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(54) **COMPOSITE STRUCTURE TYPE HIGH TENSILE STRENGTH STEEL PLATE, PLATED PLATE OF COMPOSITE STRUCTURE TYPE HIGH TENSILE STRENGTH STEEL AND METHOD FOR THEIR PRODUCTION**

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(57) **ABSTRACT**

The invention proposes a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability, wherein the steel sheet has a composition comprising C: 0.01-0.08 mass %, Si: not more than 2.0 mass %, Mn: not more than 3.0 mass %, P: not more than 0.10 mass %, S: not more than 0.02 mass %, A1: 0.005-0.20 mass %, N: not more than 0.02 mass % and V: 0.01-0.5 mass %, provided that V and C satisfy a relationship of $0.5\times C/12 \leqq V/51 \leqq 3\times C/12$, and the remainder being Fe and inevitable impurities, and has a microstructure consisting of a ferrite phase as a primary phase and a secondary phase including martensite phase at an area ratio of not less than 1% to a whole of the microstructure and a high-strength dual-phase galvanized steel sheet comprising a galvanized coating on the above steel sheet as well as a method of producing the same.

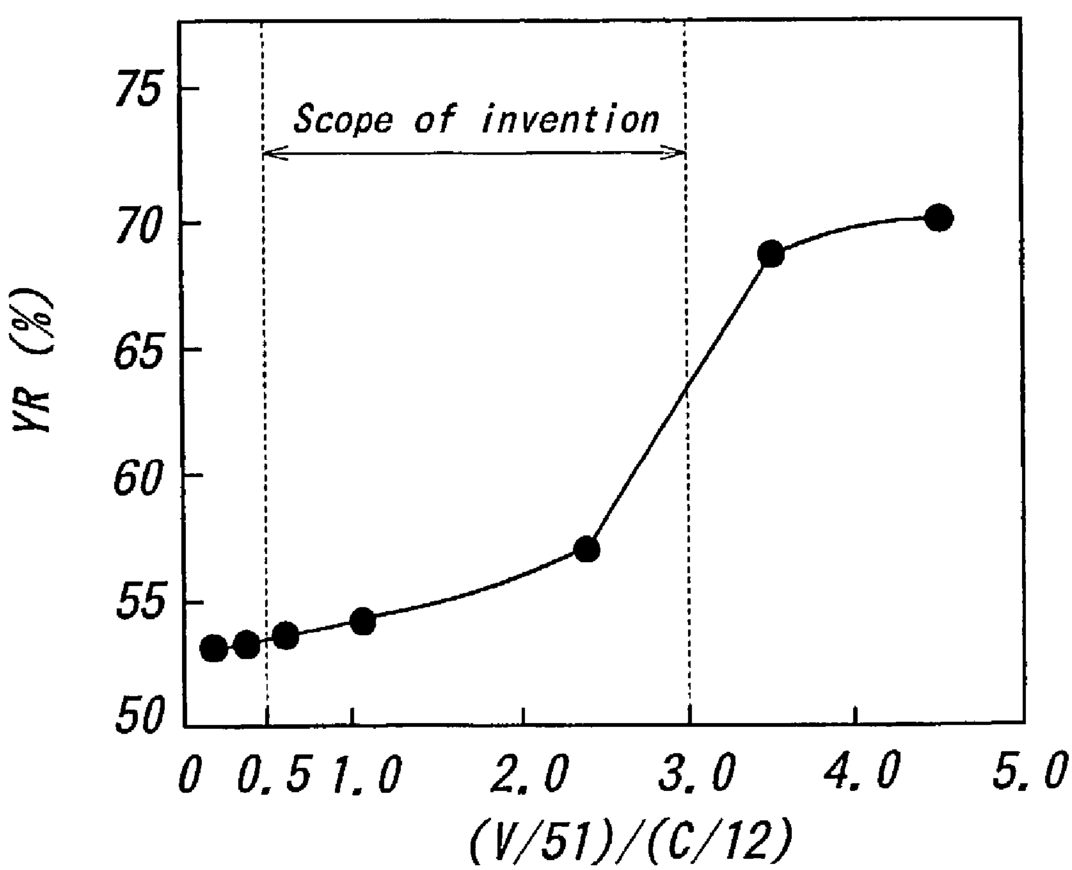


FIG. 1a

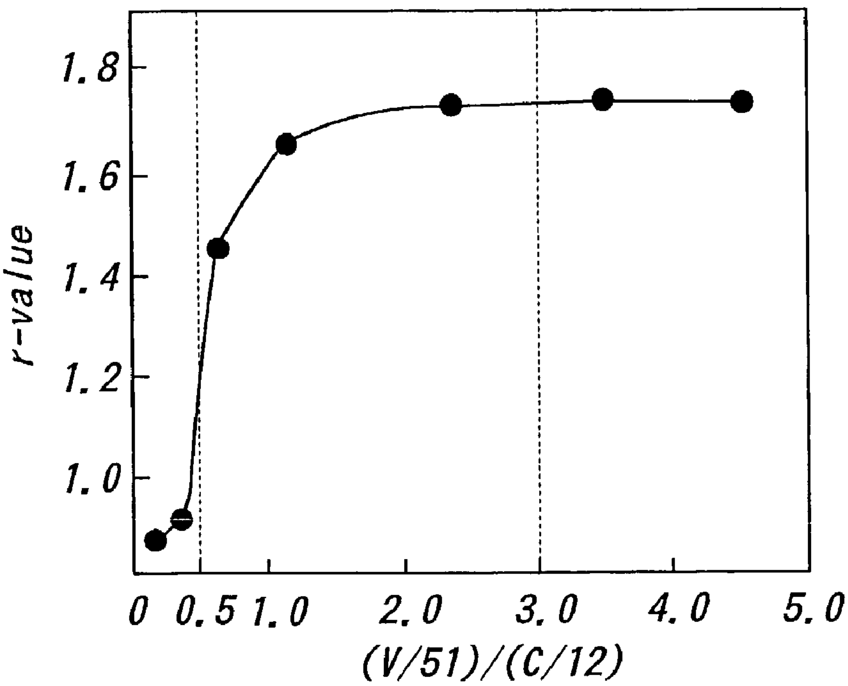

1.8
1.6
1.4
1.2
1.0
r-value
0
0.5
1.0
2.0
3.0
4.0
5.0
(V/51)/(C/12)

FIG. 1b

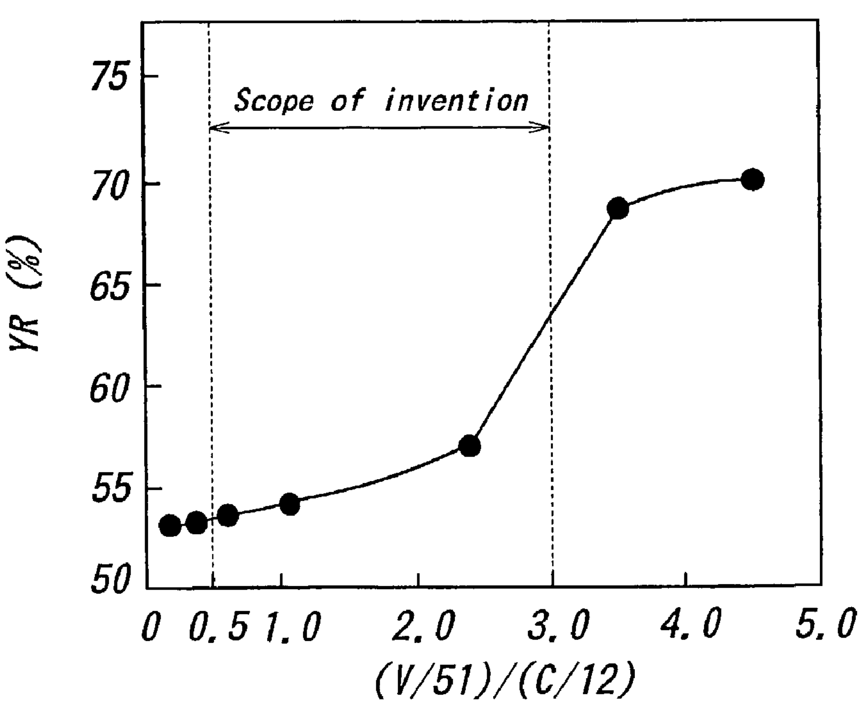

75
70
65
60
55
50
YR (%)
Scope of invention
0
0.5
1.0
2.0
3.0
4.0
5.0
(V/51)/(C/12)

FIG. 2a

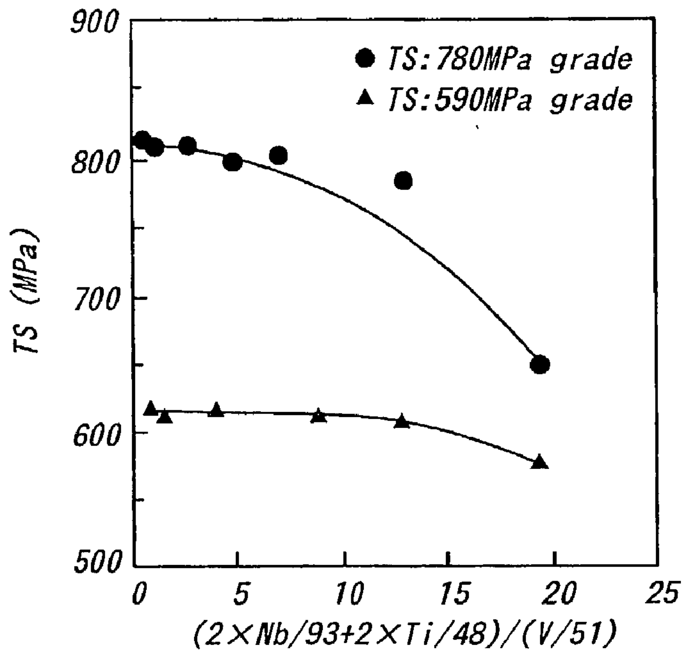
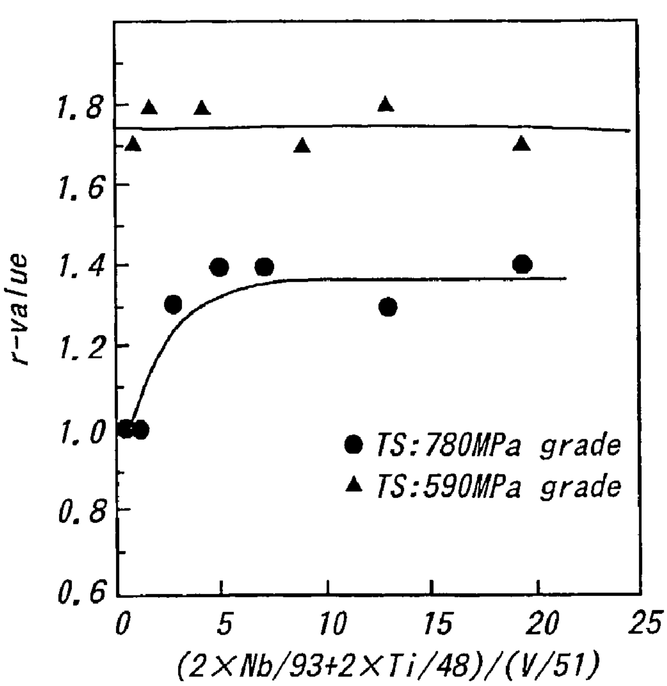

TS:780MPa grade
TS:590MPa grade
TS (MPa)
900
800
700
600
500
0
5
10
15
20
25
(2×Nb/93+2×Ti/48)/(V/51)

FIG. 2b


TS:780MPa grade
TS:590MPa grade
r-value
1.8
1.6
1.4
1.2
1.0
0.8
0.6
0
5
10
15
20
25
(2×Nb/93+2×Ti/48)/(V/51)

COMPOSITE STRUCTURE TYPE HIGH TENSILE STRENGTH STEEL PLATE, PLATED PLATE OF COMPOSITE STRUCTURE TYPE HIGH TENSILE STRENGTH STEEL AND METHOD FOR THEIR PRODUCTION

TECHNICAL FIELD

[0001] This invention relates to a high-strength dual-phase steel sheet having an excellent deep drawability, and particularly to a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability and a high strength dual phase galvanized steel sheet having an excellent deep drawability which have a tensile strength of 440 MPa or more and are suitable for use in steel sheets for vehicles as well as a method of producing the same.

BACKGROUND ART

[0002] Recently, it is required to improve a fuel consumption in a vehicle from a viewpoint of the maintenance of the global environment, and also it is required to improve a safety of a vehicle body from a viewpoint of the protection of crews during the collision of the vehicle. To this end, investigations for achieving both the lightening and strengthening of the vehicle body are positively proceeding.

[0003] In order to simultaneously satisfy the lightening and strengthening of the vehicle body, it is said that the high-strengthening of raw materials constituting the parts is effective, and recently, high-strength steel sheets are positively used as a part of the vehicle.

[0004] Most of the parts for the vehicle body are formed by press working of the steel sheet as a raw material. To this end, the high-strength steel sheet used is required to have an excellent press formability. In order to improve the press formability, it is necessary to have a high Lankford value (r-value), a high ductility (E1) and a low yield stress (YS) as mechanical properties of the steel sheet.

[0005] However, in general, as the steel sheet becomes highly strengthened, the r-value and the ductility lower and the press formability is degraded, while the yield stress rises to degrade the shapability and hence the problem of spring-back is apt to occur.

[0006] And also, a high corrosion resistance is required according to a position of the vehicle part to be applied, so that various surface-treated steel sheets having an excellent corrosion resistance are used as a steel sheet for the vehicle parts up to now. Among these surface-treated steel sheets, a galvanized steel sheet is manufactured in a continuous galvanizing equipment conducting recrystallization annealing and galvanizing at the same line, so that the provision of an excellent corrosion resistance and a cheap production are possible. And also, an alloyed galvanized steel sheet obtained by subjecting to a heat treatment after the galvanization is excellent in the weldability and press formability in addition to the excellent corrosion resistance. Therefore, they are widely used.

[0007] In order to further advance the lightening and strengthening of the vehicle body, in addition to the development of the high-strength cold rolled steel sheet having the excellent press formability, it is desired to develop a high-strength galvanized steel sheet having an excellent corrosion resistance through the continuous galvanizing line.

[0008] As a typical example of the high-strength steel sheet having a good press formability is mentioned a dual-phase steel sheet having a dual-phase microstructure of a soft ferrite phase and a hard martensite phase. Especially, the dual-phase steel sheet produced by cooling with a gas jet after the continuous annealing is low in the yield stress and possesses a high ductility and an excellent baking hardenability. The above dual-phase steel sheet is generally good in the workability, but has a drawback that the workability under severer condition is poor and particularly, the r-value is low and the deep drawability is bad.

[0009] And also, when the galvanization is applied for providing the excellent corrosion resistance, the continuous galvanizing line is general to set up the annealing equipment and the plating equipment continuously. To this end, in case of subjecting to the galvanization, the cooling after the annealing is constrained by a plating temperature and can not drop down to a temperature lower than the plating temperature at once and hence the cooling is interrupted. At a result, an average cooling rate necessarily becomes smaller. Therefore, when the galvanized steel sheet is produced in the continuous galvanizing line, it is difficult to generate martensite phase produced under a cooling condition of a large cooling rate into the steel sheet after the galvanization. To this end, it is generally difficult to produce the high-strength galvanized steel sheet having a dual-phase microstructure of a ferrite phase and a martensite phase through the continuous galvanizing line.

[0010] Under such unfavorable conditions, it is attempted to increase the r-value of the dual-phase steel sheet to improve the deep drawability. For example, JP-B-55-10650 discloses a technique that a box annealing is carried out at a temperature ranging from a recrystallization temperature to A_{c3} transformation point after the cold rolling and thereafter the continuous annealing inclusive of quenching and tempering is carried out after the heating to 700-800° C. in order to obtain the mixed microstructure. In this method, however, the quenching and tempering are carried out during the continuous annealing, so that the yield stress is high and hence a low yield ratio can not be obtained. The steel sheet having such a high yield stress is not suitable for the press formability and has a drawback that the shapability in the pressed parts is bad.

[0011] And also, a method for lowering the high yield stress is disclosed in JP-A-55-100934. In this method, the box annealing is first carried out in order to obtain a high r-value, wherein the temperature in the box annealing is made to a two-phase region of ferrite (α)-austenite (γ) and Mn is enriched from α phase to γ phase during the soaking. As the Mn enriched phase preferentially becomes γ phase during the continuous annealing, the dual-phase microstructure is obtained even at a cooling rate as in the gas jet cooling, and further the yield stress becomes low. In this method, however, it is required to conduct the box annealing at a relatively high temperature being the α-γ two-phase region over a long time for enriching Mn, so that there are many problems in production steps such as a frequent occurrence of adhesion between steel sheets inside a coil resulted from the thermal expansion in the annealing, an occurrence of temper color, a lowering of service life in an inner cover for a furnace body and the like. Therefore, it was difficult to industrially stably produce high-strength steel sheets possessing a high r-value and a low yield stress up to now.

[0012] In addition, JP-B-1-35900 discloses a technique wherein the dual-phase cold rolled steel sheet having a very high r-value and a low yield stress of r-value=1.61, YS=224 MPa and TS=482 MPa can be produced by cold rolling a steel having a composition of 0.012 mass % C-0.32 mass % Si-0.53 mass % Mn-0.03 mass % P-0.051 mass % Ti, heating to 870° C. corresponding to α-γ two-phase region and thereafter cooling at an average cooling rate of 100° C./s. However, the high cooling rate of 100° C./s is difficult to attain in the gas jet cooling usually used in the continuous annealing line or continuous galvanizing line after the cold rolling, and is required to use an equipment for water-quenching, and also a problem becomes actual in the surface treatment of the water-quenched steel sheet, so that there are problems in the production equipment and the materials.

[0013] Furthermore, it is attempted to produce the high-strength dual-phase galvanized steel sheet. In the past, as the method of producing the high-strength dual-phase galvanized steel sheet is generally used a method wherein the formation of low-temperature transformation phase is facilitated by using a steel added with a large amount of an alloying element such as Cr or Mo for enhancing a hardenability. However, the addition of the large amount of the alloying element undesirably brings about the rise of the production cost.

[0014] Moreover, as is disclosed in JP-B-62-40405 and the like, there is proposed a method of producing the high-strength dual-phase galvanized steel sheet by defining the cooling rate after the annealing or the plating in the continuous galvanizing line. However, this method is not actual from the constraint on the equipment for the continuous galvanizing line and also the steel sheet obtained by this method is not said to have a sufficient ductility.

DISCLOSURE OF THE INVENTION

[0015] It is, therefore, an object of the invention to solve the aforementioned problems and to provide high-strength dual-phase cold rolled steel sheets having an excellent deep drawability and high-strength dual-phase galvanized steel sheets having an excellent deep drawability as well as a method of producing the same.

[0016] Moreover, the term "galvanized steel sheet" used herein means to include a galvanized steel sheet obtained by subjecting to a galvanization containing aluminum or the like in addition to zinc and an alloyed galvanized steel sheet obtained by subjecting to a heat (alloying) treatment for diffusing iron of the matrix steel sheet into the plated layer after the galvanization.

[0017] In order to achieve the above object, the inventors have made various studies with respect to an influence of the alloying element upon the microstructure and the recrystallization texture in the steel sheet. As a result, it has been found that by limiting C in a steel slab to a lower content and rationalizing V content in relation to C content, before the recrystallization annealing, C in the steel is precipitated as a V carbide to decrease solid-solute C as far as possible to thereby develop {111} recrystallization texture to obtain a high r-value and subsequently the V carbide is dissolved by heating to α-γ two-phase region to enrich C in austenite for easily generating martensite in a subsequent cooling process, whereby the high-strength dual-phase cold rolled steel sheet and high-strength dual-phase galvanized steel sheet having a high r-value and an excellent deep drawability can be produced stably.

[0018] The results of fundamental experiments performed by the inventors will be explained below.

[0019] In this case, the experiments are performed with respect to a high-strength dual-phase cold rolled steel sheet of TS: 590 MPa grade and a high-strength dual-phase cold rolled steel sheet of TS: 780 MPa grade.

[0020] Firstly, the fundamental experiment in the high-strength dual-phase cold rolled steel sheet of TS: 590 MPa grade is performed under the following conditions. Each of various sheet bars having a basic composition of C: 0.03 mass %, Si: 0.02 mass %, Mn: 1.7 mass %, P: 0.01 mass %, S: 0.005 mass %, Al: 0.04 mass % and N: 0.002 mass % and different V contents by adding V within a range of 0.03-0.55 mass % is heated to 1250° C. and soaked, and then subjected to three-pass rolling at a finisher delivery temperature of 900° C. to obtain a hot rolled steel sheet having a thickness of 4.0 mm.

[0021] In addition, the same procedure as described above is conducted with respect to various sheet bars having a basic composition of C: 0.03 mass %, Si: 0.02 mass %, Mn: 1.7 mass %, P: 0.01 mass %, S: 0.005 mass %, Al: 0.04 mass % and N: 0.002 mass % and different values of (2×Nb [mass %]/93+2×Ti [mass %]/48)/(V [mass %]/51) by adding V, Nb and Ti within ranges of 0.03-0.04 mass %, 0.01-0.18 mass % and 0.01-0.18 mass %, respectively, so as to satisfy a relationship of $0.5\times C\ [\text{mass}\ \%]/12 \leqq (V\ [\text{mass}\ \%]/51+2\times Nb\ [\text{mass}\ \%]/93+2\times Ti\ [\text{mass}\ \%]/48) \leqq 3\times C\ [\text{mass}\ \%]/12$.

[0022] Moreover, the hot rolled steel sheet after the finish rolling is subjected to a temperature holding treatment of 650° C.×1 hour as a coiling treatment. Subsequently, the sheet is subjected to a cold rolling at a rolling reduction of 70% to obtain a cold rolled steel sheet having a thickness of 1.2 mm. Next, the cold rolled steel sheet is subjected to a recrystallization annealing at 850° C. for 60 seconds and cooled at a cooling rate of 30° C./s.

[0023] On the other hand, the fundamental experiment in the high-strength dual-phase cold rolled steel sheet of TS:780 MPa grade is performed under the following conditions.

[0024] Each of various sheet bars having a basic composition of C: 0.04 mass %, Si: 0.70 mass %, Mn: 2.6 mass %, P: 0.04 mass %, S: 0.005 mass %, Al: 0.04 mass % and N: 0.002 mass % and different values of (2×Nb/93+2×Ti/48)/(V/51) by adding V, Nb and Ti within ranges of 0.02-0.06 mass %, 0.01-0.12 mass % and 0.01-0.12 mass %, respectively, so as to satisfy a relationship of $0.5\times C\ [\text{mass}\ \%]/12 \leqq (V\ [\text{mass}\ \%]/51+2\times Nb\ [\text{mass}\ \%]/93+2\times Ti\ [\text{mass}\ \%]/48) \leqq 3\times C\ [\text{mass}\ \%]/12$is heated to 1250° C. and soaked, and then subjected to three-pass rolling at a finisher delivery temperature of 900° C. to obtain a hot rolled steel sheet having a thickness of 4.0 mm. Moreover, the sheet after the finish rolling is subjected to a temperature holding treatment of 650° C.×1 hour as a coiling treatment. Subsequently, the sheet is subjected to a cold rolling at a rolling reduction of 70% to obtain a cold rolled steel sheet having a thickness of

1.2 mm. Next, the cold rolled steel sheet is subjected to a recrystallization annealing at 850° C. for 60 seconds and cooled at a cooling rate of 30° C./s.

[0025] With respect to the thus obtained cold rolled steel sheets is conducted out a tensile test to investigate tensile properties. The tensile test is carried out by using JIS No. 5 tensile test piece. The r-value is determined as an average r-value $\{=(r_L+_C+2\times r_D)/4\}$ in a rolling direction (r_L), a direction (r_D) inclined at 45 degree with respect to the rolling direction and a direction (r_C) perpendicular (90°) to the rolling direction.

[0026] **FIGS. 1***a* and **1***b* show an influence of V content in a steel slab upon r-value and yield ratio of a cold rolled steel sheet (YR=yield stress (YS)/tensile strength (TS)×100(%)) in cold rolled steel sheets of TS: 590 MPa grade produced by using a steel slab containing V but not containing Nb and Ti, V. Moreover, an abscissa in **FIGS. 1***a* and **1***b* is an atomic ratio ((V/51)/(C/12)) of V content to C content, and an ordinate is r-value in **FIG. 1***a* and yield ratio (YR) in **FIG. 1***b*.

[0027] As seen from **FIGS. 1***a* and **1***b*, a high r-value and a low yield ratio are obtained by limiting V content in the steel slab to a range of 0.5-3.0 as the atomic ratio to C content and it is possible to produce high-strength dual-phase cold rolled steel sheet having an excellent deep drawability.

[0028] In the steel sheet according to the invention, the inventors found that a high r-value is obtained because solid-solute C and N are less and {111} recrystallization texture is strongly developed before the recrystallization annealing. And also, the inventors found that by annealing at α-γ two-phase region is dissolved V carbide and the solid-solute C is enriched into austenite phase in large quantity and the austenite can be easily transformed into martensite in the subsequent cooling process to obtain a dual-phase microstructure of ferrite and martensite.

[0029] Although Ti and Nb have mainly been used as a carbide forming element in the past, the inventors paid notice to V having a solubility of carbide higher than those of Ti and Nb for effectively obtaining the solid-solute C by annealing at a higher temperature region. That is, it is found that since V carbide easily dissolves as compared with Ti carbide and Nb carbide in the annealing at a high temperature, a sufficient amount of solid-solute C for transforming austenite to martensite is obtained by annealing at the α-γ two-phase region. In addition, it is clear that this phenomenon is most remarkably generated by V, but the similar result is obtained by adding Nb and Ti together.

[0030] Although the invention is based on the above knowledge, the following knowledge is obtained to achieve another invention.

[0031] The inventors compared r-values in the high-strength dual-phase cold rolled steel sheets of TS: 590 MPa grade and TS: 780 MPa produced by using steel slabs containing Nb and Ti in addition to V and made clear the followings. **FIGS. 2***a* and **2***b* show an influence of V, Nb and Ti contents in the steel slab upon tensile strength (TS) and Lankford value (r-value) of a cold rolled steel sheet in the cold rolled steel sheets of TS: 590 MPa grade and TS: 780 MPa grade produced by using the V, Nb and Ti containing steel slab. Moreover, an abscissa in **FIGS. 2***a* and **2***b* is an atomic ratio (2×Nb/93+2×Ti/48)/(V/51) of Nb and Ti contents to V content, and an ordinate is tensile strength (TS) in **FIG. 2***a* and r-value in **FIG. 2***b*.

[0032] According to the above results, in the TS: 780 MPa grade, the high-strengthening is attempted by large quantities of solid-solution strengthening elements, so that the r-value is lowered as compared with that of the TS: 590 MPa grade by the increase of the solid-solute C content or the like. In the TS: 780 MPa grade, however, the r-value is considerably improved when the value of (2×Nb/93+2×Ti/48)/(V/51) is a range of not less than 1.5. Such a characteristic in the TS: 780 MPa grade that the r-value is remarkably improved when the value of (2×Nb/93+2×Ti/48)/(V/51) is a range of not less than 1.5 is not recognized in the TS: 590 MPa grade.

[0033] Although the detail of causes on the above result is not clear, it is considered that in the system containing a large amount of an element resulted in the lowering of the r-value such as solid-solute C or the like as in the TS: 780 MPa grade, Nb and Ti easily precipitate the solid-solute C and N as a compound as compared with V and the solid-solute C and N contents after the hot rolling become less to improve the r-value. Moreover, when the value of (2×Nb/93+2×Ti/48)/(V/51) exceeds 15, TS considerably lowers, which is unfavorable for obtaining the high-strength dual-phase cold rolled steel sheet of TS: 780 MPa grade. This is considered due to the fact that as Nb carbide and Ti are hardly dissolved as compared with V carbide, if the addition quantities of the Nb and Ti contents are larger than that of the V content, the C content enriched in austenite phase is largely decreased in the annealing at the α-γ two-phase region is widely decreased and martensite phase generated after the cooling is softened.

[0034] The invention is accomplished by further examining based on the above knowledge. The summary of the invention is as follows.

[0035] (1) A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability, characterized in that the steel sheet has a composition comprising C: 0.01-0.08 mass %, Si: not more than 2.0 mass %, Mn: not more than 3.0 mass %, P: not more than 0.10 mass %, S: not more than 0.02 mass %, Al: 0.005-0.20 mass %, N: not more than 0.02 mass % and V: 0.01-0.5 mass % provided that V and C satisfy a relationship represented by the following equation (i):

$$0.5\times C/12 \leqq V/51 \leqq 3\times C/12 \quad \text{(i)}$$

[0036] and the remainder being Fe and inevitable impurities, and has a microstructure consisting of a ferrite phase as a primary phase and a secondary phase including martensite phase at an area ratio of not less than 1% to a whole of the microstructure.

[0037] (2) A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to the item (1), wherein the steel sheet has a composition comprising further not more than 0.3 mass % in total of one or tow of Nb: more than 0 mass % but not more than 0.3 mass % and Ti: more than 0 mass % but not more than 0.3 mass % provided that V, Nb, Ti and C satisfy a relationship represented by the following equation (ii) instead of the equation (i):

$$0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48) \leqq 3\times C/12 \quad \text{(ii)}$$

[0038] and the remainder being Fe and inevitable impurities.

[0039] Moreover, it is preferable that one or two of Nb: 0.001-3.0 mass % and Ti: 0.001-0.3 mass % is not more than 0.3 mass % in total.

[0040] (3) A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to the item (2), wherein the steel sheet comprises C: 0.03-0.08 mass %, Si: 0.1-2.0 mass %, Mn: 1.0-3.0 mass %, P: not more than 0.05 mass % and S: not more than 0.01 mass % and V, Nb and Ti satisfy a relationship of $1.5 \leqq (2\times Nb/93+2\times Ti/48)/(V/51) \leqq 15$.

[0041] (4) A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to any one of the items (1) to (3), wherein the steel sheet further comprises one or two of the following A group and B group:

[0042] A group: not more than 2.0 mass % in total of one or two of Cr and Mo;

[0043] B group: not more than 2.0 mass % in total of one or two of Cu and Ni.

[0044] (5) A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability, which comprises hot rolling a steel slab having a composition comprising C: 0.01-0.08 mass %, Si: not more than 2.0 mass %, Mn: not more than 3.0 mass %, P: not more than 0.10 mass %, S: not more than 0.02 mass %, Al: 0.005-0.20 mass %, N: not more than 0.02 mass % and V: 0.01-0.5 mass % provided that V and C satisfy a relationship represented by the following equation (iii):

$$0.5\times C/12 \leqq V/51 \leqq 3\times C/12 \quad \text{(iii)}$$

[0045] and the remainder being Fe and inevitable impurities, pickling, cold rolling and then subjecting to a continuous annealing at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point.

[0046] (6) A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to the item (5), wherein the steel sheet has a composition comprising further not more than 0.3 mass % in total of one or tow of Nb: more than 0 mass % but not more than 0.3 mass % and Ti: more than 0 mass % but not more than 0.3 mass % provided that V, Nb, Ti and C satisfy a relationship represented by the following equation (iv) instead of the equation (iii):

$$0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48) \leqq 3\times C/12 \quad \text{(iv)}$$

[0047] and the remainder being Fe and inevitable impurities.

[0048] Moreover, it is preferable that one or two of Nb: 0.001-0.3 mass % and Ti: 0.001-0.3 mass % is not more than 0.3 mass % in total.

[0049] (7) A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to the item (6), wherein the steel slab comprises C: 0.03-0.08 mass %, Si: 0.1-2.0 mass %, Mn: 1.0-3.0 mass %, P: not more than 0.05 mass % and S: not more than 0.01 mass % and V, Nb and Ti satisfy a relationship of $1.5 \leqq (2\times Nb/93+2\times Ti/48)/(V/51) \leqq 15$.

[0050] (8) A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to any one of the items (5)-(7), wherein the steel slab further comprises one or two of the following A-group and B-group:

[0051] A-group: not more than 2.0 mass % in total of one or two of Cr and Mo;

[0052] B-group: not more than 2.0 mass % in total of one or two of Cu and Ni.

[0053] (9) A high-strength dual-phase galvanized steel sheet having an excellent deep drawability comprising a galvanized coating on the steel sheet disclosed in any one of the items (1)-(4).

[0054] (10) A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability, wherein a galvanization is carried out after the continuous annealing at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point in the production method described in any one of the items (5)-(7).

[0055] (11) A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability according to the item (10), which further comprising a continuous annealing step between the cold rolling step and the continuous annealing step at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point.

[0056] (12) A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability according to the item (10) or (11), wherein the steel slab further comprises one or two of the following A-group and B-group:

[0057] A-group: not more than 2.0 mass % in total of one or two of Cr and Mo;

[0058] B-group: not more than 2.0 mass % in total of one or two of Cu and Ni.

[0059] The cold rolled steel sheet and the galvanized steel sheet according to the invention are high-strength dual-phase steel sheets having a tensile strength (TS) of not less than 440 MPa and an excellent deep drawability.

[0060] At first, the reason of limiting the composition in the cold rolled steel sheet and the galvanized steel sheet according to the invention will be explained below. Moreover, mass % represents simply as "%".

[0061] C: 0.01-0.08%

[0062] C is an element for increasing the strength of the steel sheet and further promoting the formation of a dual-phase microstructure of ferrite and martensite, and is necessary to contain not less than 0.01%, preferably not less than 0.015% from a viewpoint of the formation of the dual-phase microstructure in the invention. Moreover, if it is intended to increase the strength to TS: not less than 540 MPa and TS: not less than 780 MPa, the C content is preferable to be not less than 0.015% and not less than 0.03%, respectively. On the other hand, when the C content exceeds 0.08%, the development of {111} recrystallization texture is obstructed to degrade the deep drawability. Therefore, the invention limits the C content to 0.01-0.08%. When it is particularly required to increase the strength of the steel sheet, it is preferable to be 0.03-0.08%. Moreover, it is preferable to be not more than 0.05% from a viewpoint of the deep drawability.

[0063] Si: Not More Than 2.0%

[0064] Although Si is a useful reinforcing element capable of increasing the strength of the steel sheet without remarkably lowering the ductility of the steel sheet, if the content exceeds 2.0%, the deterioration of the deep drawability is caused, but also the surface properties are degraded. Therefore, Si is limited to not more than 2.0%. Moreover, if it is intended to increase the strength to TS: not less than 780 MPa, it is preferable to be not less than 0.1% for ensuring the required strength. And also, it is preferable to be not less than 0.01% for increasing the strength to TS: not less than 440 MPa which is a main object of the invention.

[0065] Mn: Not More Than 3.0%

[0066] Mn has an action reinforcing the steel and further has an action of lessening a critical cooling rate for the obtention of the dual-phase microstructure of ferrite and martensite to promote the formation of the dual-phase microstructure of ferrite and martensite, so that it is preferable to contain a content in accordance with the cooling rate after the recrystallization annealing. And also, Mn is an effective element preventing the hot tearing through S, so that it is preferable to contain an appropriate content in accordance with S content. However, when the Mn content exceeds 3.0%, the deep drawability and weldability are degraded. In the invention, therefore, the Mn content is limited to not more than 3.0%. Moreover, the Mn content is preferable to be not less than 0.5% for remarkably developing the above effect, and particularly it is preferable to be not less than 1.0% for increasing the strength to TS: not less than 780 MPa. And also, it is preferable to be not less than 0.1% for increasing the strength to TS: not less than 440 MPa which is a main object of the invention.

[0067] P: Not More Than 0.10%

[0068] P has an action reinforcing the steel and can be contained in a required amount in accordance with the desired strength. When the P content exceeds 0.10%, the press formability is degraded. Therefore, the P content is limited to not more than 0.10%. Moreover, if a more excellent press formability is required, the P content is preferable to be not more than 0.08%. Furthermore, when large quantities of C, Mn and the like are contained in order to ensure TS: not less than 780 MPa, the P content is preferable to be not more than 0.05% in order to prevent the degradation of the weldability. In addition, if it is intended to increase the strength to TS: not less than 440 MPa, it is preferable to be not less than 0.001%.

[0069] S: Not More Than 0.02%

[0070] S is existent as an inclusion in the steel sheet and is an element bringing about the degradation of the ductility and the formability of the steel sheet, particularly the stretch-flanging property. Therefore, it is preferable to be decreased as far as possible, and when it is decreased to not more than 0.02%, S does not exert a bad influence, so that the S content is 0.02% as an upper limit in the invention. Moreover, when the more excellent stretch-flanging property is required, or when the large quantities of C, Mn and the like are contained in order to ensure TS: not less than 780 MPa, if the excellent weldability is required, the S content is preferable to be not more than 0.01%, more preferably not more than 0.005%. On the other hand, the S content is preferable to be not less than 0.0001% considering a cost for the removal of S in the steelmaking process.

[0071] Al: 0.005-0.20%

[0072] Al is added to the steel as a deoxidizing element and is a useful element for improving the cleanliness of the steel, but the addition effect is not obtained at less than 0.005%. On the other hand, when it exceeds 0.20%, the more deoxidizing effect is not obtained and the deep drawability is inversely degraded. Therefore, the Al content is limited to 0.005-0.20%. Moreover, the invention does not exclude a steelmaking method through deoxidization other than the Al deoxidization. For example, Ti deoxidization or Si deoxidization may be conducted. The steel sheets made by these deoxidizing methods are included within a scope of the invention. In this case, even if Ca, REM and the like are added to the molten steel, the characteristics of the steel sheet according to the invention are not obstructed, so that the steel sheet including Ca, REM and the like is naturally included within the scope of the invention.

[0073] N: Not More Than 0.02%

[0074] N is an element increasing the strength of the steel sheet by the solid-solution hardening and the strain ageing hardening, but when N content exceeds 0.02%, the nitride is increased in the steel sheet to remarkably degrade the deep drawability of the steel sheet. Therefore, the N content is limited to not more than 0.02%. Moreover, in case of requiring the more improvement of the press formability, the N content is preferable to be not more than 0.01%, more preferably not more than 0.004%. In this case, considering the cost for denitrification in the steelmaking process, the N content is preferable to be not less than 0.0001%.

[0075] V: 0.01-0.5% and $0.5\times C/12 \leqq V/51 \leqq 3\times C/12$

[0076] V is a most important element in the invention. Before the recrystallization, the solid-solute C is precipitated and fixed as V carbide to develop the {111} recrystallization texture, whereby a high r-value can be obtained. Moreover, V dissolves the V carbide in the annealing at α-γ two-phase region to enrich a large quantity of the solid-solute C in austenite phase, which is easily transformed into martensite at the subsequent cooling process, whereby the dual-phase steel sheet having a dual-phase microstructure of ferrite and martensite can be obtained. Such an effect becomes effective when the V content is not less than 0.01%, more preferably not less than 0.02% and satisfies $0.5\times C/12 \leqq V/51$ in relation to the C content. On the other hand, when the V content exceeds 0.5% or when it is $V/51 > 3\times C/12$ in relation to the C content, the dissolution of the V carbide at the α-γ two-phase region hardly occurs and the dual-phase microstructure of ferrite and martensite is hardly obtained. Therefore, the V content is limited to 0.01-0.5% and to $0.5\times C/12 \leqq V/51 \leqq 3\times C/12$. Moreover, $V/51 \leqq 2\times C/12$ is preferable for obtaining the dual-phase microstructure of ferrite and martensite.

[0077] In addition to the above composition, it is further preferable to contain not more than 0.3 (mass) % in total of one or two of Nb: more than 0% but not more than 0.3 (mass) % and Ti: more than 0% but not more than 0.3%, and that V, Nb, Ti contents satisfy $0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48) \leqq 3\times C/12$ in relation to the C content in place of that the V and C content satisfy $0.5\times C/12 \leqq V/51 \leqq 3\times C/12$.

[0078] Not More Than 0.3% in Total of One or Tow of Nb: More Than 0% but not More Than 0.3% and Ti: More Than 0% but not More Than 0.3%, and V, Nb, Ti and C Satisfy $0.5\times C/12 \leqq (V/51+2-Nb/93+2\times Ti/48) \leqq 3\times C/12$

[0079] Nb and Ti are carbide forming elements likewise V and have the same action as V mentioned above. That is, a high r-value can be obtained by precipitating and fixing the solid-solute C as Nb and Ti carbides before the recrystallization to develop the {111} recrystallization texture, and also a dual-phase steel sheet having a dual-phase microstructure of ferrite and martensite can be obtained by dissolving the Nb and Ti carbides in the annealing at the α-γ two-phase region to enrich a large quantity of the solid-solute C in austenite phase and transforming into martensite in the subsequent cooling process. Moreover, as the above effect of Nb and Ti is considerably small as compared with that of V, when only Nb and Ti are added to the steel slab without adding V, the deep drawability aiming at the invention can not be enhanced sufficiently.

[0080] Therefore, it is preferable to add Nb and Ti of more than 0%. More preferably, each of the Nb and Ti contents is not less than 0.001%. In this case, it is preferable to satisfy $0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48)$ in relation to the C and V contents for developing the above effect. On the other hand, when each of Nb and Ti contents or both in total thereof exceeds 0.3%, or when the Nb and Ti contents satisfy $(V/51+2\times Nb/93+2\times Ti/48) > 3\times C/12$ in relation to the C and V contents, the dissolution of the carbide at the α-γ two-phase region hardly occurs and hence the dual-phase microstructure of ferrite and martensite is hardly obtained. Therefore, it is preferable that when either Nb or Ti is merely added, each of the Nb content and the Ti content is within a range of more than 0% but not more than 0.3%, and when both of Nb and Ti are added together, the Nb and Ti contents are not more than 0.3% in total and satisfy $0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48) \leqq 3\times C/12$ in relation to the V and C contents.

[0081] On the other hand, if it is intended to increase the strength to TS: not less than 780 MPa, the deep drawability is apt to be easily degraded by the addition of large quantities of solid-solution strengthening elements such as C, Mn and the like. In this case, the V, Nb and Ti contents are further desirable to be a range of $1.5 \leqq (2\times Nb/93+2\times Ti/48)/(V/51) \leqq 15$. The reason why $(2\times Nb/93+2\times Ti/48)/(V/51)$ is limited to not less than 1.5 is considered due to the fact that although the detail of the cause is not clear, the formation of carbide after the hot rolling is promoted to decrease the solid-solute C by adding large quantities of Nb and Ti as compared with V and hence the {111} recrystallization texture is easily developed. Moreover, in order to ensure the strength of TS: not less than 780 MPa, $(2\times Nb/93+2\times Ti/48)/(V/51)$ is desirable to be a range of not more than 15.

[0082] Furthermore, in addition to the above steel composition, the steel according to the invention is preferable to further comprise one or two of the following A-group and B-group:

[0083] A-group: not more than 2.0% in total of one or two of Cr and Mo;

[0084] B-group: not more than 2.0% in total of one or two of Cu and Ni.

[0085] A-Group: Not More Than 2.0% in Total of One or Two of Cr and Mo

[0086] All of Cr and Mo in the A-group have an action of decreasing the critical cooling rate for providing the dual-phase microstructure of ferrite and martensite to promote the formation of the dual-phase microstructure of ferrite and martensite likewise Mn and can be included, if necessary. The lower limits of the Cr content and Mo content preferable for obtaining the above effect are Cr: 0.05%, Mn: 0.05%. However, when one or two of Cr and Mo exceed 2.0% in total, the deep drawability is degraded. To this end, one or more of Cr and Mo in the A-group is preferable to be limited to not more than 2.0% in total.

[0087] B-Group: Not More Than 2.0% in Total of One or Two of Cu and Ni

[0088] Cu and Ni in the B-group have an action of reinforcing the steel and may be included at a required amount in accordance with the desired strength. However, when the content of Cu and Ni added alone or together exceeds 2.0% in total, it tends to degrade the deep drawability. To this end, one or more of Cu and Ni is preferable to be not more than 2.0% in total. Moreover, the lower limits of the Cu and Ni contents preferable for obtaining the above effect is Cu: 0.05% and Ni: 0.05%, respectively.

[0089] Although elements other than the above elements are not particularly limited in the invention, there is no problem even if B, Ca, Zr, REM and the like is included within a range of the usual steel composition.

[0090] In this case, B is an element having an action of improving the hardenability in the steel and may be included, if necessary. However, when the B content exceeds 0.003%, the above effect is saturated, so that the B content is preferable to be not more than 0.003%. Moreover, a more desirable range is 0.001-0.002%. Ca and REM have an action of controlling the form of sulfide inclusion and also have an effect of improving the stretch-flanging property. Such an effect is saturated when one or two selected from Ca and REM exceed 0.01% in total. To this end, the content of one or two of Ca and REM is preferable to be not more than 0.01% in total. Moreover, a more preferable range is 0.001-0.005%.

[0091] The reminder other than the above elements is Fe and inevitable impurities. As the inevitable impurity are mentioned, for example, Sb, Sn, Zn, Co and the like. As acceptable ranges of their contents are Sb: not more than 0.01%, Sn: not more than 0.1%, Zn: not more than 0.01% and Co: not more than 0.1%.

[0092] Next, the microstructure of the steel sheet according to the invention will be explained.

[0093] The cold rolled steel sheet according to the invention has a microstructure consisting of ferrite phase as a primary phase and a secondary phase including not less than 1% of martensite phase at an area ratio with respect to a whole of the microstructure.

[0094] In order to provide the cold rolled steel sheet having a low yield stress (YS), a high ductility (E1) and an excellent deep drawability, it is required to render the microstructure of the steel sheet according to the invention into a dual-phase microstructure consisting of a ferrite phase as a primary phase and a secondary phase including a martensite phase. It is preferable that the ferrite phase as a primary phase is not less than 80% at an area ratio and hence the secondary phase is not more than 20%. When the area ratio of the ferrite phase is less than 80%, it is difficult to ensure the high ductility and the press formability tends to lower. And also, when a good ductility is required, it is preferable that the ferrite phase is not less than 85% at the area ratio and hence the secondary phase is not more than 15%. Moreover, in order to utilize the advantage of the dual-phase microstructure, the ferrite phase is required to be not more than 99%.

[0095] In the invention, the secondary phase is required to include the martensite phase at the area ratio of not less than 1% with respect to the whole of the microstructure. When the martensite is less than 1% at the area ratio, the low yield stress (YS) and the high ductility (E1) can not be satisfied simultaneously. More preferably, the martensite phase is not less than 3% but not more than 20% at the area ratio. In case of requiring a good ductility, the martensite phase is preferable to be not more than 15% at the area ratio. Moreover, the secondary phase may be constituted by only the martensite phase at the area ratio of not less than 1% or by mixed phases of the martensite phase at the area ratio of not less than 1% and any of a pearlite phase, a bainite phase and a retained austenite as an additional phase and is not especially limited. In the latter case, the pearlite phase, the bainite phase and the retained austenite are preferable to be not more than 50% in total at the area ratio with respect to the microstructure of the secondary phase in order to more effectively develop the effect of the martensite phase.

[0096] The cold rolled steel sheet and the galvanized steel sheet having the above microstructure are steel sheets having a low yield stress, a high ductility and an excellent deep drawability.

[0097] Next, the method of producing the cold rolled steel sheet and the galvanized steel sheet according to the invention will be explained.

[0098] The composition of the steel slab used in the production method of the invention is the same as the compositions of the aforementioned cold rolled steel sheet and the galvanized steel sheet, so that the explanation on the reason of the limitation in the steel slab is omitted.

[0099] The cold rolled steel sheet according to the invention is produced by using a steel slab having a composition of the above range as a starting material and successively subjecting this starting material to a hot rolling step of subjecting to a hot rolling to obtain a hot rolled steel sheet, a pickling step of pickling the hot rolled steel sheet, a cold rolling step of subjecting the hot rolled steel sheet to a cold rolling to obtain a cold rolled steel sheet, and a recrystallization annealing step of subjecting the cold rolled steel sheet to a recrystallization annealing to obtain a cold rolled annealed steel sheet.

[0100] And also, the galvanized steel sheet according to the invention is produced by using a steel slab having a composition of the above range as a starting material and successively subjecting this starting material to a hot rolling step of subjecting to a hot rolling to obtain a hot rolled steel sheet, a pickling step of pickling the hot rolled steel sheet, a cold rolling step of subjecting the hot rolled steel sheet to a cold rolling to obtain a cold rolled steel sheet, and a continuous galvanization step of subjecting the cold rolled steel sheet to a recrystallization annealing and a galvanizing to obtain a galvanized steel sheet. Furthermore, it is produced by subjecting the cold rolled steel sheet to a step of annealing and pickling before the continuous galvanization step, if necessary.

[0101] The steel slab used is preferable to be produced by a continuous casting process in order to prevent the macrosegregation of the components, but may be produced by an ingot casting process or a thin slab casting process. Furthermore, in addition to the conventional process of cooling to a room temperature once after the production of the steel slab and again heating, energy-saving processes such as a process for inserting a hot steel slab into a heating furnace without cooling, a process for direct sending rolling or direct rolling immediately after slight heat-holding and the like can be applied without problems.

[0102] The above starting material (steel slab) is subjected to the hot rolling step of forming the hot rolled steel sheet by heating and hot rolling. In the hot rolling step, there is particularly no problem even in the use of usual rolling conditions as long as the hot rolled steel sheet having a desired thickness can be produced. Moreover, preferable hot rolling conditions are mentioned below for the reference.

[0103] Slab Heating Temperature: not Lower Than 900° C.

[0104] The slab heating temperature is desirable to be made lower as far as possible in order to improve the deep drawability by coarsening the precipitate to develop the {111} recrystallization texture. However, when the slab heating temperature is lower than 900° C., the rolling load increases and the risk of causing troubles in the hot rolling increases. To this end, the slab heating temperature is preferable to be not lower than 900° C. And also, the upper limit of the slab heating temperature is more preferable to be 1300° C. in terms of the lowering of the yield resulted from the increase of scale loss accompanied with the increase of the oxide weight. Moreover, it goes without saying that the utilization of a so-called sheet bar heater of heating the sheet bar in the hot rolling is an effective process from a viewpoint that the slab heating temperature is lowered and the troubles in the hot rolling are prevented.

[0105] Finisher Delivery Temperature: not Lower Than 700° C.

[0106] The finisher delivery temperature (FDT) is preferable to be not lower than 700° C. in order to obtain a uniform microstructure of the hot rolled parent sheet for providing an excellent deep drawability after the cold rolling and the recrystallization annealing. That is, when the finish deformation temperature is lower than 700° C., not only the microstructure of the hot rolled parent sheet becomes non-uniform, but also the rolling load in the hot rolling becomes higher and the risk of causing the trouble in the hot rolling is increased.

[0107] Coiling Temperature: not More Than 800° C.

[0108] The coiling temperature is preferable to be not higher than 800° C. That is, when the coiling temperature exceeds 800° C., the scale increases and the yield tends to lower due to the scale loss. And also, when the coiling temperature is lower than 200° C., the shape of the steel sheet remarkably is disordered and the risk of causing problems in the actual use increases, so that the lower limit of the coiling temperature is more preferable to be 200° C.

[0109] As mentioned above, in the hot rolling step according to the invention, it is preferable that the steel slab is heated above 900° C., subjected to the hot rolling at the finish deformation temperature of not lower than 700° C., and coiled at the coiling temperature of not higher than 800° C.

[0110] Moreover, in the hot rolling step according to the invention, a lubrication rolling may be conducted in a part of the finish rolling or between passes thereof in order to reduce the rolling load in the hot rolling. In addition, the application of the lubrication rolling is effective from a viewpoint of the uniformization of the steel sheet shape and the homogenization of the material. Also, the coefficient of friction in the lubrication rolling is preferable to be within a range of 0.10-0.25.

[0111] Further, the hot rolling step is preferable to be a continuous rolling process wherein the sheet bars located in front and rear are joined to each other and continuously subjected to the finish rolling. The application of the continuous rolling process is desirable from a viewpoint of the operating stability in the hot rolling.

[0112] Next, the hot rolled steel sheet is subjected to the pickling for the removal of the scale. The pickling step is sufficient according to the usual manner and it is preferable to use a treating solution such as hydrochloric acid, sulfuric acid or the like as a pickling solution.

[0113] Moreover, the cold rolled steel sheet is formed by subjecting the hot rolled steel sheet to the cold rolling. The cold rolling conditions are not especially limited as long as the cold rolled steel sheet having desired size and shape can be obtained, but it is preferable that a rolling reduction in the cold rolling is not less than 40%. When the rolling reduction is less than 40%, the {111} recrystallization texture is not developed and the excellent deep drawability can not be obtained.

[0114] The cold rolled steel sheet according to the invention is subjected to a recrystallization annealing in the subsequent recrystallization annealing step to obtain a cold rolled annealed steel sheet. The recrystallization annealing is carried out in a continuous annealing line. On the other hand, the galvanized steel sheet according to the invention is produced by subjecting the cold rolled steel sheet to recrystallization annealing and galvanizing in the continuous galvanization line after the cold rolling. In this case, the annealing temperature in the recrystallization annealing is required to be conducted at a (α+γ) two-phase region within a temperature range from A_{C1} transformation point to A_{C3} transformation point. This is due to the fact that the annealing is carried out at (α+γ) two-phase region to dissolve the carbides of V, Ti and Nb to thereby distribute an amount of solid-solute C sufficient to transform austenite to martensite into the austenite phase. When the annealing temperature is lower than the A_{C1} transformation point, the microstructure is rendered into the ferrite single phase and the martensite can not be generated, while when it is higher than the A_{C3} transformation point, the crystal grains are coarsened and the microstructure is rendered into the austenite single phase and the {111} recrystallization texture is not developed and hence the deep drawability is deteriorated remarkably.

[0115] In the cold rolled steel sheet according to the invention, the cooling in the recrystallization annealing is preferable to be conducted at a cooling rate of not less than 5° C./s in order to produce the martensite phase to obtain the dual-phase microstructure of ferrite and martensite.

[0116] On the other hand, in the galvanized steel sheet according to the invention, it is preferable to quench to a temperature region of 380-530° C. after the above recrystallization annealing. When a stop temperature of the quenching is lower than 380° C., the defective plating easily occurs, while when it exceeds 530° C., the unevenness easily occurs on the plated surface. Moreover, the cooling rate is preferable to be not less than 5° C./s in order to produce the martensite phase to obtain the dual-phase microstructure of ferrite and martensite. After the above quenching, the galvanization is carried out by dipping in a galvanizing bath. In this case, Al concentration in the galvanizing bath is preferable to be within a range of 0.12-0.145 mass %. When the Al concentration in the galvanizing bath is less than 0.12 mass %, the alloying excessively advances and the plating adhesion (resistance to powdering) tends to be deteriorated, while when it exceeds 0.145 mass %, the defective plating easily occurs.

[0117] And also, the plated layer may be subjected to an alloying treatment after the galvanization. Moreover, the alloying treatment is preferable to be conducted so that Fe content in the plated layer is 9-12%.

[0118] As the alloying treatment, it is preferable to conduct the alloying of the galvanized layer by reheating up to a temperature region of 450-550° C. After the alloying treatment, it is preferable to cool at a cooling rate of not less than 5° C./s to 300° C. The alloying at a high temperature is difficult to form the martensite phase and there is caused a fear of degrading the ductility of the steel sheet, while when the alloying temperature is lower than 450° C., the progress of the alloying is slow and the productivity tends to lower. Furthermore, when the cooling rate after the alloying treatment is extremely small, the formation of the martensite becomes difficult. To this end, the cooling rate at a temperature region from after the alloying treatment to 300° C. is preferable to be not less than 5° C./s.

[0119] Moreover, if it is required to further improve the plating property, it is preferable that after the cold rolling and before being subjected to the continuous galvanization, the annealing is separately conducted in the continuous annealing line and subsequently an enriched layer of components in the steel produced on the surface of the steel sheet is removed by pickling and thereafter the above treatment is conducted in the continuous galvanization line. In this case, the pickling may be carried out in the pickling line or in the pickling bath arranged in the continuous galvanization line. Also, the atmosphere in the continuous annealing line is preferable to be a reducing atmosphere with respect to the steel sheet in order to prevent the formation of the scale, and it is generally sufficient to use a nitrogen gas containing several % of H_2. The annealing is preferable to be conducted under a condition that a temperature of the steel sheet reaching in the continuous annealing line is not lower than the A_{C1} transformation point decided by the components in the steel. Because it is required to promote the enrichment of the alloying element on the surface of the steel sheet and to enrich the alloying element in the secondary phase by once forming the dual-phase microstructure in the continuous annealing line. In the steel sheet after the annealing in the continuous annealing line, there is a tendency that P among the components in the steel is diffused to segregate on the surface of the steel sheet and Si, Mn, Cr and the like enrich as an oxide, so that it is preferable to remove the enriched layer formed on the surface of the steel sheet by the pickling. Then, the same annealing as in the above is performed in the continuous galvanization line. In order to develop the characteristics as the dual-phase microstructure, the annealing in the continuous galvanization line is preferable to be performed at (α+γ) two-phase region within a temperature range of from the A_{C1} transformation point to the A_{C3} transformation point. In this case, the reason why the annealing is performed at not lower than the A_{C1} transformation point in both the continuous annealing line and the continuous galvanization line is due to the fact that the dual-phase microstructure is formed as mentioned above. Once an enriching place of the element as the secondary phase is formed by forming the dual-phase microstructure as a final microstructure in the continuous annealing line, it becomes possible to enrich the alloying element to some degree at this place. Desirably, it is sufficient to obtain the same dual-phase microstructure as in a final product after the cooling, so that the alloying element is more preferable to be enriched in the vicinity of a triple point of grain boundary (intersection of the grain boundary formed by three crystal grains). Thereafter, when the annealing is performed at the two-phase region in the continuous galvanization line, the alloying element is further enriched in the secondary phase or γ-phase and hence the γ-phase easily transforms into the martensite phase during the cooling process. Moreover, the term "alloying element" used herein means a substitutional alloying element such as Mn, Mo or the like, which makes a situation that diffusion hardly occurs and enrichment easily occurs at the temperature in the annealing step in order to lower the yield ratio.

[0120] And also, the cold rolled steel sheet after the recrystallization annealing process and the galvanized steel sheet after the plating process or after the alloying process may be subjected to a temper rolling at a rolling reduction of not more than 10% for correcting the shape and adjusting the surface roughness and the like. Furthermore, the cold rolled steel sheet according to the invention can be applied as not only a cold rolled steel sheet for the working but also a blank of a surface treated steel sheet for the working. As the surface treated steel sheet for the working are mentioned tin-plated steel sheets, porcelain enamels and so on in addition to the aforementioned galvanized steel sheets (including alloyed sheets). There is no problem even when they are subjected to a treatment such as resin or fat coating, various paintings, electroplating or the like. Moreover, the galvanized steel sheet according to the invention may be subjected to a special treatment after the galvanization in order to improve the chemical conversion property, weldability, press formability, corrosion resistance and the like.

BRIEF DESCRIPTION OF THE DRAWINGS

[0121] **FIG. 1***a* is a graph showing an influence of V and C contents in steel upon a Lankford value (r-value).

[0122] **FIG. 1***b* is a graph showing an influence of V and C contents in steel upon a yield ratio (YR=yield stress(YS)/tensile stress(TS)×100(%)).

[0123] **FIG. 2***a* is a graph showing an influence of a relationship among Nb, Ti and V contents upon a tensile strength (TS) in the high-strength dual-phase cold rolled steel sheets of TS: 590 MPa grade and TS: 780 MPa grade.

[0124] **FIG. 2***b* is a graph showing an influence of a relationship among Nb, Ti and V contents upon a Lankford value (r-value) in the high-strength dual-phase cold rolled steel sheets of TS: 590 MPa grade and TS: 780 MPa grade.

BEST MODE FOR CARRYING OUT THE INVENTION

[0125] Each of molten steels having compositions shown in Tables 1-4 is made in a converter and subjected to a continuous casting process to obtain a slab. In this case, each of the slabs having the compositions shown in Tables 1 and 2 is prepared for the purpose of experiments with respect to the cold rolled steel sheet, and each of the slabs having the compositions shown in Tables 3 and 4 is prepared for the purpose of experiments with respect to the galvanized steel sheet. Especially, the slabs shown in Tables 2 and 4 are prepared for the purpose of obtaining the cold rolled steel sheet and galvanized steel sheet of TS: not less than 780 MPa, respectively. Then, the steel slab is heated to 1150° C. and subjected to a hot rolling under conditions of a finish deformation temperature: 900° C. and a coiling temperature: 650° C. at a hot rolling step to obtain a hot rolled steel strip having a thickness of 4.0 mm. Subsequently, the hot rolled steel strip is pickled and subjected to a cold rolling at a rolling reduction of 70% at a cold rolling step to obtain a cold rolled steel strip or a cold rolled sheet having a thickness of 1.2 mm. Next, each of the cold rolled steel sheets in Tables 1 and 2 is subjected to a recrystallization annealing at an annealing temperature shown in Tables 5 and 6 in a continuous annealing line. The thus obtained cold rolled sheet is further subjected to a temper rolling at a rolling reduction of 0.8%. With respect to the galvanized steel sheets, each of the cold rolled sheets in Tables 3 and 4 is subjected to a recrystallization annealing at an annealing temperature shown in Tables 7 and 8 and further to a galvanizing in a galvanizing bath having an Al concentration of 0.13% in a continuous galvanization line. Moreover, with respect to a part of steel sheets (Steel sheet Nos. 52, 68, 69 and 70 in Table 7), the steel sheet after the cold rolling is subjected to an annealing at 830° C. in a continuous annealing line and then pickled and annealed and galvanized at a galvanizing bath temperature of 480° C. under an Al concentration in the bath of 0.13% in a continuous galvanization line and further the thus obtained steel strip (galvanized steel sheet) is subjected to a temper rolling at a rolling reduction of 0.8%. With respect to the steel sheets 75 and 77 in Table 7, they are subjected to an alloying treatment at an alloying temperature of 520° C. after the galvanization.

[0126] A test piece is cut out from the obtained steel strip and a microstructure thereof with respect to a section (C section) perpendicular to the rolling direction is imaged by using an optical microscope or a scanning electron microscope to measure a structure ratio of ferrite phase as a primary phase and a kind and a structure ratio of a secondary phase by using an image analysis device. In this case, a specimen for observing the microstructure is subjected to a mirror-like polishing and an etching with an alcohol solution containing 2% HNO_3 and then used for the observation. And also, a tensile test piece of JIS No. 5 is cut out from the steel strip and subjected to a tensile test according to the definition of JIS Z 2241 to measure a yield stress (YS), a tensile strength (TS), an elongation (E1), a yield ratio (YR) and a Lankford value (r-value). These results are shown in Tables 5-8.

TABLE 1(a)

Steel	Chemical composition (mass %)												
No.	C	Si	Mn	P	S	Al	N	V	Nb	Ti	Cr	Mo	Cu	Ni
1-A	0.030	0.02	1.55	0.01	0.004	0.032	0.002	0.132	—	—	—	—	—	—
1-B	0.028	0.02	1.48	0.01	0.001	0.032	0.002	0.105	0.042	—	—	0.15	—	—
1-C	0.032	0.03	1.72	0.01	0.005	0.028	0.002	0.085	0.035	0.035	0.05	—	—	—
1-D	0.020	0.02	1.63	0.01	0.005	0.033	0.002	0.065	—	—	—	—	0.12	0.08
1-E	0.031	0.02	1.56	0.01	0.006	0.033	0.002	0.122	0.045	—	—	0.18	—	—
1-F	0.029	0.02	1.48	0.01	0.003	0.032	0.002	0.210	0.115	0.125	—	—	—	—
1-G	0.032	0.02	1.65	0.01	0.004	0.032	0.002	0.045	—	—	—	—	—	—
1-H	0.020	0.22	2.02	0.06	0.004	0.032	0.002	0.132	—	—	—	—	—	
1-I	0.022	0.52	1.85	0.03	0.001	0.032	0.002	0.105	0.042	—	—	0.15	—	—
1-J	0.028	0.33	1.72	0.01	0.005	0.028	0.002	0.085	0.035	0.035	0.05	—	—	—
1-K	0.011	0.21	1.53	0.01	0.003	0.028	0.002	0.032	0.030	—	—	—	—	—
1-L	0.022	0.52	1.52	0.01	0.002	0.033	0.002	0.125	—	0.022	—	—	—	—
1-M	0.019	0.53	1.43	0.05	0.001	0.032	0.002	0.105	—	—	0.05	0.15	—	—

Steel				Transformation point (° C.)		
No.	X*1	Y*2	Z*3	A_{c1}	A_{c3}	Remarks
1-A	1.04	—	—	725	860	Acceptable example
1-B	—	1.27	0.44	705	855	Acceptable example
1-C	—	1.45	1.33	710	850	Acceptable example
1-D	0.76	—	—	715	855	Acceptable example
1-E	—	1.30	0.40	705	855	Acceptable example
1-F	—	4.88	1.26	725	855	Comparative example
1-G	0.33	—	—	715	850	Comparative example
1-H	1.55	—	—	725	860	Acceptable example
1-I	—	1.62	0.44	705	865	Acceptable example
1-J	—	1.66	1.33	710	860	Acceptable example
1-K	—	1.39	1.03	710	860	Acceptable example
1-L	—	1.84	0.37	715	865	Acceptable example
1-M	1.30	—	—	710	850	Acceptable example

(Note)
*1X = (V/51)/(C/12)
*2Y = (V/51 + 2 × Nb/93 + 2 × Ti/48)/(C/12)
*3Z = (2 × Nb/93 + 2 × Ti/48)/(V/51)

[0127]

TABLE 1(b)

Steel	Chemical composition (mass %)												
No.	C	Si	Mn	P	S	Al	N	V	Nb	Ti	Cr	Mo	Cu	Ni
1-N	0.021	0.33	1.72	0.06	0.003	0.030	0.002	0.115	—	—	0.05	0.15	0.15	0.15
1-O	0.020	0.41	2.02	0.02	0.002	0.029	0.002	0.072	0.042	0.010	0.05	0.15	0.10	0.10

TABLE 1(b)-continued

1-P	0.007	0.35	1.76	0.01	0.005	0.029	0.002	0.073	—	—	—	—	—	—
1-Q	0.112	0.33	1.74	0.01	0.003	0.028	0.002	0.352	—	—	—	—	—	—
1-R	0.021	0.52	1.52	0.01	0.002	0.033	0.002	0.008	—	—	—	—	—	—
1-S	0.023	0.53	1.43	0.05	0.001	0.032	0.002	0.622	—	—	—	—	—	—
1-T	0.021	0.33	1.72	0.06	0.003	0.030	0.002	0.049	0.0005	—	—	—	—	—
1-U	0.025	0.41	1.75	0.04	0.002	0.029	0.002	0.041	0.325	—	—	—	—	—
1-V	0.019	0.35	1.76	0.05	0.001	0.032	0.002	0.052	—	0.0005	—	—	—	—
1-W	0.023	0.33	1.72	0.06	0.003	0.030	0.002	0.033	—	0.306	—	—	—	—
1-X	0.018	0.02	1.48	0.01	0.003	0.032	0.002	0.030	0.001	0.001	—	—	—	—
1-Y	0.021	0.02	1.65	0.01	0.004	0.032	0.002	0.329	—	—	—	—	—	—

Steel				Transformation point (° C.)		
No.	X*1	Y*2	Z*3	A_{c1}	A_{c3}	Remarks
1-N	1.29	—	—	705	855	Acceptable example
1-O	—	1.64	0.93	715	850	Acceptable example
1-P	2.45	—	—	714	882	Comparative example
1-Q	0.74	—	—	714	859	Comparative example
1-R	0.09	—	—	722	879	Comparative example
1-S	6.36	—	—	723	972	Comparative example
1-T	—	0.56	0.01	714	910	Acceptable example
1-U	—	3.74	8.69	716	891	Comparative example
1-V	—	0.66	0.02	714	906	Acceptable example
1-W	—	6.99	19.7	714	1033	Comparative example
1-X	—	0.43	0.07	708	863	Comparative example
1-Y	3.69	—	—	706	886	Comparative example

(Note)
*1X = (V/51)/(C/12)
*2Y = (V/51 + 2 × Nb/93 + 2 × Ti/48)/(C/12)
*3Z = (2 × Nb/93 + 2 × Ti/48)/(V/51)

[0128]

TABLE 2

Steel	Chemical composition (mass %)												
No.	C	Si	Mn	P	S	Al	N	V	Nb	Ti	Cr	Mo	Cu	Ni
2-A	0.039	0.50	2.85	0.01	0.005	0.031	0.002	0.151	—	—	—	—	—	—
2-B	0.038	0.75	2.52	0.01	0.001	0.035	0.002	0.088	0.121	—	—	0.29	—	—
2-C	0.042	0.74	2.53	0.01	0.006	0.033	0.002	0.092	0.110	0.152	0.09	—	—	—
2-D	0.041	0.70	2.55	0.01	0.008	0.032	0.002	0.087	—	0.064	—	—	0.08	0.1
2-E	0.048	0.72	2.52	0.01	0.005	0.034	0.002	0.153	0.202	0.005	—	0.31	—	—
2-F	0.040	0.77	2.55	0.01	0.007	0.036	0.002	0.524	0.193	0.262	—	—	—	—
2-G	0.038	0.73	2.56	0.01	0.006	0.033	0.002	0.040	0.011	0.009	—	—	—	—
2-H	0.043	0.95	2.95	0.05	0.005	0.032	0.002	0.095	0.002	0.119	—	0.27	—	—
2-I	0.042	0.82	2.78	0.04	0.009	0.035	0.002	0.141	0.045	0.053	—	—	—	—
2-J	0.048	0.91	2.73	0.04	0.006	0.036	0.002	0.033	0.185	0.155	0.13	0.13	0.12	0.11
2-K	0.042	0.82	2.78	0.04	0.009	0.035	0.002	0.008	—	—		—	—	—
2-L	0.038	0.91	2.73	0.04	0.006	0.036	0.002	0.522	—	—		—	—	—
2-M	0.038	0.76	2.57	0.03	0.001	0.034	0.002	0.087	0.0005	—	—	—	—	—
2-N	0.039	0.76	2.55	0.03	0.001	0.035	0.002	0.032	0.056	—	—	—	—	—
2-O	0.042	0.73	2.49	0.03	0.001	0.036	0.002	0.092	0.382	—	—	—	—	—
2-P	0.043	0.75	2.52	0.03	0.001	0.035	0.002	0.088	0.453	—	—	—	—	—

TABLE 2-continued

2-Q	0.041	0.70	2.55	0.03	0.002	0.032	0.002	0.098	—	0.0005	—	—	—	—
2-R	0.038	0.71	2.58	0.03	0.002	0.030	0.002	0.025	—	0.037	—	—	—	—
2-S	0.039	0.74	2.57	0.04	0.002	0.030	0.002	0.079	—	0.186	—	—	—	—
2-T	0.042	0.71	2.51	0.02	0.002	0.029	0.002	0.089	—	0.356	—	—	—	—

Steel				Transformation point (° C.)		
No.	X*1	Y*2	Z*3	A_{c1}	A_{c3}	Remarks
2-A	0.91	—	—	707	842	Acceptable example
2-B	—	1.37	1.51	718	868	Acceptable example
2-C	—	3.00	4.82	719	915	Acceptable example
2-D	—	1.28	1.56	715	875	Acceptable example
2-E	—	1.89	1.52	717	870	Acceptable example
2-F	—	7.60	1.47	718	972	Comparative example
2-G	—	0.44	0.78	717	851	Comparative example
2-H	—	1.92	2.68	717	931	Acceptable example
2-I	—	1.70	1.15	717	900	Acceptable example
2-J	—	2.77	16.13	722	932	Acceptable example
2-K	0.04	—	—	717	866	Comparative example
2-L	3.23	—	—	720	923	Comparative example
2-M	—	0.54	0.01	718	868	Acceptable example
2-N	—	0.56	1.92	718	868	Acceptable example
2-O	—	2.86	4.55	718	868	Acceptable example
2-P	—	3.20	5.65	718	868	Comparative example
2-Q	—	0.57	0.01	715	875	Acceptable example
2-R	—	0.64	3.15	715	875	Acceptable example
2-S	—	2.86	5.00	715	875	Acceptable example
2-T	—	4.74	8.50	715	875	Comparative example

(Note)
*1X = (V/51)/(C/12)
*2Y = (V/51 + 2 × Nb/93 + 2 × Ti/48)/(C/12)
*3Z = (2 × Nb/93 + 2 × Ti/48)/(V/51)

[0129]

TABLE 3(a)

Steel	Chemical composition (mass %)													
No.	C	Si	Mn	P	S	Al	N	V	Nb	Ti	Cr	Mo	Cu	Ni
3-A	0.028	0.02	1.55	0.01	0.003	0.034	0.002	0.121	—	—	—	—	—	—
3-B	0.030	0.02	1.46	0.01	0.002	0.035	0.002	0.108	0.041	—	—	0.16	—	—
3-C	0.031	0.03	1.70	0.01	0.005	0.028	0.002	0.086	0.036	0.033	0.06	—	—	—
3-D	0.021	0.02	1.65	0.01	0.005	0.034	0.002	0.068	—	—	—	—	0.14	0.07
3-E	0.032	0.02	1.52	0.01	0.004	0.033	0.002	0.124	0.044	—	—	0.15	—	—
3-F	0.026	0.02	1.52	0.01	0.003	0.035	0.002	0.122	0.112	0.122	—	—	—	—
3-G	0.032	0.02	1.62	0.01	0.005	0.032	0.002	0.042	—	—	—	—	—	—
3-H	0.021	0.21	2.02	0.06	0.003	0.030	0.002	0.130	—	—	—	—	—	—

TABLE 3(a)-continued

Steel No.	C	Si	Mn	P	S	Al	N	V	Nb	Ti	Cr	Mo	Cu	Ni
3-I	0.024	0.52	1.88	0.04	0.001	0.032	0.002	0.105	0.033	—		0.16	—	—
3-J	0.026	0.32	1.72	0.01	0.004	0.026	0.002	0.088	0.035	0.032	0.08	—	—	—
3-K	0.020	0.70	1.55	0.01	0.003	0.028	0.002	0.073	0.045	—	—	—	—	—
3-L	0.012	0.21	1.51	0.01	0.002	0.033	0.002	0.055	—	0.018	—	—	—	—
3-M	0.018	0.50	1.56	0.03	0.004	0.035	0.002	0.108	—	—	0.05	0.15	—	—

Steel				Transformation point (° C.)		
No.	X*1	Y*2	Z*3	A_{c1}	A_{c3}	Remarks
3-A	1.02	—	—	725	860	Acceptable example
3-B	—	1.20	0.42	705	855	Acceptable example
3-C	—	1.48	1.27	710	850	Acceptable example
3-D	0.76	—	—	715	855	Acceptable example
3-E	—	1.27	0.39	705	855	Acceptable example
3-F	—	4.56	3.13	725	855	Comparative example
3-G	0.31	—	—	715	850	Comparative example
3-H	1.46	—	—	725	860	Acceptable example
3-I	—	1.38	0.34	705	860	Acceptable example
3-J	—	1.76	1.21	710	860	Acceptable example
3-K	—	1.44	0.68	715	870	Acceptable example
3-L	—	1.83	0.70	710	865	Acceptable example
3-M	1.41	—	—	710	860	Acceptable example

(Note)
*1X = (V/51)/(C/12)
*2Y = (V/51 + 2 × Nb/93 + 2 × Ti/48)/(C/12)
*3Z = (2 × Nb/93 + 2 × Ti/48)/(V/51)

[0130]

TABLE 3(b)

Steel	Chemical composition (mass %)													
No.	C	Si	Mn	P	S	Al	N	V	Nb	Ti	Cr	Mo	Cu	Ni
3-N	0.020	0.39	1.73	0.05	0.001	0.031	0.002	0.110	—	—	0.05	0.15	0.15	0.15
3-O	0.021	0.28	1.95	0.02	0.005	0.029	0.002	0.075	0.038	0.01	0.05	0.15	0.10	0.10
3-P	0.008	0.32	1.75	0.01	0.005	0.032	0.002	0.075	—	—	—	—	—	—
3-Q	0.095	0.34	1.73	0.01	0.003	0.029	0.002	0.361	—	—	—	—	—	—
3-R	0.023	0.49	1.54	0.01	0.002	0.030	0.002	0.007	—	—	—	—	—	—
3-S	0.024	0.51	1.47	0.03	0.001	0.031	0.002	0.597	—	—	—	—	—	—
3-T	0.022	0.35	1.75	0.05	0.003	0.029	0.002	0.109	0.0005	—	—	—	—	—
3-U	0.023	0.44	1.78	0.04	0.003	0.027	0.002	0.065	0.319	—	—	—	—	—
3-V	0.021	0.35	1.73	0.05	0.001	0.034	0.002	0.099	—	0.0005	—	—	—	—
3-W	0.025	0.36	1.77	0.05	0.002	0.032	0.002	0.132	—	0.321	—	—	—	—
3-X	0.020	0.02	1.51	0.01	0.003	0.033	0.002	0.035	0.001	0.001	—	—	—	—
3-Y	0.023	0.02	1.66	0.01	0.003	0.035	0.002	0.308	—	—	—	—	—	—

TABLE 3(b)-continued

Steel No.	X*1	Y*2	Z*3	Transformation point (° C.) A_{c1}	A_{c3}	Remarks
3-N	1.29	—	—	705	865	Acceptable example
3-O	—	1.55	0.84	715	865	Acceptable example
3-P	2.21	—	—	714	881	Comparative example
3-Q	0.89	—	—	714	872	Comparative example
3-R	<u>0.07</u>	—	—	722	874	Comparative example
3-S	<u>5.85</u>	—	—	721	949	Comparative example
3-T	—	1.17	0.01	714	906	Acceptable example
3-U	—	<u>4.24</u>	5.38	717	891	Comparative example
3-V	—	1.12	0.01	715	913	Acceptable example
3-W	—	<u>7.66</u>	5.17	715	1031	Comparative example
3-X	—	<u>0.45</u>	0.09	707	866	Comparative example
3-Y	<u>3.15</u>	—	—	707	882	Comparative example

(Note)
*1X = (V/51)/(C/12)
*2Y = (V/51 + 2 × Nb/93 + 2 × Ti/48)/(C/12)
*3Z = (2 × Nb/93 + 2 × Ti/48)/(V/51)

[0131]

TABLE 4

Steel	Chemical composition (mass %)													
No.	C	Si	Mn	P	S	Al	N	V	Nb	Ti	Cr	Mo	Cu	Ni
4-A	0.038	0.48	2.88	0.01	0.004	0.033	0.002	0.158	—	—	—	—	—	—
4-B	0.041	0.77	2.51	0.01	0.001	0.035	0.002	0.056	0.171	—	—	0.31	—	—
4-C	0.040	0.76	2.49	0.01	0.007	0.034	0.002	0.068	0.120	0.125	0.09	—	—	—
4-D	0.038	0.72	2.54	0.01	0.009	0.033	0.002	0.085	—	0.058	—	—	0.08	0.07
4-E	0.049	0.74	2.53	0.01	0.006	0.036	0.002	0.039	0.075	0.005	—	0.31	—	—
4-F	0.039	0.75	2.55	0.01	0.007	0.035	0.002	0.183	0.191	0.260	—	—	—	—
4-G	0.046	0.73	2.57	0.01	0.007	0.038	0.002	0.011	0.013	0.015	—	—	—	—
4-H	0.039	0.93	2.95	0.05	0.004	0.039	0.002	0.016	0.003	0.108	—	0.27	—	—
4-I	0.041	0.80	2.80	0.05	0.009	0.033	0.002	0.138	0.042	0.065	—	—	—	—
4-J	0.047	0.92	2.78	0.04	0.006	0.034	0.002	0.025	0.175	0.143	0.15	—	—	—
4-K	0.043	0.84	2.76	0.04	0.007	0.034	0.002	0.007	—	—		—	—	—
4-L	0.038	0.93	2.75	0.04	0.006	0.035	0.002	<u>0.553</u>	—	—		—	—	—
4-M	0.042	0.80	2.65	0.02	0.003	0.031	0.002	0.096	0.0005	—	—	—	—	—
4-N	0.041	0.81	2.68	0.02	0.003	0.030	0.002	0.029	0.065	—	—	—	—	—
4-O	0.043	0.78	2.67	0.03	0.003	0.029	0.002	0.087	0.295	—	—	—	—	—
4-P	0.041	0.77	2.69	0.02	0.002	0.030	0.002	0.079	<u>0.521</u>	—	—	—	—	—
4-Q	0.039	0.76	2.74	0.03	0.003	0.031	0.002	0.105	—	0.0005	—	—	—	—
4-R	0.043	0.79	2.74	0.02	0.004	0.033	0.002	0.035	—	0.042	—	—	—	—
4-S	0.040	0.80	2.75	0.03	0.002	0.032	0.002	0.087	—	0.182	—	—	—	—
4-T	0.038	0.81	2.77	0.02	0.003	0.032	0.002	0.089	—	0.290	—	—	—	—

TABLE 4-continued

Steel No.	X^{*1}	Y^{*2}	Z^{*3}	Transformation point (° C.) A_{c1}	Transformation point (° C.) A_{c3}	Remarks
4-A	0.98	—	—	706	842	Acceptable example
4-B	—	1.40	3.35	719	865	Acceptable example
4-C	—	2.74	5.84	720	905	Acceptable example
4-D	—	1.29	1.45	716	876	Acceptable example
4-E	—	0.63	2.38	717	859	Acceptable example
4-F	—	5.70	4.16	718	971	Comparative example
4-G	—	0.29	4.19	717	851	Comparative example
4-H	—	1.50	14.55	717	922	Acceptable example
4-I	—	1.85	1.33	716	909	Acceptable example
4-J	—	2.61	19.83	723	923	Acceptable example
4-K	0.04	—	—	718	867	Comparative example
4-L	3.42	—	—	721	924	Comparative example
4-M	—	0.54	0.01	719	865	Acceptable example
4-N	—	0.58	2.46	719	865	Acceptable example
4-O	—	2.25	3.72	719	865	Acceptable example
4-P	—	3.73	7.23	719	865	Comparative example
4-Q	—	0.64	0.01	716	876	Acceptable example
4-R	—	0.68	2.55	716	876	Acceptable example
4-S	—	2.79	4.45	716	876	Acceptable example
4-T	—	4.37	6.92	716	876	Comparative example

(Note)
$^{*1}X = (V/51)/(C/12)$
$^{*2}Y = (V/51 + 2 \times Nb/93 + 2 \times Ti/48)/(C/12)$
$^{*3}Z = (2 \times Nb/93 + 2 \times Ti/48)/(V/51)$

[0132]

TABLE 5(a)

		Cold rolling										
		Annealing temperature in continuous annealing line	Microstructure				Mechanical properties of cold rolled steel sheet					
			Ferrite phase	Second phase			Tensile properties					
Steel sheet No.	Steel No.	(° C.)	Area ratio (%)	Kind^{*1}	Area ratio of martensite (%)	Area ratio of second phase (%)	YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
1	1-A	830	92	M	8	8	330	600	31	55	1.8	Invention example
2	1-B	830	90	M	10	10	330	610	30	54	1.8	Invention example
3	1-B	980	0	P, B, M	15	100	650	720	22	90	0.9	Comparative example

TABLE 5(a)-continued

		Cold rolling										
		Annealing temperature in continuous annealing line	Microstructure				Mechanical properties of cold rolled steel sheet					
			Ferrite phase	Second phase			Tensile properties					
Steel sheet No.	Steel No.	(° C.)	Area ratio (%)	Kind*1	Area ratio of martensite (%)	Area ratio of second phase (%)	YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
4	1-B	680	100	—	0	0	450	530	29	85	0.8	Comparative Example
5	1-C	830	92	M	8	8	340	600	31	57	1.8	Invention example
6	1-D	830	90	M	10	10	330	610	30	54	1.4	Invention example
7	1-E	830	92	M	8	8	310	570	33	54	1.7	Invention example
8	1-F	830	100	—	0	0	510	600	27	85	1.8	Comparative example
9	1-G	830	93	M	7	7	330	610	31	54	0.8	Comparative example
10	1-H	850	92	M	8	8	350	630	29	56	1.9	Invention example
11	1-I	850	93	M	7	7	330	620	30	53	1.9	Invention example
12	1-J	850	92	M	8	8	330	610	33	54	1.8	Invention example
13	1-K	830	92	M	8	8	245	450	38	54	1.9	Invention example
14	1-L	830	93	M	7	7	330	605	30	55	1.8	Invention example

(Note)
*1F is abbreviation of ferrite phase, M is abbreviation of matensite phase, P is abbreviation of perlite phase and B is abbreviation of beinite phase.

[0133]

TABLE 5(b)

		Cold rolling										
		Annealing temperature in continuous annealing line	Microstructure				Mechanical properties of cold rolled steel sheet					
			Ferrite phase	Second phase			Tensile properties					
Steel sheet No.	Steel No.	(° C.)	Area ratio (%)	Kind*1	Area ratio of martensite (%)	Area ratio of second phase (%)	YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
15	1-M	830	92	M	8	8	340	620	30	55	1.7	Invention example
16	1-N	830	93	M	7	7	320	600	31	53	1.7	Invention example
17	1-O	830	92	M, B	6	8	340	625	29	54	1.8	Invention example
18	1-P	830	100	—	0	0	425	520	34	82	1.9	Comparative Example
19	1-Q	830	65	M	35	35	395	670	29	59	0.8	Comparative example
20	1-R	850	69	M	31	31	370	620	30	60	0.8	Comparative example
21	1-S	850	100	—	0	0	495	615	30	80	1.7	Comparative example
22	1-T	850	92	M	8	8	355	575	32	62	1.7	Invention example
23	1-U	850	100	—	0	0	470	580	31	81	1.8	Comparative example
24	1-V	830	91	M	9	9	350	570	32	61	1.7	Invention example

TABLE 5(b)-continued

Steel sheet No.	Steel No.	Cold rolling: Annealing temperature in continuous annealing line (° C.)	Microstructure: Ferrite phase Area ratio (%)	Microstructure: Second phase Kind*[1]	Microstructure: Second phase Area ratio of martensite (%)	Microstructure: Second phase Area ratio of second phase (%)	Mechanical properties of cold rolled steel sheet: Tensile properties YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
25	1-W	850	100	—	0	0	480	595	31	81	1.8	Comparative example
26	1-X	830	72	M	28	28	350	560	31	63	0.8	Comparative example
27	1-Y	830	100	—	0	0	475	590	30	81	1.7	Comparative example

(Note)
*[1]F is abbreviation of ferrite phase, M is abbreviation of matensite phase, P is abbreviation of perlite phase and B is abbreviation of beinite phase.

[0134]

TABLE 6(a)

Steel sheet No.	Steel No.	Cold rolling: Annealing temperature in continuous annealing line (° C.)	Microstructure: Ferrite phase Area ratio (%)	Microstructure: Second phase Kind*[1]	Microstructure: Second phase Area ratio of martensite (%)	Microstructure: Second phase Area ratio of second phase (%)	Mechanical properties of cold rolled steel sheet: Tensile properties YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
28	2-A	780	90	M	10	10	560	825	19	68	1.1	Invention example
29	2-B	780	87	M	13	13	550	810	19	68	1.3	Invention example
30	2-B	950	0	P, B, M	19	100	740	860	16	86	0.7	Comparative example
31	2-B	680	100	—	0	0	625	770	22	81	0.8	Comparative Example
32	2-C	750	88	M	12	12	540	805	20	67	1.3	Invention example
33	2-D	760	88	M	12	12	545	810	19	67	1.2	Invention example
34	2-E	770	87	M	13	13	550	820	20	67	1.3	Invention example
35	2-F	780	100	—	0	0	660	830	19	80	1.4	Comparative example
36	2-G	780	69	M	31	31	540	820	20	66	0.7	Comparative example
37	2-H	760	81	M	19	19	620	930	15	67	1.3	Invention example
38	2-I	780	83	M	17	17	590	860	17	69	1.1	Invention example

(Note)
*[1]F is abbreviation of ferrite phase, M is abbreviation of matensite phase, P is abbreviation of perlite phase and B is abbreviation of beinite phase.

[0135]

TABLE 6(b)

		Cold rolling										
		Annealing	Microstructure				Mechanical properties of					
		temperature in	Ferrite	Second phase			cold rolled steel sheet					
Steel		continuous	phase		Area ratio of	Area ratio	Tensile properties					
sheet No.	Steel No.	annealing line (° C.)	Area ratio (%)	Kind*1	martensite (%)	of second phase (%)	YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
39	2-J	780	87	M	13	13	445	660	27	67	1.4	Invention example
40	2-K	760	68	M	32	32	570	850	18	67	0.8	Comparative example
41	2-L	780	100	—	0	0	690	835	19	83	1.3	Comparative example
42	2-M	780	85	M	15	15	525	805	20	65	1.1	Invention example
43	2-N	760	88	M	12	12	530	800	20	66	1.3	Invention example
44	2-O	780	90	M	10	10	525	790	21	66	1.3	Invention example
45	2-P	780	100	—	0	0	650	795	21	82	1.3	Comparative example
46	2-Q	760	87	M	13	13	540	810	19	67	1.1	Invention example
47	2-R	760	88	M	12	12	545	815	15	67	1.3	Invention example
48	2-S	780	90	M	10	10	540	810	19	67	1.3	Invention example
49	2-T	780	100	—	0	0	665	785	20	85	1.4	Comparative example

(Note)
*1F is abbreviation of ferrite phase, M is abbreviation of matensite phase, P is abbreviation of perlite phase and B is abbreviation of beinite phase.

[0136]

TABLE 7(a)

		Galvanizing										
		Annealing	Microstructure				Mechanical properties of					
		temperature in	Ferrite	Second phase			galvanized steel sheet					
Steel		continuous	phase		Area ratio of	Area ratio	Tensile properties					
sheet No.	Steel No.	annealing line (° C.)	Area ratio (%)	Kind*1	martensite (%)	of second phase (%)	YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
50	3-A	830	92	M	8	8	330	610	31	54	1.7	Invention example
51	3-B	830	90	M	10	10	330	620	30	53	1.7	Invention example
52	3-B	830	92	M	8	8	350	630	30	56	1.6	Invention example
53	3-B	980	0	P, B, M	12	100	660	720	22	92	0.9	Comparative Example
54	3-B	680	100	—	0	0	460	540	28	85	0.8	Comparative example
55	3-C	830	90	M	10	10	340	610	31	56	1.7	Invention example
56	3-D	830	92	M	8	8	340	620	30	55	1.4	Invention example
57	3-E	830	94	M	6	6	320	580	32	55	1.6	Invention example
58	3-F	830	100	—	0	0	510	600	27	85	1.7	Comparative example

TABLE 7(a)-continued

Steel sheet No.	Steel No.	Galvanizing: Annealing temperature in continuous annealing line (° C.)	Microstructure: Ferrite phase Area ratio (%)	Second phase: Kind*1	Second phase: Area ratio of martensite (%)	Second phase: Area ratio of second phase (%)	Mechanical properties of galvanized steel sheet, Tensile properties: YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
59	3-G	830	92	M	8	8	330	610	30	54	0.8	Comparative example
60	3-H	850	93	M	7	7	340	630	30	54	1.8	Invention example
61	3-I	850	92	M	8	8	340	620	31	55	1.8	Invention example
62	3-J	850	92	M	8	8	320	610	31	52	1.7	Invention example
63	3-K	830	92	M, B	6	8	330	610	30	54	1.6	Invention example
64	3-L	830	92	M	8	8	248	450	37	55	1.7	Invention example
65	3-M	830	93	M	7	7	340	620	30	55	1.6	Invention example

(Note)
*1F is abbreviation of ferrite phase, M is abbreviation of matensite phase, P is abbreviation of perlite phase and B is abbreviation of beinite phase.

[0137]

TABLE 7(b)

Steel sheet No.	Steel No.	Galvanizing: Annealing temperature in continuous annealing line (° C.)	Microstructure: Ferrite phase Area ratio (%)	Second phase: Kind*1	Second phase: Area ratio of martensite (%)	Second phase: Area ratio of second phase (%)	Mechanical properties of galvanized steel sheet, Tensile properties: YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
66	3-N	830	92	M	8	8	320	600	31	53	1.6	Invention example
67	3-O	830	93	M	7	7	340	625	29	54	1.7	Invention example
68	3-H	830	92	M	8	8	340	620	30	55	1.8	Invention example
69	3-K	830	93	M	7	7	320	600	31	53	1.6	Invention example
70	3-M	830	92	M	8	8	320	610	31	52	1.6	Invention example
71	3-P	830	100	—	<u>0</u>	0	420	510	34	82	1.8	Comparative example
72	3-Q	830	66	M	34	34	390	670	27	58	0.8	Comparative example
73	3-R	850	68	M	32	32	385	615	30	63	0.8	Comparative example
74	3-S	850	100	—	<u>0</u>	0	500	605	31	83	1.6	Comparative example
75	3-T	850	91	M	9	9	350	580	31	60	1.7	Invention example
76	3-U	850	100	—	<u>0</u>	0	480	575	32	83	1.6	Comparative example
77	3-V	830	91	M	9	9	340	580	31	59	1.7	Invention example
78	3-W	850	100	—	<u>0</u>	0	490	600	30	82	1.7	Comparative example

TABLE 7(b)-continued

Steel sheet No.	Steel No.	Galvanizing: Annealing temperature in continuous annealing line (° C.)	Microstructure: Ferrite phase Area ratio (%)	Second phase: Kind*[1]	Second phase: Area ratio of martensite (%)	Second phase: Area ratio of second phase (%)	Mechanical properties of galvanized steel sheet: Tensile properties YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
79	3-X	830	70	M	30	30	340	565	32	60	0.8	Comparative example
80	3-Y	830	100	—	0	0	490	600	30	82	1.7	Comparative example

(Note)
*[1]F is abbreviation of ferrite phase, M is abbreviation of matensite phase, P is abbreviation of perlite phase and B is abbreviation of beinite phase.

[**0138**]

TABLE 8(a)

Steel sheet No.	Steel No.	Galvanizing: Annealing temperature in continuous annealing line (° C.)	Microstructure: Ferrite phase Area ratio (%)	Second phase: Kind*[1]	Second phase: Area ratio of martensite (%)	Second phase: Area ratio of second phase (%)	Mechanical properties of galvanized steel sheet: Tensile properties YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
81	4-A	780	91	M	9	9	560	815	19	69	1.1	Invention example
82	4-B	780	89	M	11	11	555	805	19	69	1.4	Invention example
83	4-B	950	0	P,B,M	21	100	735	850	16	86	0.8	Comparative example
84	4-B	680	100	—	0	0	620	760	22	82	0.8	Comparative Example
85	4-C	4	89	M	11	11	545	800	20	68	1.3	Invention example
86	4-D	760	88	M	12	12	550	805	19	68	1.4	Invention example
87	4-E	770	90	M	10	10	550	810	20	68	1.3	Invention example
88	4-F	780	100	—	0	0	675	815	19	83	1.5	Comparative example
89	4-G	780	92	M	8	8	550	810	20	68	0.8	Comparative example
90	4-H	760	83	M	17	17	635	935	15	68	1.3	Invention example
91	4-I	780	85	M	15	15	590	855	17	69	1.1	Invention example

(Note)
*[1]F is abbreviation of ferrite phase, M is abbreviation of matensite phase, P is abbreviation of perlite phase and B is abbreviation of beinite phase.

[0139]

TABLE 8(b)

		Galvanizing										
		Annealing temperature in continuous annealing line	Microstructure				Mechanical properties of galvanized steel sheet					
			Ferrite phase	Second phase			Tensile properties					
Steel sheet No.	Steel No.	(° C.)	Area ratio (%)	Kind*1	Area ratio of martensite (%)	Area ratio of second phase (%)	YS (MPa)	TS (MPa)	El (%)	YR (%)	r-value	Remarks
92	4-J	780	85	M	15	15	440	665	25	68	1.4	Invention example
93	4-K	760	67	M	33	33	560	860	18	65	0.8	Comparative example
94	4-L	780	100	—	0	0	695	840	19	83	1.4	Comparative example
95	4-M	780	86	M	14	14	510	810	20	63	1.1	Invention example
96	4-N	760	89	M	11	11	525	800	20	66	1.3	Invention example
97	4-O	780	89	M	11	11	525	795	20	66	1.3	Invention example
98	4-P	780	100	—	0	0	660	805	20	82	1.4	Comparative example
99	4-Q	760	87	M	13	13	525	810	19	65	1.1	Invention example
100	4-R	760	86	M	14	14	530	810	19	65	1.2	Invention example
101	4-S	780	89	M	11	11	540	820	18	66	1.3	Invention example
102	4-T	780	100	—	0	0	660	790	20	84	1.3	Comparative example

(Note)
*1F is abbreviation of ferrite phase, M is abbreviation of matensite phase, P is abbreviation of perlite phase and B is abbreviation of beinite phase.

[0140] As seen from the results shown in Tables 5 and 6, the cold rolled steel sheets in all invention examples have a low yield stress (YS), a high elongation (E1) and a low yield ratio (YR) and further indicate a high r-value and are excellent in the deep drawability, and have a tensile strength (TS) of not less than 440 MPa. On the contrary, in the comparative examples being outside the range of the invention, the yield stress (YS) is high, the elongation (E1) is low, or the r-value is low. Particularly, the somewhat lowering of the r-value accompanied with the high-strengthening is observed in the high-strength steel sheets of TS: not less than 780 MPa shown in Table 6, for example, the steel sheet No. 28 produced by using the steel No. 2-A containing V and no Nb and Ti and the steel sheet No. 38 produced by using the steel No. 2-I containing V, Nb and Ti and satisfying a relationship of $0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48) \leqq 3\times C/12$ but satisfying a relationship of $(2\times Nb/93+2\times Ti/48)/(V/51)<0.5$. On the other hand, the r-value is improved in the steel sheet Nos. 29, 32, 33 and 34 produced by using the steel Nos. 2-B, 2-C, 2-D and 2-E containing V, Nb and Ti and satisfying both relationships of $0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48) \leqq 3\times C/12$ and $1.5 \leqq (2\times Nb/93+2\times Ti/48)/(V/51) \leqq 15$.

[0141] And also, the results obtained with respect to the galvanized steel sheets are shown in Tables 7 and 8. Even in these galvanized steel sheets, the results similar to those of the above cold rolled steel sheets are obtained.

[0142] In the steel sheet according to the invention, excellent properties are obtained even by the production process conducting the galvanization.

[0143] Industrial Applicability

[0144] The invention develops an industrially remarkable effect that the high-strength cold rolled steel sheet and galvanized steel sheet having an excellent deep drawability can be produced stably. When the cold rolled steel sheet and the galvanized steel sheet according to the invention are applied to vehicle parts, there are effects that the press forming is easy and they can sufficiently contribute to reduce the weight of the vehicle body.

1. A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability, characterized in that the steel sheet has a composition comprising C: 0.01-0.08 mass %, Si: not more than 2.0 mass %, Mn: not more than 3.0 mass %, P: not more than 0.10 mass %, S: not more than 0.02 mass %, Al: 0.005-0.20 mass %, N: not more than 0.02 mass % and V: 0.01-0.5 mass %, provided that V and C satisfy a relationship of $0.5\times C/12 \leqq V/51 \leqq 3\times C/12$, and the remainder being Fe and inevitable impurities, and has a microstructure consisting of a ferrite phase as a primary phase and a secondary phase including martensite phase at an area ratio of not less than 1% to a whole of the microstructure.

2. A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability, characterized in that

the steel sheet has a composition comprising C: 0.01-0.08 mass %, Si: not more than 2.0 mass %, Mn: not more than 3.0 mass %, P: not more than 0.10 mass %, S: not more than 0.02 mass %, Al: 0.005-0.20 mass %, N: not more than 0.02 mass % and V: 0.01-0.5 mass % and further comprising not more than 0.3 mass % in total of one or two of Nb: more than 0 mass % but not more than 0.3 mass % and Ti: more than 0 mass % but not more than 0.3 mass %, provided that V, Nb, Ti and C satisfy a relationship of $0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48) \leqq 3\times C/12$, and the remainder being Fe and inevitable impurities, and has a microstructure consisting of a ferrite phase as a primary phase and a secondary phase including martensite phase at an area ratio of not less than 1% to a whole of the microstructure.

3. A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to claim 2, wherein the steel sheet comprises not more than 0.3 mass % in total of one or two of Nb: 0.001-0.3 mass % and Ti: 0.001-0.3 mass %.

4. A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to claim 2, wherein the steel sheet comprises C: 0.03-0.08 mass %, Si: 0.1-2.0 mass %, Mn: 1.0-3.0 mass %, P: not more than 0.05 mass % and S: not more than 0.01 mass %, provided that V, Nb and Ti satisfy a relationship of $1.5 \leqq (2\times Nb/93+2\times Ti/48)/(V/51) \leqq 15$.

5. A high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to any one of claims **1-4**, wherein the steel sheet further comprises one or two of the following A-group and B-group:

A-group: not more than 2.0 mass % in total of one or two of Cr and Mo;

B-group: not more than 2.0 mass % in total of one or two of Cu and Ni.

6. A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability, which comprises hot rolling a steel slab having a composition comprising C: 0.01-0.08 mass %, Si: not more than 2.0 mass %, Mn: not more than 3.0 mass %, P: not more than 0.10 mass %, S: not more than 0.02 mass %, Al: 0.005-0.20 mass %, N: not more than 0.02 mass % and V: 0.01-0.5 mass %, provided that V and C satisfy a relationship of $0.5\times C/12 \leqq V/51 \leqq 3\times C/12$, and the remainder being Fe and inevitable impurities, pickling, cold rolling and then subjecting to a continuous annealing at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point.

7. A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability, which comprises hot rolling a steel slab having a composition comprising C: 0.01-0.08 mass %, Si: not more than 2.0 mass %, Mn: not more than 3.0 mass %, P: not more than 0.10 mass %, S: not more than 0.02 mass %, Al: 0.005-0.20 mass %, N: not more than 0.02 mass % and V: 0.01-0.5 mass % and further comprising not more than 0.3 mass % in total of one or two of Nb: more than 0 mass % but not more than 0.3 mass % and Ti: more than 0 mass % but not more than 0.3 mass %, provided that V, Nb, Ti and C satisfy a relationship of $0.5\times C/12 \leqq (V/51+2\times Nb/93+2\times Ti/48) \leqq 3\times C/12$, and the remainder being Fe and inevitable impurities, pickling, cold rolling and then subjecting to a continuous annealing at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point.

8. A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to claim 7, wherein the steel slab comprises not more than 0.3 mass % in total of one or two of Nb: 0.001-0.3 mass % and Ti: 0.001-0.3 mass %.

9. A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to claim 7, wherein the steel slab comprises C: 0.03-0.08 mass %, Si: 0.1-2.0 mass %, Mn: 1.0-3.0 mass %, P: not more than 0.05 mass % and S: not more than 0.01 mass %, provided that V, Nb and Ti satisfy a relationship of $1.5 \leqq (2\times Nb/93+2\times Ti/48)/(V/51) \leqq 15$.

10. A method of producing a high-strength dual-phase cold rolled steel sheet having an excellent deep drawability according to any one of claims **6-9**, wherein the steel slab further comprises one or two of the following A-group and B-group: A-group: not more than 2.0 mass % in total of one or two of Cr and Mo; B-group: not more than 2.0 mass % in total of one or two of Cu and Ni.

11. A high-strength dual-phase galvanized steel sheet having an excellent deep drawability comprising a galvanized coating on the steel sheet as claimed in claim 1.

12. A high-strength dual-phase galvanized steel sheet having an excellent deep drawability comprising a galvanized coating on the steel sheet as claimed in claim 2.

13. A high-strength dual-phase galvanized steel sheet having an excellent deep drawability comprising a galvanized coating on the steel sheet as claimed in claim 3.

14. A high-strength dual-phase galvanized steel sheet having an excellent deep drawability comprising a galvanized coating on the steel sheet as claimed in claim 4.

15. A high-strength dual-phase galvanized steel sheet having an excellent deep drawability comprising a galvanized coating on the steel sheet as claimed in claim 5.

16. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability, characterized by subjecting to a galvanization after the continuous annealing at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point in the method claimed in claim 6.

17. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability according to claim 16, characterized by further comprising a continuous annealing step between the cold rolling step and the continuous annealing step at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point.

18. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability, characterized by subjecting to a galvanization after the continuous annealing at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point in the method claimed in claim 7.

19. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability according to claim 18, characterized by further comprising a continuous annealing step between the cold rolling step and the continuous annealing step at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point.

20. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability, characterized by subjecting to a galvanization after the continuous annealing at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point in the method claimed in claim 8.

21. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability according to claim 20, characterized by further comprising a continuous annealing step between the cold rolling step and the continuous annealing step at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point.

22. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability, characterized by subjecting to a galvanization after the continuous annealing at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point in the method claimed in claim 9.

23. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability according to claim 22, characterized by further comprising a continuous annealing step between the cold rolling step and the continuous annealing step at a temperature range from a A_{C1} transformation point to a A_{C3} transformation point.

24. A method of producing a high-strength dual-phase galvanized steel sheet having an excellent deep drawability according to any one of claims **16-23**, wherein the steel slab further comprises one or two of the following A-group and B-group:

A-group: not more than 2.0 mass % in total of one or two of Cr and Mo;

B-group: not more than 2.0 mass % in total of one or two of Cu and Ni.

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