an increase in YS and a decrease in El (elongation)) does not occur, unlike a conventional aging steel sheet. In order to prevent the occurrence of a trouble in actual press forming, it is necessary that in aging at room temperature for 3 months before press forming, an increase in YS is 30 MPa or less, a decrease in elongation is 2% or less, and a recovery of yield point elongation is 0.2% or less.

[0108] In the present invention, the surface of the cold-rolled steel sheet may be coated by hot-dip galvanization or alloying hot-dip galvanization without any problem, and TS, BH and ΔTS are equivalent to those before plating. AS the plating method, electro-galvanization, hot-dip galvanization, alloying hot-dip galvanization, electro-tinning, electric chromium plating, electro-nickelizing, and the like may be preferably used.

[0109] For reference, description will now be made of forming conditions and conditions for subsequent heat treatment for increasing strength when the steel sheet of the present invention is molded, for example, press-molded. When the steel sheet of the present invention is subjected to press working, for example, deep drawing, the strain introduced by press working is several % to several tens%. Although the amount of strain changes with molded parts, a strain of about 5 to 10% is introduced into an inner plate and a structural member in the automobile field.

[0110] These automobile parts are heat-treated by coating and baking. However, with the steel sheet of the present invention, strength of a molded product can be effectively increased after heat treatment. In the present invention, as a method of evaluating burning hardenability in a laboratory, a tensile test specimen of JIS No. 5 size is obtained from the steel sheet in the rolling direction, and tensile strain of 10% is applied to the tensile test specimen by a tensile testing machine. Then, the specimen is heat-treated and again subjected to a tensile test. Particularly, when properties are evaluated after heat treatment in a low temperature region, the heat treatment conditions include 120°C and 20 minutes. In this test, the properties of the completed portion after heat treatment subsequent to press forming are evaluated.

[0111] Namely, in the present invention, the difference (ΔTS) between the tensile strength of the specimen after application of tensile strain and heat treatment and the tensile strength of a product is defined as the strength increasing ability of heat treatment.

[0112] In order to increase the strength of the molded product, the amount of strain introduced by forming, or the heat treatment temperature after processing is preferably as high as possible.

[0113] However, with the steel sheet of the present invention, when the amount of applied strain is about 5 to 10%, the strength can be sufficiently increased even by heat treatment at a temperature lower than conventional heat treatment, i.e., a temperature of 200°C or less, after forming. However, with a heat treatment temperature of less than 120°C, the strength cannot be sufficiently increased with the low train applied. On the other hand, with the heat treatment temperature of over 350°C after forming, softening proceeds. Therefore, the temperature of heat treatment after forming is preferably about 120 to 350°C.

[0114] The heating method is not limited, and hot gas heating, infrared furnace heating, hot-bath heating, direct current heating, induction heating, and the like can be used. Alternatively, only a portion where strength is desired to be increased is selectively heated.

Examples

[0115] In the examples below, the amount of dissolved N, the microstructure, tensile properties, the r value, strain age hardenability, and aging property were examined. The examination methods were as follows:
(1) Amount of dissolved N

The amount of dissolved N was determined by subtracting the amount of precipitated N from the total N amount of steel determined by chemical analysis. The amount of precipitated N was determined by an analysis method using a constant-potential electrolytic method.

(2) Microstructure

A test specimen was obtained from each of cold-rolled annealed steel sheets, and the microstructure of a section (C section) perpendicular to the rolling direction was imaged with an optical microscope or a scanning electron microscope. Then, the fraction of the ferrite texture and the type and the structure fraction of a second phase were determined by an image analysis apparatus.

(3) Crystal grain diameter

In the present invention, the value used as the average crystal grain diameter was a higher one of the value calculated from a photograph of a sectional structure by a quadrature method defined by ASTM, and the nominal value determined from a photograph of a sectional structure by an intercept method defined by ASTM (refer to, for example, Umemoto et al.: Heat Treatment, 24 (1984), p334).

(4) Tensile properties

A test specimen of JIS No. 5 was obtained from each of cold-rolled annealed steel sheets in the rolling direction, and a tensile test was carried out with a strain rate of 3 \times 10^{-3}/s according to the regulations of JIS Z 2241 to determine yield stress YS, tensile strength TS, and elongation El.

(5) Strain age hardenability

A test specimen of JIS No. 5 was obtained from each of cold-rolled annealed steel sheets in the rolling direction, and a tensile strain of 5% was applied as pre-deformation. Then, the specimen was subjected to heat treatment corresponding to coating and baking at 170°C for 20 minutes, and a tensile test with a strain rate of 3 \times 10^{-3}/s was performed to determine the tensile properties (yield stress TSBH, tensile strength TSBH). Then, BH amount = YSBH - YS5%, and \Delta TS = TSBH - TS were calculated. YS5% represents deformation stress in 5% pre-deformation of the produced sheet, YSBH and TSBH represent yield stress and tensile strength, respectively, after pre-deformation and heat treatment, and TS represents the tensile strength of the produced sheet.

(6) Measurement of r value

A test specimen of JIS No. 5 was obtained from each of the cold-rolled annealed steel sheets in each of the rolling direction (L direction), the direction (D direction) at 45° with the rolling direction, and the direction (C direction) at 90° with the rolling direction. The width-direction strain and thickness-direction strain of each of the test specimens were determined when a uniaxial tensile strain of 15% was applied to each specimen, and the r value of each specimen in each of the directions was determined from the following ratio of width-direction strain to thickness-direction strain:

\[ r = \ln \left( \frac{w}{w_0} \right) / \ln \left( \frac{t}{t_0} \right) \]

(wherein w0 and t0 represent the width and thickness of a specimen before the test, and w and t represent the width and thickness of a specimen after the test). The mean value was determined by the following equation:

\[ r_{\text{mean}} = \frac{(r_L + 2r_D + r_D)}{4} \]

wherein rL represents the r value in the rolling direction (L direction), rD represents the r value in the direction (D direction) at 45° with the rolling direction, and rL represents the r value in the direction (C direction) at 90° with the rolling direction. In order to improve the precision of experiment, calculation was made by using changes in elongation strain and strain.
in the width direction on the assumption that the volume was constant.

(7) Aging properties

A test specimen of JIS No. 5 was obtained from each of cold-rolled annealed steel sheets in the rolling direction, and then subjected to aging at 50°C for 200 hours, followed by a tensile test. The difference in yield elongation \( \Delta Y_{El} \) between before and after aging was determined from the obtained results to evaluate aging properties at normal temperature. When \( \Delta Y_{El} \) was zero, it was evaluated that the specimen has non-aging properties and excellent natural aging resistance.

(8) Tensile strength after forming and heat treatment

A test specimen of JIS No. 5 was obtained from each of produced sheets in the rolling direction, and then a pre-strain of 10% was applied thereto. Then, heat treatment was conducted for 20 minutes at a conventional heat treatment temperature of 120°C and a temperature of 170°C corresponding to coating and baking, and then tensile strength was determined.

(9) Decrease (\( \Delta El \)) in total elongation by natural aging

The decrease (\( \Delta El \)) in total elongation by natural aging was determined as the difference between the total elongation measured with a specimen of JIS No 5 obtained from the produced sheet in the rolling direction, and the total elongation measured with a specimen of JIS No 5 separately obtained from the produced sheet in the rolling direction after accelerated aging (retention at 100°C for 8 hours) of natural aging.

Example 1

A steel slab having each of the compositions shown in Table 1 was hot-rolled to a hot-rolled sheet having a thickness of 3.5 mm, and then cold-rolled to a cold-rolled sheet having a thickness of 0.7 mm under the conditions shown in Table 2. Then, the cold-rolled sheet was recrystallized, annealed and further galvannealed in a continuous annealing line or a continuous annealing and galvanizing line. Then, the annealed sheet was temper-rolled with a rolling reduction ratio of 1.0% to produce a cold-rolled steel sheet and a galvannealed steel sheet having both sides coated with a weight of 45 g/m² per side. In Table 2, the finisher deliver temperatures in the hot rolling process of Steel Nos. 3 and 8 are less than the Ar\(_3\) transformation point, and the finisher deliver temperatures of the others are the Ar\(_3\) transformation point or more.

The thus-obtained cold-rolled steel sheets and the galvannealed steel sheets were measured with respect to tensile strength, the \( r \) value, and a change in tensile strength after forming and heat treatment. The results are shown in Table 3.

Table 3 indicates that with all the cold-rolled steel sheets and the galvannealed steel sheets obtained according to the present invention, a high \( r \) value and excellent strain age hardenability are obtained, as compared with comparative examples. Particularly, in the suitable examples in which the crystal grain diameter is 20 \( \mu \)m or less, the decrease in elongation due to natural aging is also as low as 2.0% or less.

Example 2

A slab of steel symbol B shown in Table 1 was hot-rolled under the same production conditions as No. 2 shown in Table 2 in which the heating temperature was 1100°C, and the finisher deliver temperature of hot rolling was 900°C, and then coiled at coiling temperature of 550°C into a coil. The thus-obtained coil was cold-rolled with a reduction ratio of 80%, and then recrystallized and annealed at 840°C. With respect to the product properties of the resultant cold-rolled steel sheet, tensile strength TS was 365 MPa, and the \( r \) value was 1.7. A test specimen of JIS No. 5 was obtained from the cold-rolled steel sheet in the rolling direction, and a tensile strain of 10% was applied by a tensile test machine. Then, the specimen was subjected to heat treatment under the heat treatment conditions (temperature and time) shown in Table 4, and a tensile test was again performed. Table 4 also shows the increase in tensile strength (\( \Delta TS \)) from the tensile strength (TS = 365 MPa) of a product before application of strain.

Table 4 indicates that the increase in strength increases as the heat treatment temperature increases, and the heat treatment time increases. However, with the steel sheet of the present invention, a sufficient increase in tensile strength of 82 MPa (85% or more of an increase in heat treatment for 20 minutes) can be obtained even by heat treatment at low temperature of 120°C for a short retention time of 2 minutes. It is thus found that with the steel sheet of the present invention, good strain age hardenability can be obtained even by heat treatment at a low temperature for a short time.
In order to obtain the stable effect of increasing strength of an automobile structural member, or the like, heat treatment at a normal temperature for a normal time causes no problem. It was confirmed that with the galvanized steel sheets and the galvannealed steel sheets obtained by hot-dip galvanizing and heat alloying the cold-rolled sheets, the same results as shown in Table 4 are obtained.

All the examples of the present invention have excellent strain age hardenability and a high r value, and exhibit extremely high stable BH amount, ΔTS and mean r value regardless of variations in production conditions. It was also recognized that in the examples of the present invention, by performing continuous rolling and controlling the temperature of the sheet bar in the long direction and the width direction, the thickness precision and the shape of the produced steel sheet are improved, and variations in material properties are decreased.

Industrial Availability

According to the present invention, a cold rolled steel sheet can be obtained, in which TS is greatly increased by press forming and heat treatment while maintaining excellent deep drawability in press forming. The cold-rolled steel sheet has the excellent effect of industrially producing coated steel sheets by electro-galvanization, hot-dip galvanization, alloying hot-dip galvanization.
<table>
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<tr>
<th>Steel</th>
<th>C</th>
<th>N</th>
<th>Si</th>
<th>Mn</th>
<th>B</th>
<th>Al</th>
<th>Nb</th>
<th>P</th>
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</tr>
<tr>
<td>H</td>
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<td>0.020</td>
<td>0.35</td>
<td>0.11</td>
<td>0.0014</td>
<td>0.020</td>
<td>0.025</td>
<td>0.007</td>
<td>0.005</td>
<td>0.0041</td>
<td>-0.0011</td>
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<tr>
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<td>0.002</td>
<td>0.01</td>
<td>0.12</td>
<td>0.0007</td>
<td>0.038</td>
<td>0.055</td>
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</tr>
<tr>
<td>J</td>
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<td>0.50</td>
<td>0.12</td>
<td>0.0008</td>
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<td>0.001</td>
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<td>0.009</td>
<td>0.005</td>
<td>0.0027</td>
<td>0.0026</td>
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</table>

* Equation (1)': \( N% = \frac{14}{93} \times Nb\% + \frac{14}{27} \times Al\% + \frac{14}{11} \times B\% \) (the suitable range of this invention is 0.0015 or more)

** Equation (2)': \( C% = \frac{0.5}{12} \times Nb\% \) (the suitable range of this invention is 0 or less)
<table>
<thead>
<tr>
<th>No.</th>
<th>Steel</th>
<th>Slab heating temperature (°C)</th>
<th>Finisher delivery temperature (°C)</th>
<th>Cooling condition after finish rolling (s, °C/s)</th>
<th>Coiling temperature (°C)</th>
<th>Cold rolling reduction ratio (%)</th>
<th>Heating rate (°C/s)</th>
<th>Recrystallization annealing temperature (°C)</th>
<th>Presence of alloying hot-dip galvanization</th>
<th>Remarks</th>
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</thead>
<tbody>
<tr>
<td>1</td>
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<td>Finisher delivery temperature (°C)</td>
<td>Cooling condition after finish rolling (s, °C/s)</td>
<td>Coiling temperature (°C)</td>
<td>Cold rolling reduction ratio (%)</td>
<td>Heating rate (°C/s)</td>
<td>Recrystallization annealing temperature (°C)</td>
<td>Presence of alloying-hot-dip galvanization</td>
<td>Remarks</td>
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*The finisher delivery temperature was lower than Ar₃ transformation point. The cooling conditions after finish rolling include the cooling start time(s) and the cooling rate (°C/s).*
<table>
<thead>
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<th>No.</th>
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<th>Charge in tensile strength after forming-heat treatment</th>
<th>Remarks</th>
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<td></td>
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<td>Crystal grain diameter (μm)</td>
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<td>2.3</td>
<td>19</td>
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<td>14</td>
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<tr>
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<td>E</td>
<td>430</td>
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<td>18</td>
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<td>1.6</td>
<td>18</td>
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<tr>
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</table>
1. A cold-rolled steel sheet having excellent strain age hardenability comprising a composition, by mass %,

C: less than 0.01%;
Si: 0.005 to 1.0%;
Mn: 0.01 to 1.5%;
P: 0.1% or less;
S: 0.01% or less;
Al: 0.005 to 0.030%;
N: 0.005 to 0.040%;
B: 0.0001 to 0.0030%;
Nb: 0.005 to 0.050%; and

optionally further comprising at least one of Cu, Ni and Mo in a total of 1.0% or less, and the balance is composed of Fe and inevitable impurities, wherein N/Al is 0.30 or more, the amount of dissolved N is 0.0010% or more, and the ranges of B and Nb satisfy the following equations (1) and (2),

\[ N\% \geq 0.0015 + \frac{14}{93} \cdot Nb\% + \frac{14}{27} \cdot Al\% + \frac{14}{11} \cdot B\% \quad \ldots \quad (1) \]

\[ C\% \leq 0.5 \cdot \left( \frac{12}{93} \right) \cdot Nb\% \quad \ldots \quad (2) \]

2. A cold-rolled steel sheet having excellent strain age hardenability according to Claim 1, wherein the steel sheet has a crystal grain diameter of 20 \( \mu m \) or less.

3. A cold-rolled steel sheet having excellent strain age hardenability according to Claim 1 or 2, wherein strength after forming is increased by 60 MPa or more by heat treatment in a low temperature region of 120 to 200°C.

Table 4

<table>
<thead>
<tr>
<th>Retention time (min.)</th>
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<th>200</th>
<th>300</th>
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<td>82</td>
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<td>86</td>
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<tr>
<td>20</td>
<td>95</td>
<td>125</td>
<td>140</td>
</tr>
</tbody>
</table>

Table 5

<table>
<thead>
<tr>
<th>Al %</th>
<th>N/Al</th>
<th>TS x r value MPa</th>
<th>ΔTS MPa</th>
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</thead>
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<tr>
<td>0.020</td>
<td>0.75</td>
<td>775</td>
<td>58</td>
</tr>
<tr>
<td>0.036</td>
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<td>762</td>
<td>55</td>
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<tr>
<td>0.049</td>
<td>0.31</td>
<td>753</td>
<td>42</td>
</tr>
<tr>
<td>0.072</td>
<td>0.21</td>
<td>720</td>
<td>25</td>
</tr>
<tr>
<td>0.080</td>
<td>0.19</td>
<td>719</td>
<td>19</td>
</tr>
</tbody>
</table>
4. An electro-galvanized, hot-dip galvanized or alloyed hot-dip galvanized steel sheet having excellent strain age hardenability comprising a coated layer formed on the surface of a cold-rolled steel sheet according to Claims 1 to 3 by electro-galvanization, hot-dip galvanization, or alloying hot-dip galvanization.

5. A method of producing a cold-rolled steel sheet having excellent strain age hardenability comprising hot-rolling a steel slab under conditions in which the steel slab is cooled immediately after the end of finish rolling and coiled at a coiling temperature of 400 to 800°C, cold-rolling the hot-rolled sheet with a rolling' reduction ratio of 60 to 95%, and then recrystallizing and annealing the cold-rolled sheet at a temperature of 650 to 900°C, wherein the steel slab has a composition according to claim 1.

6. A method of producing a cold-rolled steel sheet having excellent strain age hardenability according to Claim 5, wherein in the heating-up step in the recrystallization annealing, the temperature is increased at a rate of 1 to 20°C/s in the temperature region from 500°C to the recrystallization temperature.

7. A method of producing an galvannealed steel sheet having excellent strain age hardenability comprising hot-dip galvanization and then heat alloying after recrystallization and annealing according to Claim 5 or 6.

**Patentansprüche**

1. Ein kaltgewalztes Stahlblech mit exzellenter Reckalterungseigenschaft, umfassend eine Zusammensetzung, in Gew.-%:
   - C: weniger als 0,01 %;
   - Si: 0,005 bis 1,0 %;
   - Mn: 0,01 bis 1,5 %;
   - P: 0,1 % oder weniger;
   - S: 0,01 % oder weniger;
   - Al: 0,005 bis 0,030 %;
   - N: 0,005 bis 0,040 %;
   - B: 0,0001 bis 0,0030 %;
   - Nb: 0,005 bis 0,050 %; und
   wahlweise ferner umfassend wenigstens eine von Cu, Ni und Mo in einer Gesamtmenge von 1,0 % oder weniger, und der Rest besteht aus Fe und unvermeidbaren Verunreinigungen, wobei N/Al 0,30 oder mehr ist, die Menge an gelöstem N 0,0010 % oder mehr ist und die Bereiche von B und Nb die folgenden Gleichungen (1) und (2) erfüllen:

   \[ N \% \geq 0,0015 + 14/93 \times Nb \% + 14/27 \times Al \% + 14/11 \times B \% \quad \ldots (1) \]

   \[ C \% \leq 0,5 \times (12/93) \times Nb \% \quad \ldots (2) \]

2. Ein kaltgewalztes Stahlblech mit exzellenter Reckalterungseigenschaft nach Anspruch 1, wobei das Stahlblech einen Kristallkorndurchmesser von 20 μm oder weniger aufweist.

3. Ein kaltgewalztes Stahlblech mit exzellenter Reckalterungseigenschaft nach Anspruch 1 oder 2, wobei die Festigkeit nach Umformen durch Wärmebehandlung in einem Niedrigtemperaturbereich von 120 bis 200°C mit 60 MPa oder mehr erhöht ist.

4. Ein galvanisch verzinktes, feuerverzinktes oder legiertes, feuerverzinktes Stahlblech mit exzellenter Reckalterungseigenschaft, umfassend eine auf der Oberfläche eines kaltgewalzten Stahlblechs gemäß Anspruch 1 bis 3 durch galvanisches Verzinken, Feuerverzinken oder legiertes Feuerverzinken beschichtete Schicht.


Revendications

1. Tôle d’acier laminée à froid présentant une excellente aptitude au vieillissement par contrainte comprenant la composition suivante, en % en masse :

   C : moins de 0,01 % ;
   Si : 0,005 à 1,0 % ;
   Mn : 0,01 à 1,5 % ;
   P : 0,1 % ou moins ;
   S : 0,01 % ou moins ;
   Al : 0,005 à 0,030 % ;
   N : 0,005 à 0,040 % ;
   B : 0,0001 à 0,0030 % ;
   Nb : 0,005 à 0,050 % ; et

   comprenant, en outre, éventuellement, au moins l’un de Cu, Ni et Mo pour une quantité totale de 1,0 % ou moins, et le reste étant composé de Fe et des impuretés inévitables,

dans laquelle N/Al est égal à 0,30 ou plus, la quantité de N dissous est égale à 0,0010 % ou plus, et les plages de B et Nb satisfont les équations (1) et (2) suivantes :

\[ \% \, N \geq 0,0015 + 14/93 \times \% \, \text{Nb} + 14/27 \times \% \, \text{Al} + 14/11 \times \% \, \text{B} \quad (1) \]

\[ \% \, C \leq 0,5 \times (12/93) \times \% \, \text{Nb} \quad (2). \]

2. Tôle d’acier laminée à froid présentant une excellente aptitude au vieillissement par contrainte selon la revendication 1, dans laquelle la tôle d’acier a un diamètre de grain cristallin de 20 μm ou moins.

3. Tôle d’acier laminée à froid ayant une excellente aptitude au vieillissement par contrainte selon la revendication 1 ou 2, dans laquelle la résistance après le formage augmente de 60 MPa ou plus, grâce à un traitement thermique dans une région de températures basses de 120 à 200 °C.

4. Tôle d’acier électro-galvanisé, galvanisé par trempage à chaud ou d’acier allié galvanisé par trempage à chaud présentant une excellente aptitude au vieillissement par contrainte et comprenant une couche enduite formée sur la surface d’une tôle d’acier laminée à froid selon les revendications 1 à 3 par électro-galvanisation, galvanisation par trempage à chaud ou galvanisation par trempage à chaud d’alliage.

5. Procédé de production d’une tôle d’acier laminée à froid ayant une excellente aptitude au vieillissement par contrainte et comprenant le laminage à chaud d’une brame dans des conditions dans lesquelles la brame d’acier est retroidie
immédiatement après la fin du brunissage de finition et bobinée à une température de bobinage de 400 à 800 °C, le laminage à froid de la tôle d’acier laminée à chaud avec un rapport de réduction au laminage de 60 à 95 %, puis la recristallisation et le recuit de la tôle d’acier laminée à froid à une température de 650 à 900 °C, dans lequel la brame d’acier présente une composition selon la revendication 1.

6. Procédé de production d’une tôle d’acier laminée à froid ayant une excellente aptitude au vieillissement par contrainte selon la revendication 5, dans lequel, au cours de l’étape de chauffage pendant le recuit de recristallisation, la température augmente à une cadence de 1 à 20 °C/s dans la région de températures comprises entre 500 °C et la température de recristallisation.

7. Procédé de production d’une tôle d’acier recuit après galvanisation ayant une excellente aptitude au vieillissement par contrainte et comprenant la galvanisation par trempage à chaud, puis un alliage à chaud après recristallisation et recuit selon la revendication 5 ou 6.
FIG. 6

- BH (MPa)
- ΔTS (MPa)
- TS × r VALU (MPa)

B × 10^4 (%)