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(54) FERRITIC STAINLESS STEEL AND METHOD FOR PRODUCING SAME

(57) Provided are a ferritic stainless steel having sufficient corrosion resistance and formability (high elongation, large average r-value, and small $|\Delta r|$) and excellent surface properties with less seam defects, and a method for producing the ferritic stainless steel.

A ferritic stainless steel of the present invention contains, in terms of % by mass, C: 0.005% to 0.05%, Si: 0.02% to 0.50%, Mn: 0.05% to 1.0%, P: 0.04% or less, S: 0.01% or less, Cr: 15.5% to 18.0%, Al: 0.001% to 0.10%, N: 0.01% to 0.06%, V: 0.01% to 0.25%, Ti: 0.001% to 0.020%, Nb: 0.001% to 0.030%, and the balance being Fe and unavoidable impurities, in which $V/(Ti + Nb) \geq 2.0$ is satisfied.

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Description

[Technical Field]

[0001] The present invention relates to a ferritic stainless steel that has sufficient corrosion resistance and formability and excellent surface properties free of seam defects caused by hot rolling or annealing, and also to a method for producing the ferritic stainless steel.

[Background Art]

[0002] Ferritic stainless steel, which is inexpensive and highly corrosion resistant, is used in a wide variety of applications including building material, transportation equipment, home electric appliances, kitchen instruments, automobile parts, etc., and the range of applications has seen further expansion in recent years. To be suitable for these applications, ferritic stainless steel is required to have not only corrosion resistance but also sufficient formability allowing the steel to be worked into desired shapes (in other words, the elongation needs to be large (hereinafter having sufficiently high elongation may be referred to as having ductility), the average Lankford value (hereinafter may be referred to as an "average r-value" needs to be high, and the absolute value of the r-value in-plane anisotropy (hereinafter may be referred to as $|\Delta r|$) needs to be small). Having excellent surface properties is also required if the applications require aesthetically appealing surfaces.

[0003] In this respect, Patent Literature 1 discloses a ferritic stainless steel having excellent formability and ridging resistance, the ferritic stainless steel containing, in terms of % by mass, C: 0.02% to 0.06%, Si: 1.0% or less, Mn: 1.0% or less, P: 0.05% or less, S: 0.01% or less, Al: 0.005% or less, Ti: 0.005% or less, Cr: 11% to 30%, and Ni: 0.7% or less, and satisfying $0.06 \leq (C + N) \leq 0.12$, $1 \leq N/C$, and $1.5 \times 10^{-3} \leq (V \times N) \leq 1.5 \times 10^{-2}$ (C, N, and V respectively represent the contents of the respective elements in terms of % by mass). However, Patent Literature 1 is completely silent about anisotropy. Moreover, box annealing (for example, annealing at 860°C for 8 hours) is necessary after hot rolling. Since the box annealing process requires about a week to finish if the annealing and cooling steps are also counted, low productivity arises as a problem.

[0004] Patent Literature 2 discloses a ferritic stainless steel having excellent workability and surface properties, obtained by hot rolling a steel containing, in terms of % by mass, C: 0.01% to 0.10%, Si: 0.05% to 0.50%, Mn: 0.05% to 1.00%, Ni: 0.01% to 0.50%, Cr: 10% to 20%, Mo: 0.005% to 0.50%, Cu: 0.01% to 0.50%, V: 0.001% to 0.50%, Ti: 0.001% to 0.50%, Al: 0.01% to 0.20%, Nb: 0.001% to 0.50%, N: 0.005% to 0.050%, and B: 0.00010% to 0.00500%, annealing the hot rolled sheet in a box furnace or a continuous furnace of an annealing and pickling line (AP line) in a ferrite single-phase temperature region, and performing cold rolling and finish annealing. However, when a box furnace is used, there is a problem of low productivity as with Patent Literature 1 described above. Although Patent Literature 2 makes no mention about elongation, annealing a hot rolled sheet in a continuous annealing furnace in a ferrite single-phase temperature region results in insufficient crystallization due to low annealing temperature, and the elongation is decreased compared to when box annealing is performed in a ferrite single-phase temperature region. Moreover, in general, when ferritic stainless steel such as one described in Patent Literature 2 is casted or hot-rolled, crystal grain groups (colonies) that have similar crystal orientations occur and a problem of a large $|\Delta r|$ arises.

[Citation List]

[Patent Literature]

[0005]

[PTL 1] Japanese Patent No. 3584881 (Re-publication of PCT International Publication No. WO00/60134)
[PTL 2] Japanese Patent No. 3581801 (Japanese Unexamined Patent Application Publication No. 2001-3134)

[Summary of Invention]

[Technical Problem]

[0006] The present invention addresses the issues described above and aims to provide a ferritic stainless steel that has sufficient corrosion resistance and formability and excellent surface properties free of seam defects caused by hot rolling or annealing, and a method for producing the ferritic stainless steel.

[0007] For the purposes of the present invention, sufficient corrosion resistance means that when a steel sheet is polish-finished with #600 Emery paper, has its end surface portions sealed, and subjected to 3 cycles of a salt spray cycle test (each cycle including salt spray (35°C, 5 mass% NaCl, spraying: 2 hours) → drying (60°C, relative humidity: 40%, 4 hours) → wetting (50°C, relative humidity ≥ 95%, 2 hours)) prescribed in JIS H 8502, the rust area fraction (= rust area/total steel sheet area × 100 [%]) of the steel sheet surface is 25% or less.

[0008] Sufficient formability means that a test specimen taken in a direction 90° with respect to the rolling direction exhibits that an elongation after fracture is 25% or more in a tensile test conducted according to JIS Z 2241, that the average r-value calculated from formula (1) below is 0.65 or more under a strain of 15% in a tensile test conducted according to JIS Z 2241, and that the absolute value ($|\Delta r|$) of the r-value in-plane anisotropy (hereinafter referred to as Δr) calculated from formula (2) below is 0.30 or less:

$$\text{Average r-value} = (r_L + 2 \times r_D + r_C)/4 \qquad (1)$$

$$\Delta r = (r_L - 2 \times r_D + r_C)/2 \qquad (2)$$

where r_L represents an r-value observed in a tensile test conducted in a direction parallel to the rolling direction, r_D represents an r-value observed in a tensile test conducted in a direction 45° with respect to the rolling direction, and r_C represents an r-value observed in a tensile test conducted in a direction 90° with respect to the rolling direction.

[Solution to Problem]

[0009] Studies have been made to address the issues and it has been found that a ferritic stainless steel having sufficient corrosion resistance and formability is obtained by annealing a ferritic stainless steel sheet having an appropriate composition in a ferrite-austenite dual-phase temperature region before cold-rolling a hot rolled steel sheet. Occurrence of seam defects on the steel sheet surface can be suppressed by further controlling V, Ti, and Nb contents within the above-described appropriate steel composition range so as not to cause precipitation of coarse Cr carbonitrides during hot rolling. It has been found that not only corrosion resistance and formability but also surface properties will be improved as a result.

[0010] The present invention has been made on the basis of the findings described above and can be summarized as below:

[1] A ferritic stainless steel comprising, in terms of % by mass, C: 0.005% to 0.05%, Si: 0.02% to 0.50%, Mn: 0.05% to 1.0%, P: 0.04% or less, S: 0.01% or less, Cr: 15.5% to 18.0%, Al: 0.001% to 0.10%, N: 0.01% to 0.06%, V: 0.01% to 0.25%, Ti: 0.001% to 0.020%, Nb: 0.001% to 0.030%, and the balance being Fe and unavoidable impurities, wherein V/(Ti + Nb) ≥ 2.0 is satisfied.

[2] A ferritic stainless steel comprising, in terms of % by mass, C: 0.01% to 0.05%, Si: 0.02% to 0.50%, Mn: 0.2% to 1.0%, P: 0.04% or less, S: 0.01% or less, Cr: 16.0% to 18.0%, Al: 0.001% to 0.10%, N: 0.01% to 0.06%, V: 0.01% to 0.25%, Ti: 0.001% to 0.015%, Nb: 0.001% to 0.025%, and the balance being Fe and unavoidable impurities, wherein V/(Ti + Nb) ≥ 2.0 is satisfied.

[3] The ferritic stainless steel according to [1] or [2], further comprising, in terms of % by mass, at least one selected from Cu: 0.1% to 1.0%, Ni: 0.1% to 1.0%, Mo: 0.1% to 0.5%, and Co: 0.01% to 0.5%.

[4] The ferritic stainless steel according to any one of [1] to [3], further comprising, in terms of % by mass, at least one selected from Mg: 0.0002% to 0.0050%, B: 0.0002% to 0.0050%, REM: 0.01% to 0.10%, and Ca: 0.0002% to 0.0020%.

[5] A method for producing a ferritic stainless steel, comprising hot-rolling a steel slab having the composition described in any one of [1] to [4], annealing the resulting hot rolled sheet by holding the hot rolled sheet in a temperature range of 880°C to 1000°C for 5 seconds to 15 minutes to obtain a hot-rolled and annealed sheet, cold-rolling the hot-rolled and annealed sheet, and annealing the resulting cold-rolled sheet by holding the cold-rolled sheet in a temperature range of 800°C to 950°C for 5 seconds to 5 minutes.

[0011] In this specification, % indicating the steel composition means % by mass.

[Advantageous Effects of Invention]

[0012] According to the present invention, a ferritic stainless steel having sufficient corrosion resistance and formability (high elongation, large average r-value, and small $|\Delta r|$) and excellent surface properties with less seam defects is obtained.

[Description of Embodiments]

[0013] The present invention will now be described in detail.

[0014] A ferritic stainless steel comprises, in terms of % by mass, C: 0.005% to 0.05%, Si: 0.02% to 0.50%, Mn: 0.05% to 1.0%, P: 0.04% or less, S: 0.01% or less, Cr: 15.5% to 18.0%, Al: 0.001% to 0.10%, N: 0.01% to 0.06%, V: 0.01% to 0.25%, Ti: 0.001% to 0.020%, Nb: 0.001% to 0.030%, and the balance being Fe and unavoidable impurities, in which $V/(Ti + Nb) \geq 2.0$ is satisfied. In the present invention, balancing the components in the composition is important, and in particular, balancing V, Ti, and Nb is important. It is important that V: 0.01% to 0.25%, Ti: 0.001% to 0.020%, Nb: 0.001% to 0.030% and $V/(Ti + Nb) \geq 2.0$ be satisfied. When the components of the composition are combined as such, a ferritic stainless steel having sufficient corrosion resistance, sufficient formability, and excellent surface properties with less seam defects can be obtained.

[0015] First, the technical details of the present invention are described.

[0016] The inventors have focused on a technique of achieving desired formability by annealing a hot rolled sheet for a short period of time using a continuous annealing furnace, which is a furnace with high productivity, instead of annealing a hot rolled sheet for a long period of time such as in box annealing (batch annealing). The problem of the related art that uses continuous annealing furnaces is that since annealing is performed in a ferrite single-phase temperature region, sufficient recrystallization does not occur, sufficient elongation is not achieved, and $|\Delta r|$ is large due to colonies remaining even after cold-rolled-sheet annealing. The inventors then have devised annealing a hot rolled sheet in a ferrite-austenite dual-phase region, and then cold rolling the resulting hot-rolled and annealed sheet and annealing the resulting cold rolled sheet by a normal process so that the microstructure of the steel returns to a ferrite single phase at the final stage.

[0017] That is, when a hot rolled sheet is annealed in a ferrite-austenite dual-phase temperature region higher than the ferrite single-phase temperature region, recrystallization of a ferrite phase is promoted. As a result, ferrite crystal grains to which working strain has been introduced by hot rolling do not remain after cold-rolled-sheet annealing, and elongation after cold-rolled-sheet annealing is improved. When an austenite phase is formed from a ferrite phase by hot-rolled-sheet annealing, the austenite phase is forms having crystal orientations which is different from that of the ferrite phase before annealing, and thus colonies of the ferrite phase are effectively destroyed. As a result, in a microstructure of a cold rolled and annealed sheet after cold rolling and cold-rolled-sheet annealing, a γ-fiber texture that improves the r-value is developed, colonies are disrupted, and anisotropy of the microstructure is relaxed. Thus, excellent properties such as small $|\Delta r|$ are obtained.

[0018] When hot-rolled-sheet annealing is performed in a ferrite-austenite dual-phase region, a dual-phase structure including a ferrite phase and a martensite phase formed by transformation of an austenite phase is formed. When a hot-rolled and annealed sheet containing the martensite phase is cold rolled, the ferrite phase near the martensite phase preferentially deform due to the martensite phase being harder than the ferrite phase, and rolling strain concentrates on those ferrite phase, thereby further increasing the number of recrystallization sites during cold-rolled-sheet annealing. As a result, recrystallization during cold-rolled-sheet annealing is further promoted, and anisotropy of the microstructure after the cold-rolled-sheet annealing is further relaxed.

[0019] However, it has been found that when a hot rolled sheet made of a steel having a conventional composition is annealed in the ferrite-austenite dual-phase region, line-like defects (hereinafter may be referred to as seam defects) occur in the rolling direction after the cold-rolled-sheet annealing, and the surface properties are significantly degraded, which poses a new problem.

[0020] In order to achieve both formability and surface properties, the inventors have investigated the cause of occurrence of seam defects resulting from hot-rolled-sheet annealing in the ferrite-austenite dual-phase region. It has been found that seam defects are caused by a significantly hard martensite phase that exists in a surface layer portion of a steel sheet after hot-rolled-sheet annealing. In other words, when the significantly hard martensite phase is present in a surface layer portion of a steel sheet after the hot-rolled-sheet annealing, strains concentrate at the interfaces between the significantly hard martensite phase and the ferrite phase during the subsequent cold rolling process and cause microcracks that will form seam defects after the cold-rolled-sheet annealing. The martensite phase is formed as a result of transformation of an austenite phase, which has been formed in the hot-rolled-sheet annealing in the ferrite-austenite dual-phase region, as cooling proceeds. The hardness of the martensite grains in the microstructure has been studied. It has been found that while most part of the martensite phase has a Vickers hardness (HV) of about 300 to 400, some part of the martensite phase has shown significantly high hardness with HV exceeding 500, and that microcracks that occur in cold rolling occur at the interfaces between the ferrite phase and the significantly hard martensite phase with HV exceeding 500.

[0021] The inventors have examined what causes the significantly hard martensite phase having HV exceeding 500 to locally occur after the hot-rolled-sheet annealing, and pursued the technology to overcome this issue. As a result, it has been found that the significantly hard martensite phase is formed when coarse Cr carbonitrides are present before the hot-rolled-sheet annealing. The mechanism behind this is presumably as follows. In hot-rolled-sheet annealing,the austenite phase is formed by the dissolution of the Cr carbonitrides that have precipitated during hot rolling. If the Cr carbonitrides before the hot-rolled-sheet annealing are coarse, the amount of carbon (C) supplied to the austenite phase increases. As a result, the region in which coarse Cr carbonitrides was dissolved locally exhibits higher C concentration compared to the region in which coarse Cr carbonitrides did not dissolved. The significantly hard martensite phase is formed after the hot-rolled-sheet annealing from this austenite phase having high C concentrations.

[0022] The inventors then focused on the technology for preventing precipitation of coarse Cr carbonitrides during hot rolling. As a result they have found that precipitation of coarse Cr carbonitrides during hot rolling can be avoided when the steel composition contains V, Ti, and Nb in amounts of V: 0.01% to 0.25%, Ti: 0.001% to 0.020%, and Nb: 0.001% to 0.030%, and satisfies V/(Ti + Nb) ≥ 2.0.

[0023] In other words, it has been found that when appropriate amounts of these elements are contained, complex carbonitrides (Cr, V, Ti, Nb)(C, N) containing V, Ti and Nb precipitate instead of Cr carbonitrides during hot rolling and these complex carbonitrides precipitate more finely and homogeneously than the Cr carbonitrides, thereby suppressing occurrence of coarse Cr carbonitrides.

[0024] This effect is exhibited by incorporation of an appropriate amount of V. Titanium (Ti) and Nb have higher affinity to C and N than Cr, and more easily form carbonitrides than Cr. When Ti or Nb is used alone, Ti or Nb will precipitate as Ti(C, N) or Nb(C, N) separate from Cr carbonitrides, and the effect of suppressing formation of coarse Cr carbonitrides is not obtained.

[0025] Vanadium (V) is also an element that has strong affinity to C and N. Vanadium (V) tends to form complex carbonitrides (Cr, V, Ti, Nb)(C, N) with Cr, Ti, and Nb, and Cr carbonitrides precipitate as (Cr, V, Ti, Nb)(C, N) if an appropriate amount of V is contained in addition to Ti and Nb. Since these (Cr, V, Ti, Nb)(C, N) precipitates containing V, Ti, and Nb which have a smaller diffusion rate than Cr, growth or coarsening after precipitation is governed by diffusion of V, Ti, and Nb. Thus, the precipitates are finer than conventional Cr carbonitrides and occurrence of coarse carbonitrides during hot rolling can be effectively suppressed.

[0026] It has become clear that due to these effects, the formation of the significantly hard martensite phase caused by dissolution of coarse Cr carbonitrides during hot-rolled-sheet annealing in the ferrite-austenite dual-phase region is suppressed, and occurrence of seam defects after the cold-rolled-sheet annealing is significantly reduced.

[0027] In other words, in order to achieve desired formability without degrading the surface properties by conducting hot-rolled-sheet annealing for a short period of time using a continuous annealing furnace instead of conducting hot-rolled-sheet annealing for a long period of time such as in box annealing (batch annealing), it is necessary not only to conduct hot-rolled-sheet annealing for a short period of time in the ferrite-austenite dual-phase region but also to have the steel composition contain an appropriate blend of V, Ti, and Nb.

[0028] The composition of the ferritic stainless steel according to the present invention will now be described. In the description below, % means % by mass unless otherwise noted.

C: 0.005% to 0.05%

[0029] Carbon (C) has an effect of expanding the dual-phase, which has a ferrite phase and an austenite phase, temperature region during hot-rolled-sheet annealing due to promoting formation of an austenite phase. In order to obtain this effect, the C content needs to be 0.005% or more. At a C content exceeding 0.05%, the steel sheet hardens and ductility is degraded. Moreover, the significantly hard martensite phase is formed after the hot-rolled-sheet annealing even according to the present invention, and seam defects are induced after cold-rolled-sheet annealing. Accordingly, the C content is to be within the range of 0.005% to 0.05%. The lower limit is preferably 0.01% and more preferably 0.015%. The upper limit is preferably 0.035%, more preferably 0.03%, and yet more preferably 0.025%.

Si: 0.02% to 0.50%

[0030] Silicon (Si) is an element that acts as a deoxidizing agent during steel melting. In order to obtain this effect, the Si content needs to be 0.02% or more. At a Si content exceeding 0.50%, however, the steel sheet hardens and the rolling load during hot rolling increases. Moreover, the ductility after the cold-rolled-sheet annealing is deteriorated. Accordingly, the Si content is to be in the range of 0.02% to 0.50%, preferably in the range of 0.10% to 0.35%, and more preferably in the range of 0.25% to 0.30%.

Mn: 0.05% to 1.0%

[0031] As with carbon (C), manganese (Mn) has an effect of expanding the dual-phase, which has a ferrite phase and an austenite phase, temperature region during hot-rolled-sheet annealing due to promoting formation of an austenite phase. In order to obtain this effect, the Mn content needs to be 0.05% or more. At a Mn content exceeding 1.0%, however, the amount of MnS generated increases and corrosion resistance is deteriorated. Accordingly, the Mn content is to be in the range of 0.05% to 1.0%. The lower limit is preferably 0.1% and more preferably 0.2%. The upper limit is preferably 0.8%, more preferably 0.35%, and yet more preferably 0.3%.

P: 0.04% or less

[0032] Phosphorus (P) is an element that promotes intergranular fracture by intergranular segregation, and thus the P content is preferably as low as possible. The upper limit is to be 0.04%, preferably 0.03% or less, and more preferably 0.01% or less.

S: 0.01% or less

[0033] Sulfur (S) is an element that degrades ductility, corrosion resistance, etc., by forming sulfide-based inclusions such as MnS. Its adverse effects are particularly notable when the S content exceeds 0.01%. Thus, the S content is preferably as low as possible, and the upper limit of the S content in the present invention is 0.01%, more preferably 0.007% or less, and yet more preferably 0.005% or less.

Cr: 15.5% to 18.0%

[0034] Chromium (Cr) is an element that has an effect of improving corrosion resistance by forming a passivation film on a steel sheet surface. In order to obtain this effect, the Cr content needs to be 15.5% or more. At a Cr content exceeding 18.0%, however, an austenite phase is not sufficiently formed during hot-rolled-sheet annealing and desired material properties are not obtained. Accordingly, the Cr content is to be in the range of 15.5% to 18.0%, more preferably in the range of 16.0% to 18.0%, and yet more preferably in the range of 16.0% to 17.0%.

Al: 0.001% to 0.10%

[0035] As with Si, aluminum (Al) is an element that acts as a deoxidizing agent. In order to obtain this effect, the Al content needs to be 0.001% or more. At an Al content exceeding 0.10%, however, the amount of the Al-based inclusions such as Al_2O_3 increases, and the surface properties tend to be degraded. Thus, the Al content is to be in the range of 0.001% to 0.10%, preferably in the range of 0.001% to 0.07%, more preferably in the range of 0.001% to 0.05%, and yet more preferably in the range of 0.001% to 0.03%.

N: 0.01% to 0.06%

[0036] As with C and Mn, nitrogen (N) has an effect of promoting expanding the dual-phase, which has a ferrite phase and an austenite phase, temperature region during hot-rolled-sheet annealing due to promoting formation of an austenite phase. In order to obtain this effect, the N content needs to be 0.01% or more. At a N content exceeding 0.06%, however, ductility is significantly degraded, and corrosion resistance is degraded due to accelerated precipitation of Cr nitrides. Accordingly, the N content is to be in the range of 0.01% to 0.06%, preferably in the range of 0.01% to 0.05%, and more preferably in the range of 0.02% to 0.04%.

V: 0.01% to 0.25%

[0037] Vanadium (V) is an extremely critical element in the present invention. Vanadium (V) is characterized by having higher affinity to C and N than Cr, and when $V/(Ti + Nb) \geq 2.0$ is satisfied, vanadium combined with Cr, Ti, and Nb precipitates as (Cr, V, Ti, Nb)(C, N) during hot rolling and suppresses precipitation of coarse Cr carbonitrides. Due to this effect, formation of the extremely C-rich austenite phase is suppressed during hot-rolled-sheet annealing, the significantly hard martensite phase is not formed after hot-rolled-sheet annealing, and occurrence of surface seam defects resulting from microcracks generated during cold rolling is prevented. In order to obtain this effect, the V content must be 0.01% or more. At a V content exceeding 0.25%, however, workability is degraded and the production cost rises. Accordingly, the V content is to be in the range of 0.01% to 0.25%, preferably in the range of 0.03% to 0.20%, and more preferably in the range of 0.05% to 0.15%.

Ti: 0.001% to 0.020%, Nb: 0.001% to 0.030%, $V/(Ti + Nb) \geq 2.0$

[0038] As with V, Ti and Nb are elements that have higher affinity to C and N than Cr and have an effect of suppressing precipitation of coarse Cr carbonitrides by forming (Cr, V, Ti, Nb)(C, N) with V and Cr during hot rolling if the steel contains V. In order to obtain this effect, 0.001% or more of Ti and 0.001% or more of Nb must be contained while satisfying $V/(Ti + Nb) \geq 2.0$. However, when the Ti content exceeds 0.020% or the Nb content exceeds 0.030%, Ti(C, N) and Nb(C, N) independently precipitate during hot rolling instead of (Cr, V, Ti, Nb)(C, N). Thus, the effect of suppressing formation of coarse Cr carbonitrides is not obtained, and desired surface properties cannot be obtained. Accordingly, the Ti content is to be in the range of 0.001% to 0.020%, and the Nb content is to be in the range of 0.001% to 0.030%. The Ti content is preferably in the range of 0.001% to 0.015% and more preferably in the range of 0.003% to 0.010%. The Nb content is preferably in the range of 0.001% to 0.025% and more preferably in the range of 0.005% to 0.020%. When V/(Ti + Nb) is less than 2.0, V needed to form composite carbonitrides becomes deficient, and Ti, Nb, and V each independently form carbides or nitrides; thus, formation of coarse Cr carbonitrides cannot be sufficiently suppressed. Thus, V/(Ti + Nb) is to be 2.0 or more, preferably 3.0 or more, and more preferably 4.0 or more. In contrast, if V/(Ti + Nb) exceeds 30.0, V, Ti, and Nb are not used in forming composite carbonitrides even when the V, Ti, and Nb contents are in the designated ranges, and the amount of V in a dissolved state in the matrix is increased. Thus, the steel sheet becomes hard and elongation decreases. The upper limit of V/(Ti + Nb) is preferably 30.0.

[0039] The balance is Fe and unavoidable impurities.

[0040] Although the advantageous effects of the present invention are achieved by the composition described above, the following elements may be contained in order to further improve manufacturability and material properties.

At least one selected from Cu: 0.1% to 1.0%, Ni: 0.1% to 1.0%, Mo: 0.1% to 0.5%, and Co: 0.01% to 0.5%

[0041] Copper (Cu) and nickel (Ni) are both an element that improves corrosion resistance and are preferably contained if particularly high corrosion resistance is required. Moreover, Cu and Ni have an effect of promoting expanding the dual-phase, which has a ferrite phase and an austenite phase, temperature region during hot-rolled-sheet annealing due to promoting formation of an austenite phase. These effects are notable when each element is contained in an amount of 0.1% or more. However, a Cu content exceeding 1.0% is not preferable since hot workability is degraded. If Cu is to be contained, the Cu content is to be 1.0% or less, preferably in the range of 0.2% to 0.8%, and more preferably in the range of 0.3% to 0.5%. A Ni content exceeding 1.0% is not preferable since workability is degraded. If Ni is to be contained, the Ni content is to be 1.0% or less, preferably in the range of 0.1% to 0.6%, and more preferably in the range of 0.1% to 0.3%.

[0042] Molybdenum (Mo) is an element that improves corrosion resistance and it is effective to use Mo when a particularly high corrosion resistance is required. This effect becomes notable at a Mo content of 0.1% or more. However, a Mo content exceeding 0.5% is not preferable since formation of an austenite phase during hot-rolled-sheet annealing is insufficient and desired material properties are not obtained. Thus, if Mo is to be contained, the Mo content is to be 0.1% to 0.5% or less and preferably in the range of 0.1% to 0.3%.

[0043] Cobalt (Co) is an element that improves toughness. This effect is obtained at a Co content of 0.01% or more. At a Co content exceeding 0.5%, workability is degraded. Thus, if Co is to be contained, the Co content is to be 0.5% or less and preferably in the range of 0.01% to 0.2%.

At least one selected from Mg: 0.0002% to 0.0050%, B: 0.0002% to 0.0050%, REM: 0.01% to 0.10%, and Ca: 0.0002% to 0.0020%

Mg: 0.0002% to 0.0050%

[0044] Magnesium (Mg) is an element that has an effect of improving hot workability. In order to obtain this effect, the Mg content needs to be 0.0002% or more. At a Mg content exceeding 0.0050%, however, the surface quality is degraded. Thus, if Mg is to be contained, the Mg content is to be in the range of 0.0002% to 0.0050%, preferably in the range of 0.0005% to 0.0035%, and more preferably in the range of 0.0005% to 0.0020%.

B: 0.0002% to 0.0050%

[0045] Boron (B) is an element effective for preventing secondary working embrittlement. In order to obtain this effect, the B content needs to be 0.0002% or more. At a B content exceeding 0.0050%, however, hot workability is degraded. Thus, if B is to be contained, the B content is to be in the range of 0.0002% to 0.0050%, preferably in the range of 0.0005% to 0.0035%, and more preferably in the range of 0.0005% to 0.0020%.

REM: 0.01% to 0.10%

[0046] A rare earth metal (REM) is an element that improves oxidation resistance and particularly has an effect of improving corrosion resistance of a weld zone by suppressing formation of oxide films in the weld zone. In order to obtain this effect, the REM content needs to be 0.01% or more. At a REM content exceeding 0.10%, however, manufacturability such as a pickling property during cold-rolled-sheet annealing is degraded. Moreover, since REM is an expensive element, excessive incorporation thereof increases the manufacturing cost, which is not preferable. If REM is to be contained, the REM content is to be in the range of 0.01% to 0.10%.

Ca: 0.0002% to 0.0020%

[0047] Calcium (Ca) is a component effective for preventing nozzle clogging caused by crystallization of Ti-based inclusions that easily occur during continuous casting. In order to obtain this effect, the Ca content needs to be 0.0002% or more. At a Ca content exceeding 0.0020%, however, CaS is generated and corrosion resistance is degraded. Accordingly, if Ca is to be contained, the Ca content is to be in the range of 0.0002% to 0.0020%, preferably in the range of 0.0005% to 0.0015%, and more preferably in the range of 0.0005% to 0.0010%.

[0048] Next, a method for producing a ferritic stainless steel according to the present invention is described. A ferritic stainless steel according to the present invention is obtained by hot-rolling a steel slab having the composition described above, annealing the resulting hot rolled sheet by holding the hot rolled sheet in a temperature range of 880°C to 1000°C for 5 seconds to 15 minutes to obtain a hot-rolled and annealed sheet, cold-rolling the hot-rolled and annealed sheet, and annealing the resulting cold-rolled sheet by holding the cold-rolled sheet in a temperature range of 800°C to 950°C for 5 seconds to 5 minutes.

[0049] First, a molten steel having the composition described above is melted by using a known method such as one of a steel converter, an electric furnace, a vacuum melting furnace, or the like and formed into a steel material (slab) by a continuous casting method or an ingoting-blooming method. The slab is heated at 1100°C to 1250°C for 1 to 24 hours and hot rolled, or directly hot rolled as casted so as to form a hot rolled sheet.

[0050] Next, hot rolling is performed. In coiling, the coiling temperature is preferably 500°C or more and 850°C or less. A coiling temperature less than 500°C is not preferable since recrystallization after coiling is insufficient and ductility after cold-rolled-sheet annealing may sometimes be degraded. Grain size may increase when the hot rolled sheet is coiled at a temperature exceeding 850°C and surface deterioration may occur during press-forming. Accordingly, the coiling temperature is preferably in the range of 500°C to 850°C.

[0051] Subsequently, hot-rolled-sheet annealing of holding the hot rolled sheet at a temperature of 880°C to 1000°C, which is a ferrite-austenite dual-phase region temperature, is performed for 5 seconds to 15 minutes.

[0052] The hot-rolled-sheet annealing is an important step of the present invention for obtaining desired surface properties and formability. At a hot-rolled-sheet annealing temperature less than 880°C, sufficient recrystallization does not occur and the effects of the present invention brought by the dual-phase annealing may not be obtained since this temperature is in the ferrite single phase region. However, if the annealing temperature exceeds 1000°C, dissolution of carbides is accelerated, the C concentration of the austenite phase is increasingly increased, and the significantly hard martensite phase is formed after the hot-rolled-sheet annealing. Thus, desired surface properties are not obtained.

[0053] When the hot-rolled-sheet annealing temperature exceeds 1000°C, the amount of the austenite phase is decreased. Thus, the amount of the martensite phase formed after the hot-rolled-sheet annealing is decreased. As a result, an effect of relaxing the anisotropy of the microstructure caused by concentration of rolling strain onto the ferrite phase near the martensite phase when the microstructure containing the ferrite phase and the martensite phase is cold rolled cannot be sufficiently obtained, and a desired $|\Delta r|$ cannot be obtained.

[0054] If the annealing time is less than 5 seconds, formation of the austenite phase and recrystallization of the ferrite phase are not sufficient even by annealing at the designated temperature, and thus desired formability is not obtained. If the annealing time is more than 15 minutes, some of (Cr, V, Ti, Nb)(C, N) dissolves, promoting an increase in C concentration in the austenite phase. Thus, due to the same mechanism described above, the desired surface properties are not obtained.

[0055] If the annealing time is longer than 15 minutes, an excessive increase in the C content in the martensite phase formed by transformation of the austenite phase after hot-rolled-sheet annealing occurs due to the mechanism described above. The martensite phase decomposes into carbides and the ferrite phase during cold-rolled-sheet annealing; however, if the C concentration is excessively large, the martensite phase transform to the ferrite phase containing a large quantity of carbides. As a result, after the cold-rolled-sheet annealing, the microstructure turns into a mixed grain microstructure constituted by ferrite grains that have less intragranular and intergranular carbides and ferrite grains that have excessive intragranular and intergranular carbides. In such a microstructure, there is a difference in hardness between grains with less carbides and grains with more carbides; hence, deformation strain concentrates at the interface between two grains, large voids starting from intergranular carbides are easily formed, and ductility is degraded.

[0056] Thus, the hot-rolled-sheet annealing involves holding a temperature of 880°C to 1000°C for 5 seconds to 15 minutes, preferably holding a temperature of 900°C to 1000°C for 15 seconds to 15 minutes, and more preferably holding a temperature of 900°C to 1000°C for 15 seconds to 3 minutes.

[0057] Next, cold rolling and cold-rolled-sheet annealing are performed. If needed, pickling is performed to obtain a product.

[0058] Cold rolling is preferably performed at a reduction of 50% or more from the viewpoint of formability and shape correction. In the present invention, cold rolling and annealing may be repeated twice or more, and a stainless steel foil having a thickness of 200 μm or less may be formed by cold rolling.

[0059] In annealing the cold rolled sheet, a temperature of 800°C to 950°C is held for 5 seconds to 5 minutes in order to obtain excellent formability.

[0060] The cold-rolled-sheet annealing is an important step for converting a ferrite-martensite dual-phase microstructure formed by hot-rolled-sheet annealing into a ferrite single-phase microstructure. At a cold-rolled-sheet annealing temperature less than 800°C, recrystallization does not occur sufficiently, and desired ductility and average r-value cannot be obtained. If the cold-rolled-sheet annealing temperature exceeds 950°C and if the steel composition is one such this temperature is in the ferrite-austenite dual-phase temperature region, the steel sheet becomes hard due to formation of the martensite phase after cold-rolled-sheet annealing and desired ductility cannot obtained. Even if the steel composition is one such this temperature is in the ferrite single phase temperature region, excessive coarsening of crystal grains degrades glossiness of the steel sheet, which is not preferable from the viewpoint of the surface quality. If the annealing time is less than 5 seconds, recrystallization of ferrite phase does not sufficiently occur even at the designated annealing temperature; thus, desired ductility and average r-value cannot be obtained. An annealing time exceeding 5 minutes is not preferred from the viewpoint of surface quality since crystal grains coarsen excessively and glossiness of the steel sheet is degraded. Thus, the cold-rolled-sheet annealing is to be carried out by holding a temperature in the range of 800°C to 950°C for 5 seconds to 5 minutes, and preferably in the range of 850°C to 900°C for 15 seconds to 3 minutes. If higher glossiness is desirable, bright annealing (BA) may be conducted.

[0061] Grinding, polishing, or the like process may be performed to further improve surface properties.

[EXAMPLE 1]

[0062] The present invention will now be described in more detail through examples.

[0063] A stainless steel having a chemical composition shown in Table 1 was melted in a 50 kg small-scale vacuum melting furnace. The resulting steel ingot was heated at 1150°C for 1 hour and then hot-rolled into a hot rolled sheet having a thickness of 3.5 mm. The hot rolled sheet was subjected to hot-rolled-sheet annealing under conditions described in Table 2, and the surface of the resulting annealed sheet was descaled by a shot blast treatment and pickling. Pickling involved immersing the sheet in a 20 mass% sulfuric acid solution at a temperature of 80°C for 120 seconds and then immersing the sheet in a 15 mass% nitric acid-3 mass% hydrofluoric acid mixed solution at a temperature of 55°C for 60 seconds. The pickled sheet was cold-rolled to a thickness of 0.7 mm, and the cold-rolled-sheet annealing was performed under the conditions described in Table 2. The resulting annealed sheet was subjected to a descaling treatment that involved electrolytic pickling in a 18 mass% aqueous Na_2SO_4 solution having a solution temperature of 80°C under a condition of 25 C/dm^2, and electrolytic pickling in a 10 mass% aqueous HNO_3 solution having a solution temperature of 50°C under a condition of 30 C/dm^2. As a result, a cold-rolled, pickled, and annealed sheet was obtained.

[0064] The cold-rolled, pickled, and annealed sheet thus obtained was evaluated for the following properties.

(1) Surface property evaluation

[0065] The number of seam defects which are 5 mm or longer and present per square meter of a steel sheet was counted after cold-rolled-sheet annealing. Samples with 5 or less seam defects per square meter of the steel sheet observed on the surface of the cold-rolled and annealed sheet were rated Pass (P), and samples with more than 5 seam defects were rated Fail (F).

(2) Evaluation of ductility

[0066] A JIS 13B tensile test specimen was sampled from the cold-rolled, pickled, and annealed sheet in a direction 90° with respect to the rolling direction, and a tensile test was conducted in accordance with JIS Z 2241 to measure elongation after fracture. Samples with elongation after fracture of 25% or more were rated Pass (P) and samples elongation after fracture of less than 25% were rated Fail (F).

(3) Evaluation of average r-value and |Δrl

[0067] JIS 13B tensile test specimens were taken from the cold-rolled, pickled, and annealed sheet in a direction (L direction) parallel to the rolling direction, a direction (D direction) 45° with respect to the rolling direction, and a direction (C direction) 90° with respect to the rolling direction. The tensile test in accordance with JIS Z 2241 was conducted up to 15% strain and interrupted. The r-values of the respective directions were measured and the average r-value ($= (r_L + 2r_D + r_C)/4$) and the absolute value ($|\Delta r|$) of the r value in-plane anisotropy ($\Delta r = (r_L - 2r_D + r_C)/2$) were calculated. Here, r_L, r_D, and r_C are respectively r-values in the L direction, the D direction, and the C direction. Samples with an average r-value of 0.65 or more were rated Pass (P) and samples with an average r-value less than 0.65 were rated Fail (F). Samples with $|\Delta r|$ of 0.30 or less were rated Pass (P), and samples with $|\Delta r|$ exceeding 0.30 were rated Fail (F).

(4) Evaluation of corrosion resistance

[0068] A 60 mm × 100 mm test specimen was sampled from the cold-rolled, pickled, and annealed sheet, the surface thereof was polish-finished with #600 Emery paper, and end surfaces were sealed to prepare a test piece to be used in a salt spray cycle test prescribed in JIS H 8502. The salt spray cycle test was performed 3 cycles, each cycle including salt spray (35°C, 5% NaCl, spraying: 2 hours) → drying (60°C, relative humidity: 40%, 4 hours) → wetting (50°C, relative humidity ≥ 95%, 2 hours). The surface of the test piece after 3 cycles of the salt spray cycle test was photographed, the rust area of the test piece surface was measured by image processing, and the rust area fraction ((rust area in test piece/total area of test piece) × 100 [%]) was calculated as a ratio with respect to the total area of the test piece. Samples with a rust area fraction of 10% or less were rated Pass with particularly excellent corrosion resistance (PP), samples with a rust area fraction of more than 10% but not more than 25% were rated Pass (P), and samples with a rust area fraction more than 25% were rated Fail (F).

[0069] The evaluation results are shown in Table 2 together with the hot-rolled-sheet annealing conditions and cold-rolled-sheet annealing conditions.

[Table 1]

(mass%)

Steel Type	C	Si	Mn	P	S	Cr	Al	N	V	Ti	Nb	Cu	Ni	Mo	Mg	B	Co	REM	Ca	V/(Ti+Nb)	Note
A	0.03	0.23	0.9	0.03	0.004	18.0	0.002	0.03	0.02	0.002	0.001	-	0.1	-	-	-	-	-		6.67	*Ex.
B	0.04	0.14	0.8	0.01	0.003	17.2	0.038	0.04	0.11	0.008	0.001	-	0.2	-	-	-	-	-		12.2	*Ex.
C	0.02	0.14	0.8	0.03	0.003	16.4	0.001	0.04	0.08	0.003	0.001	-	0.2	-	-	-	-	-		20	*Ex.
D	0.03	0.40	0.4	0.02	0.002	17.3	0.027	0.03	0.20	0.010	0.014	-	0.3	-	-	-	-	-		8.3	*Ex.
E	0.04	0.26	0.7	0.03	0.009	16.4	0.003	0.02	0.07	0.011	0.009	-	0.1	-	-	-	-	-		3.5	*Ex.
F	0.03	0.08	0.2	0.01	0.008	16.6	0.079	0.05	0.12	0.001	0.006	-	0.2	-	-	-	-	-		17.1	*Ex.
G	0.01	0.33	0.8	0.02	0.005	16.3	0.067	0.02	0.12	0.009	0.001	-	0.3	-	-	-	-	-		12.0	*Ex.
H	0.03	0.21	0.8	0.03	0.003	17.5	0.081	0.02	0.14	0.014	0.025	-	0.5	-	-	-	-	-		3.6	*Ex.
I	0.04	0.47	0.5	0.03	0.002	17.3	0.035	0.05	0.01	0.001	0.003	-	0.1	-	-	-	-	-		2.5	*Ex.
J	0.02	0.33	0.4	0.03	0.004	16.1	0.001	0.04	0.23	0.015	0.022	-	0.4	-	-	-	-	-		6.2	*Ex.
K	0.03	0.14	0.4	0.03	0.007	16.5	0.043	0.03	0.08	0.004	0.007	-	-	-	-	-	-	-		7.3	*Ex.
L	0.04	0.18	0.9	0.03	0.005	16.6	0.013	0.04	0.06	0.005	0.010	0.3	-	-	-	-	-	-		4.0	*Ex.
M	0.04	0.21	0.7	0.04	0.003	16.4	0.003	0.03	0.06	0.009	0.009	-	0.7	-	-	-	-	-		3.3	*Ex.
N	0.03	0.22	0.8	0.02	0.005	16.8	0.009	0.02	0.07	0.007	0.011	-	-	0.3	-	-	-	-		3.9	*Ex.
O	0.03	0.11	0.8	0.03	0.004	16.2	0.003	0.03	0.06	0.006	0.001	-	-	-	0.0017	-	-	-		8.6	*Ex.
P	0.01	0.13	0.9	0.02	0.005	16.6	0.007	0.02	0.05	0.010	0.001	-	-	-	-	0.0022	-	-		4.5	*Ex.
Q	0.05	0.24	0.8	0.03	0.007	16.2	0.010	0.03	0.06	0.013	0.016	-	-	-	-	-	0.14	-		2.1	*Ex.
R	0.04	0.26	0.8	0.01	0.005	16.4	0.009	0.02	0.05	0.010	0.013	-	-	-	-	-	-	0.02		2.2	*Ex.
BA	0.023	0.31	0.24	0.02	0.004	16.3	0.005	0.03	0.04	0.018	0.002	-	-	-	-	-	-	-	0.0014	2.0	*Ex.
BB	0.021	0.29	0.21	0.03	0.004	16.3	0.006	0.04	0.04	0.009	0.011	-	-	-	-	-	-	-	-	2.0	*Ex.
BC	0.008	0.28	0.63	0.04	0.006	16.2	0.007	0.03	0.04	0.006	0.013	-	0.1	-	-	-	-	-	-	2.1	*Ex.
BD	0.024	0.24	0.09	0.03	0.005	16.3	0.005	0.04	0.05	0.008	0.012	-	-	-	-	-	-	-	-	2.5	*Ex.
BE	0.021	0.25	0.77	0.02	0.004	15.6	0.004	0.03	0.04	0.007	0.010	-	-	-	-	-	-	-	-	2.4	*Ex.
BF	0.023	0.34	0.78	0.03	0.006	16.4	0.008	0.05	0.11	0.018	0.017	-	0.3	-	-	-	-	-	-	3.1	*Ex.
BG	0.022	0.31	0.78	0.02	0.004	16.5	0.006	0.05	0.12	0.013	0.029	-	-	-	-	-	-	-	-	2.9	*Ex.
BH	0.022	0.25	0.82	0.04	0.006	16.6	0.007	0.04	0.04	0.007	0.012	0.1	-	-	-	-	-	-	-	2.1	*Ex.
BI	0.023	0.19	0.76	0.03	0.004	16.4	0.005	0.03	0.05	0.007	0.014	-	0.2	-	-	0.0008	-	-	-	2.4	*Ex.
BJ	0.025	0.28	0.79	0.03	0.005	16.3	0.008	0.04	0.05	0.009	0.013	-	-	0.1	-	-	-	-	-	2.3	*Ex.
BK	0.021	0.24	0.74	0.04	0.004	16.7	0.008	0.04	0.04	0.011	0.008	-	-	-	-	-	-	-	-	2.1	*Ex.
BL	0.021	0.27	0.25	0.03	0.005	16.5	0.006	0.03	0.05	0.010	0.009	-	0.1	-	-	-	-	-	-	2.6	*Ex.
BM	0.018	0.31	0.26	0.03	0.007	15.7	0.003	0.04	0.11	0.010	0.002	0.2	0.1	0.3						9.2	*Ex.
BN	0.034	0.28	0.26	0.03	0.004	16.1	0.003	0.04	0.05	0.002	0.012									3.6	*Ex.
BO	0.024	0.26	0.24	0.03	0.005	16.2	0.005	0.03	0.06	0.008	0.019					0.0017		0.02		2.2	*Ex.
BP	0.022	0.35	0.25	0.02	0.004	16.1	0.003	0.04	0.04	0.014	0.003	0.1							0.0012	2.4	*Ex.
BQ	0.020	0.26	0.27	0.03	0.005	16.6	0.027	0.03	0.03	0.005	0.002									4.6	*Ex.
BR	0.024	0.32	0.35	0.02	0.002	16.3	0.004	0.02	0.07	0.002	0.006									8.8	*Ex.
BS	0.032	0.30	0.29	0.03	0.003	17.8	0.001	0.04	0.06	0.007	0.005				0.0015		0.15			5.0	*Ex.
BT	0.016	0.27	0.24	0.03	0.006	16.0	0.002	0.02	0.06	0.003	0.002									12.0	*Ex.
BU	0.018	0.28	0.34	0.03	0.009	16.3	0.002	0.03	0.14	0.008	0.006									10.0	*Ex.
S	0.04	0.18	1.0	0.02	0.006	17.4	0.048	0.04	0.003	0.012	0.021	-	0.5	-	-	-	-	-		0.1	**CEx
T	0.02	0.16	0.9	0.04	0.003	17.8	0.042	0.05	0.28	0.014	0.023	-	0.1	-	-	-	-	-		7.6	**CEx

U	0.02	0.16	0.8	0.03	0.005	17.1	0.017	0.03	0.18	0.023	0.034	-	0.1	-	-	-	-	-		3.2	**CEx
V	0.04	0.24	0.5	0.04	0.006	13.9	0.077	0.05	0.05	0.011	0.009	-	0.3	-	-	-	-	-		2.5	**CEx
W	0.03	0.25	0.6	0.04	0.008	18.4	0.051	0.05	0.04	0.004	0.008	-	-	-	-	-	-	-		3.3	**CEx
X	0.08	0.37	0.7	0.02	0.004	16.7	0.031	0.06	0.10	0.009	0.014	-	0.2	-	-	-	-	-		4.3	**CEx
Y	0.003	0.30	0.8	0.01	0.006	16.3	0.024	0.05	0.08	0.013	0.017	-	0.2	-	-	-	-	-		2.7	**CEx
Z	0.02	0.18	0.7	0.02	0.008	16.1	0.002	0.03	0.02	0.004	0.008	-	-	-	-	-	-	-		1.7	**CEx
AA	0.03	0.24	0.8	0.03	0.011	16.5	0.004	0.04	0.03	0.011	0.014	-	0.1	-	-	-	-	-		1.2	**CEx
AB	0.021	0.29	0.31	0.03	0.0005	16.3	0.003	0.04	0.04	0.015	0.012	-	0.1	-	-	-	-	-		1.5	**CEx

Underlines indicate items outside the scope of the invention.
*Ex.: Invention Example, **CEx.: Comparative Example

[Table 2]

No.	Steel type	Hot-rolled-sheet annealing		Cold-rolled-sheet annealing		Surface property (No. of seam defects per square meter of steel sheet)	Ductility	Average r - value	|Δr|	Corrosion resistance	Note
		Temperature (°C)	Time (sec)	Temperature (°C)	Time (sec)						
1	A	919	57	860	30	2	P	P	P	P	*Ex.
2	B	922	60	860	30	1	P	P	P	P	*Ex.
3	C	922	58	860	30	1	P	P	P	P	*Ex.
4	C	961	61	860	30	0	P	P	P	P	*Ex.
5	C	893	88	860	30	0	P	P	P	P	*Ex.
6	D	921	61	860	30	0	P	P	P	P	*Ex.
7	E	923	57	860	30	1	P	P	P	P	*Ex.
8	E	957	124	860	30	1	P	P	P	P	*Ex.
9	F	923	58	860	30	1	P	P	P	P	*Ex.
10	F	989	46	860	30	0	P	P	P	P	*Ex.
11	F	920	69	860	30	0	P	P	P	P	*Ex.
12	G	920	62	860	30	2	P	P	P	P	*Ex.
13	H	920	63	860	30	1	P	P	P	P	*Ex.
14	I	919	63	860	30	3	P	P	P	P	*Ex.
15	J	920	58	860	30	0	P	P	P	P	*Ex.
16	K	923	58	860	30	0	P	P	P	P	*Ex.
17	L	918	59	860	30	2	P	P	P	PP	*Ex.
18	M	922	61	860	30	1	P	P	P	PP	*Ex.
19	N	921	57	860	30	3	P	P	P	PP	*Ex.
20	O	921	61	860	30	2	P	P	P	P	*Ex.
21	P	920	58	860	30	3	P	P	P	P	*Ex.
22	Q	917	60	860	30	4	P	P	P	P	*Ex.
23	R	917	59	860	30	3	P	P	P	P	*Ex.

(continued)

No.	Steel type	Hot-rolled-sheet annealing		Cold-rolled-sheet annealing		Surface property (No. of seam defects per square meter of steel sheet)	Ductility	Average r - value	|Δrl	Corrosion resistance	Note
		Temperature (°C)	Time (sec)	Temperature (°C)	Time (sec)						
33	BA	922	58	860	30	1	P	P	P	P	*Ex.
34'	BB	958	59	860	30	0	P	P	P	P	*Ex.
35	BC	920	61	860	30	0	P	P	P	P	*Ex.
36	BD	917	60	860	30	1	P	P	P	P	*Ex.
37	BE	919	63	860	30	1	P	P	P	P	*Ex.
38	BF	922	57	860	30	0	P	P	P	P	*Ex.
39	BG	923	55	860	30	0	P	P	P	P	*Ex.
40	BH	920	58	900	15	1	P	P	P	P	*Ex.
41	BI	918	64	840	60	1	P	P	P	P	*Ex.
42	BJ	921	61	860	30	1	P	P	P	P	*Ex.
43	BL	924	60	860	30	1	P	P	P	P	*Ex.
44	BL	923	57	860	30	0	P	P	P	P	*Ex.
45	A	954	8	860	30	2	P	P	P	P	*Ex.
46	Q	897	118	820	120	3	P	P	P	P	*Ex.
52	BM	942	59	860	30	2	P	P	P	PP	*Ex.
53	BN	957	62	860	60	2	P	P	P	P	*Ex.
54	BO	939	60	860	60	3	P	P	P	P	*Ex.
55	BP	926	93	860	30	1	P	P	P	P	*Ex.
56	BQ	932	60	860	30	4	P	P	P	P	*Ex.
57	BR	946	63	860	30	2	P	P	P	P	*Ex.
58	BS	939	60	860	30	1	P	P	P	P	*Ex.
59	BT	937	58	860	30	0	P	P	P	P	*Ex.
60	BU	940	61	960	30	2	P	P	P	P	*Ex.

(continued)

No.	Steel type	Hot-rolled-sheet annealing		Cold-rolled-sheet annealing		Surface property (No. of seam defects per square meter of steel sheet)	Ductility	Average r - value	|Δrl	Corrosion resistance	Note
		Temperature (°C)	Time (sec)	Temperature (°C)	Time (sec)						
61	BM	982	30	900	30	0	P	P	P	PP	*Ex.
62	BA	945	7	860	30	1	P	P	P	P	*Ex.
63	BS	908	175	840	60	1	P	P	P	P	*Ex.
24	S	923	58	860	30	36	P	P	P	P	**CEx
25	T	919	57	860	30	0	F	P	P	P	**CEx
26	U	918	61	860	30	23	P	P	P	P	"CEx
27	V	917	58	860	30	2	P	P	P	F	**CEx
28	W	921	62	860	30	4	F	F	F	PP	**CEx
29	X	923	62	860	30	16	F	P	P	P	**CEx
30	Y	919	62	860	30	1	P	F		P	**CEx
31	Z	920	60	860	30	39	P	P	P	P	**CEx
32	AA	919	61	860	30	35	P	P	P	P	**CEx
47	Z	1031	60	860	30	43	P	P	F	P	**CEx
48	Z	841	64	860	30	0	F	F	F	P	**CEx
64	AB	1028	59	860	30	38	P	P	F	P	**CEx
65	AB	839	62	860	30	1	F	F	F	P	**CEx
66	AB	924	1233	860	30	54	F	P	P	P	**CEx
67	AB	921	59	780	30	38	F	F	P	P	**CEx
68	AB	922	61	960	30	40	F	P	P	P	**CEx

Underlines indicate items outside the scope of the invention.
*Ex.: Invention Example, **CEx.: Comparative Example

[0070] In Examples Nos. 1 to 23, 33 to 46, and 52 to 63 of the present invention that satisfy the ranges of the present invention, the number of seam defects observed after the cold-rolled-sheet annealing was 5 or less per square meter in all samples, which means that excellent surface properties were obtained. In addition, it is confirmed that excellent formability was obtained since the elongation after fracture was 25% or more, the average r-value was 0.65 or more, and $|\Delta r|$ was 0.30 or less. Moreover, regarding corrosion resistance, in all samples, since the rust area fraction of the test piece surface after 3 cycles of the salt spray cycle test was 25% or less, which means that excellent corrosion resistance was obtained.

[0071] In particular, Steels L, M, N, and BM (Nos. 17, 18, 19, 52, and 61) containing Cu, Ni, and Mo had a rust area fraction of 10% or less after the salt spray cycle test, and exhibited even better corrosion resistance.

[0072] In contrast, in Comparative Example No. 24 in which the V content was below the range of the present invention and $V/(Ti + Nb) \geq 2.0$ was not satisfied and in Comparative Example No. 26 in which Ti and Nb contents were beyond the ranges of the present invention, the amount of precipitates of (Cr, V, Ti, Nb)(C, N) was deficient and dissolved C and N were not sufficiently immobilized during hot-rolled-sheet annealing. As a result, significantly hard martensite phase occurred after hot-rolled-sheet annealing, and a large number of seam defects occurred after cold-rolled-sheet annealing.

[0073] In Comparative Example No. 25 in which the V content was beyond the range of the present invention, the desired average r-value and $|\Delta r|$ were obtained, but excessive incorporation of V made the steel sheet hard and the desired ductility could not be obtained.

[0074] In Comparative Example No. 27 in which the Cr content was below the range of the present invention, the desired surface properties, ductility, average r-value, and $|\Delta r|$ were obtained; however, since the Cr content was deficient, the desired corrosion resistance was not obtained.

[0075] In Comparative Example No. 28 in which the Cr content was beyond the range of the present invention, sufficient corrosion resistance was obtained, but excessive incorporation of Cr prevented formation of the austenite phase during hot-rolled-sheet annealing, and the desired ductility, average r-value, and $|\Delta r|$ could not be obtained.

[0076] In Comparative Example No. 29 in which the C content was beyond the range of the present invention, V, Ti, and Nb contents were within the range of the present invention, but C in the steel could not be sufficiently stabilized as (Cr, V, Ti, Nb)(C, N) and dissolved C remained. Thus, the significantly hard martensite phase was formed after hot-rolled-sheet annealing, and the desired surface properties were not obtained. Moreover, since the dissolved C content increased, the steel sheet strength increased significantly, and the desired ductility was not obtained.

[0077] In Comparative Example No. 30 in which the C content was below the range of the present invention, the effect of stabilizing an austenite phase by C was insufficient, a sufficient amount of the austenite phase was not formed during hot-rolled-sheet annealing in the dual phase region, and the desired average r value and $|\Delta r|$ could not be obtained.

[0078] In Comparative Example Nos. 31 and 32 in which V/(Ti + Nb) was below the range of the present invention, precipitation of (Cr, V, Ti, Nb)(C, N) during hot rolling was not sufficient, a large quantity of coarse Cr carbonitrides precipitated, and the significantly hard martensite phase was formed after hot-rolled-sheet annealing. Thus, a large number of seam defects occurred after cold-rolled-sheet annealing, and the desired surface properties were not obtained.

[0079] In Comparative Example Nos. 47 and 64, V/(Ti + Nb) was below the range of the present invention, and the hot-rolled-sheet annealing temperature was higher than the range of the present invention. Since V/(Ti + Nb) was below the range of the present invention, the increase in the C concentration of the austenite phase caused by dissolution of coarse carbides precipitated during hot rolling was promoted, the significantly hard martensite phase was formed after hot-rolled-sheet annealing and generated a large number of seam defects, and thus the desired surface properties were not obtained. Moreover, since the hot-rolled-sheet annealing temperature was higher than the range of the present invention, the amount of the austenite phase formed by annealing decreased, and the amount of the martensite phase formed after the hot-rolled-sheet annealing decreased. As a result, the microstructure anisotropy relaxing effect to be brought by the subsequent cold rolling could not be obtained, and the desired $|\Delta r|$ was not obtained.

[0080] No. 48 and No. 65 are comparative examples in which V/(Ti + Nb) was below the range of the present invention, and the hot-rolled-sheet annealing temperature was lower than the range of the present invention. Although V/(Ti + Nb) was below the range of the present invention, the hot-rolled-sheet annealing temperature was in the ferrite single-phase temperature region and the austenite phase was not formed. Thus, occurrence of seam defects resulting from formation of the significantly hard martensite phase was substantially prevented, and excellent surface properties were obtained. However since the hot-rolled-sheet annealing temperature was lower than the range of the present invention, sufficient recrystallization did not occur and the martensite phase was not formed after hot-rolled-sheet annealing. Thus, the desired ductility, average r-value, and $|\Delta r|$ were not obtained.

[0081] No. 66 is a comparative example in which V/(Ti + Nb) was below the range of the present invention and the hot-rolled-sheet annealing time was longer than the range of the present invention. Accordingly, the C concentration in the the austenite phase caused by dissolution of coarse carbides precipitated during hot rolling increased excessively, and the significantly hard martensite phase was formed after hot-rolled-sheet annealing. Thus, a large number of seam defects occurred and the desired surface properties were not obtained. Moreover, the microstructure after cold-rolled-sheet annealing was a mixed grain microstructure constituted by ferrite grains that had excessive intragranular and intergranular carbides and ferrite grains that had less grain boundaries and intergranular carbides. Thus, strain concentration occurred locally at the interfaces between the crystal grains during tensile deformation, and the desired ductility was not obtained.

[0082] No. 67 is a comparative example in which the V/(Ti + Nb) was below the range of the present invention and the cold-rolled-sheet annealing temperature was lower than the range of the present invention. Because the V/(Ti + Nb) was below the range of the present invention, a large number of seam defects occurred and the desired surface properties were not obtained. Since the cold-rolled-sheet annealing temperature was lower than the range of the present invention, recrystallization during the cold-rolled-sheet annealing was insufficient, and the deformation microstructure formed by cold rolling remained. Thus, the desired ductility and average r-value were not obtained.

[0083] No. 68 is a comparative example in which the V/(Ti + Nb) was below the range of the present invention and the cold-rolled-sheet annealing temperature was higher than the range of the present invention. Since the V/(Ti + Nb) was below the range of the present invention, a large number of seam defects occurred and the desired surface properties were not obtained. Since the cold-rolled-sheet annealing temperature was higher than the range of the present invention, annealing was conducted in the ferrite-austenite dual-phase temperature region and austenite phase occurred again. Since the austenite phase transformed into martensite phase after the cold-rolled-sheet annealing, the steel sheet hardened significantly and the desired ductility was not obtained.

[Industrial Applicability]

[0084] The ferritic stainless steel obtained in the present invention is particularly suitable for use in press-formed products such as products formed mainly by drawing and applications that require aesthetically appealing surfaces, e.g., applications to kitchen instruments and tableware.

Claims

1. A ferritic stainless steel comprising, in terms of % by mass, C: 0.005% to 0.05%, Si: 0.02% to 0.50%, Mn: 0.05% to 1.0%, P: 0.04% or less, S: 0.01% or less, Cr: 15.5% to 18.0%, Al: 0.001% to 0.10%, N: 0.01% to 0.06%, V: 0.01% to 0.25%, Ti: 0.001% to 0.020%, Nb: 0.001% to 0.030%, and the balance being Fe and unavoidable impurities, wherein V/(Ti + Nb) ≥ 2.0 is satisfied.

2. A ferritic stainless steel comprising, in terms of % by mass, C: 0.01% to 0.05%, Si: 0.02% to 0.50%, Mn: 0.2% to 1.0%, P: 0.04% or less, S: 0.01% or less, Cr: 16.0% to 18.0%, Al: 0.001% to 0.10%, N: 0.01% to 0.06%, V: 0.01% to 0.25%, Ti: 0.001% to 0.015%, Nb: 0.001% to 0.025%, and the balance being Fe and unavoidable impurities, wherein V/(Ti + Nb) ≥ 2.0 is satisfied.

3. The ferritic stainless steel according to Claim 1 or 2, further comprising, in terms of % by mass, at least one selected from Cu: 0.1% to 1.0%, Ni: 0.1% to 1.0%, Mo: 0.1% to 0.5%, and Co: 0.01% to 0.5%.

4. The ferritic stainless steel according to any one of Claims 1 to 3, further comprising, in terms of % by mass, at least one selected from Mg: 0.0002% to 0.0050%, B: 0.0002% to 0.0050%, REM: 0.01% to 0.10%, and Ca: 0.0002% to 0.0020%.

5. A method for producing a ferritic stainless steel, comprising hot-rolling a steel slab having the composition described in any one of Claims 1 to 4, annealing the resulting hot rolled sheet by holding the hot rolled sheet in a temperature range of 880°C to 1000°C for 5 seconds to 15 minutes to obtain a hot-rolled and annealed sheet, cold-rolling the hot-rolled and annealed sheet, and annealing the resulting cold-rolled sheet by holding the cold-rolled sheet in a temperature range of 800°C to 950°C for 5 seconds to 5 minutes.

INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2015/000032

A. CLASSIFICATION OF SUBJECT MATTER
C22C38/00(2006.01)i, *C22C38/28*(2006.01)i, *C22C38/54*(2006.01)i, *C21D9/48* (2006.01)n

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

B. FIELDS SEARCHED

Minimum documentation searched (classification system followed by classification symbols)
C22C38/00, C22C38/28, C22C38/54, C21D9/48

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched

Jitsuyo Shinan Koho	1922-1996	Jitsuyo Shinan Toroku Koho	1996-2015
Kokai Jitsuyo Shinan Koho	1971-2015	Toroku Jitsuyo Shinan Koho	1994-2015

Electronic data base consulted during the international search (name of data base and, where practicable, search terms used)

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X Y	JP 2001-003143 A (Nippon Steel Corp.), 09 January 2001 (09.01.2001), tables 1 to 2 (Family: none)	1-4 5
X Y	JP 6-158233 A (Nippon Steel Corp.), 07 June 1994 (07.06.1994), table 1 (Family: none)	1-4 5

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

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INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2015/000032

C (Continuation). DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X Y	JP 2002-542040 A (USINOR), 10 December 2002 (10.12.2002), table 1 & US 6622779 B1 & EP 1187691 A1 & WO 2000/064613 A1 & DE 60000938 D & DE 60000938 T & FR 2792561 A & FR 2792561 A1 & AU 3661900 A & BR 9881 A & CZ 20013777 A & SK 14612001 A & AT 228905 T & DK 1187691 T & AU 757307 B & PT 1187691 E & ES 2187456 T & PL 351310 A & TW 520306 B & CN 1347352 A & ZA 200108667 A & RU 2242325 C & SI 1187691 T & TR 200103013 T	1-4 5
Y	WO 2013/136736 A1 (JFE Steel Corp.), 19 September 2013 (19.09.2013), paragraphs [0049], [0064], [0070] to [0071] & JP 5376099 B & TW 201348463 A	5
Y	JP 2009-68034 A (JFE Steel Corp.), 02 April 2009 (02.04.2009), paragraphs [0026] to [0027]; table 2 (Family: none)	5

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- WO 0060134 A **[0005]**
- JP 3581801 B **[0005]**
- JP 2001003134 A **[0005]**