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(54) HOT-ROLLED STEEL SHEET, FORGED STEEL PART AND PRODUCTION METHODS THEREFOR

(57) A hot rolled steel sheet having a chemical composition consisting of, in mass %, C: 0.020-0.070%, Si: 0.05-1.70%, Mn: 0.60-2.50%, Al: 0.010-1.000%, N: > 0-0.0030%, P ≤ 0.050%, S ≤ 0.005%, Ti: 0.015-0.170%, Nb: 0-0.100%, V: 0-0.300%, Cu: 0-2.00%, Ni: 0-2.00%, Cr: 0-2.00%, Mo: 0-1.00%, B: 0-0.0100%, Mg: 0-0.0100%, Ca: 0-0.0100%, REM: 0-0.1000%, Zr: 0-1.000%, Co: 0-1.000%, Zn: 0-1.000%, W: 0-1.000%, Sn: 0-0.050%, the balance: Fe and impurities, a metal microstructure includes, in area %, ferrite: 5-70%, bainite: 30-95%, retained $\gamma \leq 2\%$, martensite ≤ 2%, pearlite ≤ 1%, ferrite + bainite ≥ 95%, a number density of the precipitates in ferrite grains is 1.0×10^{16}-50.0×10^{16}/cm^3, an average circle-equivalent diameter of the TiN precipitates in the steel sheet is 1.0-10.0 μm, an average of minimum distances between adjacent TiN precipitates is 10.0 μm or more, and a standard deviation of nano hardness is 1.00 GPa or less.

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Description

TECHNICAL FIELD

[0001] The present invention relates to a hot rolled steel sheet and a steel forged part, and a production method therefor.

BACKGROUND ART

[0002] High strength and high press workability are required for steel sheets used in body structures of automobiles in view of safety improvement and weight reduction. In order to meet the requirements, there has conventionally been proposed a high-strength steel sheet that has excellent hole expandability (a high burring property). For example, as a steel sheet that has excellent hole expandability (λ value), a steel sheet made, as a primary phase, of ferrite that is precipitation strengthened with fine precipitates such as Ti and Nb and a production method therefor have been reported.

[0003] Patent Document 1 discloses a hot-rolled steel sheet that is strong and has excellent stretch flangeability. Patent Document 2 discloses a high-formability high-tensile-strength hot-rolled steel sheet that has excellent material uniformity. Patent Document 3 discloses a high tensile strength hot-rolled steel sheet that has excellent elongation and stretch flangeability.

LIST OF PRIOR ART DOCUMENTS

PATENT DOCUMENT

[0004]

Patent Document 1: JP2002-105595A

Patent Document 2: JP2002-322540A

Patent Document 3: JP2002-322541A

SUMMARY OF INVENTION

TECHNICAL PROBLEM

[0005] With an increase in complexity of body structures of automobiles as well as complexity of shapes of parts, working on steel sheets for automobiles has been practiced by a mixed combination of new working elements with conventional press working elements, as with the case of sheet metal forging, instead of solely by conventional press working elements. Such conventional press working elements include, for example, deep drawing, hole expansion, bulging, bending, and ironing.

[0006] In recent press working typified by sheet metal forging, working elements for forging such as upsetting and thickening have been added to the conventional press working elements by further dispersing a pressing load and applying a partial compressive load. In other words, the sheet metal forging is a way of press working that includes mixed working elements including forging-specific working elements, in addition to conventional working elements for press working steel sheets.

[0007] In such sheet metal forging, a steel sheet is deformed into a shaped part with the steel sheet retaining an original sheet thickness or being thinned (reduced in thickness) by the conventional press working, while the sheet thickness is increased in a forged portion by a partially applied compressive force. In this way, efficient deformation can be achieved such that a sheet thickness of the steel sheet intended for a functionally necessary portion can be attained, and strength of the part can be secured.

[0008] However, Patent Documents 1 to 3 fail to describe a working including mixed working elements typified by the sheet metal forging. Further, coiling conditions for producing a hot-rolled steel sheet described in Patent Document 1 are highly strict and are not practical. Still further, the hot-rolled steel sheets described in Patent Documents 2 and 3 contain Mo, which is an expensive alloying element, in an amount of 0.07% or more, and thus have a problem of high production costs.

[0009] It has been known that a high burring steel exhibits good formability during conventional press working. However, it has been found that the sheet metal forging, which is a forming method including forging elements in addition to the conventional press working, may in some cases cause cracks in the steel sheet even at a low working ratio and end in rupture.

[0010] Specifically, in the conventional press working, press cracking appears at a point where sheet thickness necking (a reduced sheet thickness of the steel sheet) occurs. It has also been found that even in a working that is not associated with sheet thickness necking, such as sheet metal forging, cracks may be generated in the material, which may end in rupture and products may not be obtained in some cases.

[0011] Little is known about what characteristics of steel sheet govern the limit of crack generation in the sheet metal forging and how it can be improved. Accordingly, there has been a need for a high burring steel that is not prone to rupture even during sheet metal forging while conventional features of a high burring steel such as deep drawing workability, hole expandability, and bulging workability are still effective.

[0012] An object of the present invention, which has been made to solve the above problem, is to provide a hot rolled steel sheet with excellent sheet forgeability, which maintains basic features as a high burring steel and also makes it possible to improve cracking limit of a forged portion by a partially applied compressive force.

SOLUTION TO PROBLEM

[0013] The present invention has been made to solve the above problem, and the gist thereof a hot rolled steel sheet and a steel forged part, and a production method therefor, as described below.

[0014]

(1) A hot rolled steel sheet having a chemical composition consisting of, in mass %,
C: 0.020 to 0.070%,
Si: 0.05 to 1.70%,
Mn: 0.60 to 2.50%,
Al: 0.010 to 1.000%,
N: more than 0% to 0.0030% or less,
P: 0.050% or less,
S: 0.005% or less,
Ti: 0.015 to 0.170%,
Nb: 0 to 0.100%,
V: 0 to 0.300%,
Cu: 0 to 2.00%,
Ni: 0 to 2.00%,
Cr: 0 to 2.00%,
Mo: 0 to 1.00%,
B: 0 to 0.0100%,
Mg: 0 to 0.0100%,
Ca: 0 to 0.0100%,
REM: 0 to 0.1000%,
Zr: 0 to 1.000%,
Co: 0 to 1.000%,
Zn: 0 to 1.000%,
W: 0 to 1.000%,
Sn: 0 to 0.050%, and,
the balance: Fe and impurities, wherein
when a width and a thickness of the steel sheet in a cross section perpendicular to a rolling direction of the steel sheet are defined as W and t, respectively, a metal microstructure includes, in area %, at a position 1/4W or 3/4W from an end face of the steel sheet and 1/4t or 3/4t from a surface of the steel sheet,
ferrite: 5 to 70%,
bainite: 30 to 95%,
retained austenite: 2% or less,
martensite: 2% or less, and,
pearlite: 1% or less, and
a total of ferrite and bainite: 95% or more, and wherein
the ferrite contains precipitates including Ti in grains of the ferrite,
a number density of the precipitates including Ti is 1.0×10^{16} to $50.0 \times 10^{16}/cm^3$,
the steel sheet includes TiN precipitates,
an average circle-equivalent diameter of the TiN precipitates is 1.0 to 10.0 μm,
an average of minimum distances between adjacent TiN precipitates is 10.0 μm or more, and
a standard deviation of nano hardness is 1.00 GPa or less.

(2) The hot rolled steel sheet according to the above (1), in which an average circle-equivalent diameter of the precipitates including Ti is 1.00 to 3.00 nm.
(3) The hot rolled steel sheet according to the above (1) or (2), in which
a tensile strength is 780 MPa or more,
a product of a uniform elongation and a tensile strength is 7000 MPa·% or more, and
a product of a hole expansion ratio and a tensile strength is 50000 MPa·% or more.
(4) A method of producing the hot rolled steel sheet according to any of the above (1) to (3), including:

subjecting a slab having the chemical composition according to the above (1) to a heating process, a continuous hot rolling process, a first cooling process, a second cooling process, and a coiling process, in this order, wherein
in the heating process, the slab is heated to a temperature of SRTmin°C or more to 1260°C or less, the SRTmin being represented by a formula (i) below,
the continuous hot rolling process includes rough rolling and multi-stand finish rolling of three stands or more,
an end temperature of the rough rolling is 1100°C or more,
a cumulative strain of rolling at final three stands in the multi-stand finish rolling is 0.01 to 0.10,
a rolling end temperature of the multi-stand finish rolling is at a temperature of Ar3 + 30°C or more, the Ar_3 being determined by a formula (ii) below,
in the first cooling process, cooling is started after 1.00 to 5.00 seconds after completion of the multi-stand finish rolling, the cooling is continued at an average cooling rate of 10°C/sec or more from the rolling end temperature down to a temperature range of 650 to 750°C, and thereafter the sheet is held in air for 1 to 10 seconds,
in the second cooling process, after the sheet is held in air, cooling is conducted at an average cooling rate of 10°C/sec or more from a temperature range of 600 to 740°C, and
in the coiling process, the sheet is coiled at a coiling temperature of 450 to 650°C:

$$SRTmin = 7000/\{2.75 - \log(Ti \times C)\} - 273 \ldots (i)$$

$$Ar_3 = 970 - 325 \times C + 33 \times Si + 287 \times P + 40 \times Al - 92 \times (Mn + Mo + Cu) - 46 \times (Cr + Ni) \ldots (ii)$$

where a symbol of an element in the above formula represents a content (in mass %) of the element in the hot rolled steel sheet and is substituted by zero when the element is not contained.

(5) A steel forged part obtained from the hot rolled steel sheet according to any of the above (1) to (3).
(6) A method of producing a steel forged part, in which the hot rolled steel sheet according to any of the above (1) to (3) is subjected to at least forging.

ADVANTAGEOUS EFFECTS OF INVENTION

[0015] According to the present invention, a hot rolled steel sheet with excellent sheet forgeability maintaining a good hole expandability, which is a basic feature as a high burring steel, can be provided.

BRIEF DESCRIPTION OF DRAWINGS

[0016] [Figure 1] Figure 1 shows schematic drawings illustrating a simple shear test. Figure 1(a) illustrates a specimen for a simple shear test. Figure 1(b) illustrates a specimen after a simple shear test.

DESCRIPTION OF EMBODIMENTS

[0017] The present inventors conducted intensive studies in order to solve the above problem and obtained the following findings.

(a) Equivalent Plastic Strain

[0018] The sheet metal forging includes a strain range exceeding a rupture strain in a conventional tensile test (high strain range). Since the sheet metal forging is a composite working, it cannot be evaluated simply based on tensile test

and shear test data. Accordingly, the present inventors established a new way of evaluation by introducing an "equivalent plastic strain" as an indicator.

[0019] The present inventors have found that the equivalent plastic strain can be used as an indicator to mixedly evaluate a tensile stress and a tensile strain at the time of rupture when a tensile test is conducted and a shearing stress and a shearing strain at the time of rupture when a shear test is conducted.

[0020] The equivalent plastic strain is converted using a relation between a shearing stress σs and a shear plastic strain εsp in a simple shear test into a relation between a tensile stress σ and a tensile strain ε in a uniaxial tensile test, which is different in deformation mode. Assuming an isotropic hardening rule and a plastic work conjugate relationship, a constant, conversion factor (κ) can be used to make a conversion as in the formula below. The conversion factor (κ) is calculated according to a method described later, and then an equivalent plastic strain is derived.

$$\text{uniaxial tensile test tensile stress } \sigma = \text{simple shear test shearing stress } \sigma s \times \kappa$$

$$\text{uniaxial tensile test tensile strain } \varepsilon = \text{simple shear test shear plastic strain } \varepsilon sp/\kappa$$

(b) Multi-Stage Shear Test

[0021] To determine the equivalent plastic strain, it is necessary to obtain a relation between a tensile stress and a tensile strain in a tensile test and a relation between a shearing stress and a shear strain in a shear test. However, the sheet metal forging includes deformation in a high strain range. Accordingly, when test is performed at one time in a commonly used shear test device, cracks may propagate in a specimen from a portion where the specimen is held. As a result, a test of deformation may not often be completed up to the high strain range. Therefore, there is a need for a method for reproducing a working, such as sheet metal forging, in which thinning (thickness reduction and necking) of steel sheet does not occur.

[0022] The present inventors have then chosen to divide a shear test into multiple stages, machine an initiation point of a crack in a specimen generated in a portion where the specimen is held in order to prevent the crack from propagating in the specimen after the shear test of each stage, and evaluate a test result obtained by serially connecting the shear test results. Employing the test method, it is possible to obtain the shear test results up to the high strain range and to determine a relation between a shearing stress and a shearing strain up to the high strain range.

[0023] On the other hand, a conventional tensile test method can be applied to the tensile stress and the tensile strain. For example, a JIS No. 5 specimen based on JIS Z 2241 (2011) can be used.

(c) Mechanism of Crack Generation

[0024] By employing the above-described multi-stage shear test, the evaluation method with an equivalent plastic strain, and micro-structure observations of steel sheet before and after sheet metal forging, the present inventors obtained the following findings about the mechanism of crack generation.

[0025] To attain excellent hole expandability, a microstructure made, as a primary phase, of ferrite that is precipitation strengthened with fine precipitates such as Ti and Nb (precipitation strengthened ferrite) is used for a high burring steel. On the other hand, in the case in which Ti is added, coarse TiN may be precipitated (hereafter, such precipitated TiN is also referred to simply as "TiN") unless any special production method is used. This is because TiN is a highly thermally-stable compound from a thermodynamic point of view and may preferentially be crystalized or precipitated with respect to other compounds when being cast during a steel sheet production process, when heated in hot rolling, or under a high temperature condition as with the beginning of rough rolling.

[0026] TiN is hard enough to be used for cutting tools, machine parts, press tooling for plastic molding, sporting goods, coating applications for ornaments, and the like. It is known that the hardness of TiN is on the order of Hv2000 to 2300, which demonstrates that TiN is a very hard precipitate. Accordingly, under deformation in a high strain range as in the case of sheet metal forging, a void is likely to be generated at an interface between a parent phase structure and TiN due to a difference in deformability with the parent phase structure.

[0027] Due to a difference between a hard precipitate (TiN) and a reasonably soft parent phase structure (ferrite or bainite), a void (micro cavity) may be generated at an interface between the two phases. Thereafter, as strain associated with the sheet metal forging increases, the void may grow and coalesce with an adjacent void to become a crack, ending in rupture. The present inventors have then found that the crack generation can be inhibited if the void generation can be prevented and if the void can be inhibited from coalescing with an adjacent void even when the void grows. At this time, however, it is also important that intrinsic functionality as a high burring steel is left unimpaired.

[0028] The present inventors have found the followings from the findings.

(i) To limit an average diameter of TiN.

[0029] Specifically, a void may be generated at a grain boundary of TiN, which is a hard precipitate, and thus limiting an average diameter of TiN can lead to a reduction in void generation.

(ii) To limit a distance between TiN precipitates.

[0030] Specifically, a void may be generated at a grain boundary of TiN, and thus spacing the TiN precipitates apart from each other can make it difficult for voids to coalesce with each other even when the voids grow.

(iii) To reduce variation in nano hardness.

[0031] Specifically, the void generation can be reduced by reducing a difference in hardness between a hard microstructure and a soft microstructure as much as possible.

(iv) Equivalent plastic strain at the time of rupture is 0.90 (90%) or more.

[0032] It has been confirmed that when the conditions (i) to (iii) are satisfied, equivalent plastic strain at the time of rupture reaches 0.90 (90%) or more, and a certain level of workability can be secured even in a composite working such as sheet metal forging.

(d) Effective Cumulative Strain

[0033] To obtain a microstructure satisfying the above (i) to (iv), in the multi-stand finish rolling, which is conducted by continuous rolling at multiple, three stands or more (for example, 6 or 7 stands) in hot rolling, it is necessary to perform a final finish rolling such that a cumulative strain (hereafter, also referred to as "effective cumulative strain") of rolling at final three stands is 0.01 to 0.10.

[0034] The effective cumulative strain is an indicator that takes into consideration grain recovery, recrystallization, and grain growth according to temperature during rolling and rolling reduction of a steel sheet by rolling. Accordingly, a constitutive equation that represents static recovery phenomena in a time lapse after rolling is used for determining the effective cumulative strain. The static recovery of grains in a time lapse after rolling is taken into consideration because energy accumulated as strain in rolled grains may be released in the static recovery due to vanishment of thermal dislocations of grains. Further, the vanishment of thermal dislocations may be affected by rolling temperature and lapsed time after rolling. Accordingly, taking the static recovery into consideration, the present inventors introduced an indicator described, as parameters, by the temperature during rolling, the rolling reduction of a steel sheet by rolling (logarithmic strain), and the lapsed time after rolling, and defined it as "effective cumulative strain".

[0035] By limiting the effective cumulative strain in this way, an intended microstructure can be obtained and the variation in nano hardness can be reduced. Accordingly, the void generation is inhibited at an interface between a hard microstructure and a soft microstructure, and thus sheet metal forging does not cause cracks. Therefore, a hot rolled steel sheet with excellent sheet forgeability can be obtained.

[0036] The present invention has been made based on the above-described findings. Description will now be made as to each requirement of the present invention.

(A) chemical composition

[0037] The reason for limitation on each element is as follows. It is to be noted that a symbol "%" concerning a content in the following description represents "mass %".

C: 0.020 to 0.070%

[0038] C (carbon) may bond to elements such as Nb and Ti to form a precipitate in a steel sheet and contribute to improvement in strength by precipitation strengthening. When a content of C is less than 0.020%, a sufficient effect of the action cannot be obtained. On the other hand, more than 0.070% of the content of C may cause an increase in iron-carbides, which may be an initiation point of cracking during hole expansion working, and thus the hole expansion value is degraded. Accordingly, the content of C is 0.020 to 0.070%. The content of C is preferably 0.025% or more, and more preferably 0.030% or more. In addition, the content of C is preferably 0.060% or less, and more preferably 0.050% or less.

Si: 0.05 to 1.70%

[0039] Si (silicon) has a deoxidation effect, and an effect of inhibiting precipitation of iron-carbides such as cementite in a material microstructure and contributing to improvement in ductility and hole expandability. However, when the content is excessive, ferritic transformation may be likely to occur in a high temperature range, and thus carbides including Ti may be likely to precipitate. The precipitation of carbide in a high temperature range is likely to cause variation in the amount of precipitation, resulting in material variation in strength, hole expandability, and the like. Accordingly, a content of Si is 0.05 to 1.70%.

[0040] In view of inhibiting generation of scale-based defects such as fish-scale defects and spindle-shaped scale, the content of Si is preferably 0.06% or more, and more preferably 0.08% or more. In addition, the content of Si is preferably 1.50% or less, and in view of improving chemical treatability and a corrosion resistance after painting, the content of Si is more preferably 1.00% or less.

Mn: 0.60 to 2.50%

[0041] Mn (manganese) is an element that may contribute to improvement in strengthening and hardenability of ferrite. On the other hand, excessively contained Mn may cause unnecessarily high hardenability, which may prevent ferrite from being secured sufficiently and cause slab cracking during casting. Accordingly, a content of Mn is 0.60 to 2.50%. The content of Mn is preferably 1.00% or more, and more preferably 1.50% or more. In addition, the content of Mn is preferably 2.00% or less, and more preferably 1.80% or less.

Al: 0.010 to 1.000%

[0042] Al (aluminum) has a deoxidation effect and an effect of generating ferrite, as with Si. On the other hand, an excessive content may lead to embrittlement and be likely to cause clogging of a tundish nozzle during casting. Accordingly, a content of Al is 0.010 to 1.000%. The content of Al is preferably 0.015% or more, or 0.020% or more, and more preferably 0.025% or more, or 0.030% or more. In addition, the content of Al is preferably 0.800% or less, 0.700% or less, or 0.600% or less, and more preferably 0.500% or less, or 0.400% or less.

N: more than 0% to 0.0030% or less

[0043] When excessively contained, N (nitrogen) may cause not only a decrease in ductility due to remaining dissolved nitrogen, but also a decrease in hole expandability due to precipitation of TiN. Accordingly, a content of N is 0.0030% or less. The content of N is preferably 0.0025% or less.

P: 0.050% or less

[0044] P (phosphorus) is an impurity contained in molten pig iron, and since P may degrade local ductility due to grain boundary segregation and degrade weldability, a content of P is preferably as small as possible. Accordingly, the content of P is limited to 0.050% or less. The content of P is preferably 0.030% or less or 0.020% or less. It is not particularly necessary to define a lower limit, and the lower limit is 0%. However, an excessive reduction in the content of P leads to an increase in costs during smelting, and thus the lower limit may be 0.001%.

S: 0.005% or less

[0045] S (sulfur) is also an impurity contained in molten pig iron, and since S may degrade local ductility and weldability due to formation of MnS, a content of S is preferably as small as possible. Accordingly, the content of S is limited to 0.005% or less. To improve ductility and weldability, the content of S may be 0.003% or less or 0.002% or less. It is not particularly necessary to define a lower limit, and the lower limit is 0%. However, an excessive reduction in the content of S leads to an increase in costs during smelting, and thus the lower limit may be 0.0005%.

Ti: 0.015 to 0.170%

[0046] Ti (titanium) has an effect of improving low temperature toughness because carbo-nitride or dissolved Ti may cause a delay in grain growth during hot rolling and thus refine grain diameter in a hot rolled sheet. Further, Ti may be finely dispersed in ferrite as TiC, so that it contributes to strengthening of the steel sheet through precipitation strengthening. However, an excessive content may cause saturation of the effect and make TiN, which is a hard precipitate, to likely to precipitate. Accordingly, a content of Ti is 0.015 to 0.170%. The content of Ti is preferably 0.030% or more,

0.045% or more or 0.060% or more, and more preferably 0.070% or more, 0.080% or more, 0.090% or more, or 0.100% or more. In addition, the content of Ti is preferably 0.160% or less, 0.150% or less, 0.140% or less, 0.130% or less, or 0.120% or less.

Nb: 0 to 0.100%

[0047] Nb (niobium) has an effect of improving low temperature toughness because carbo-nitride or dissolved Nb may cause a delay in grain growth during hot rolling and thus refine grain diameter in a hot rolled sheet. Further, Nb may be present as NbC, so that it contributes to strengthening of the steel sheet through precipitation strengthening. Accordingly, Nb may be contained as necessary. However, an excessive content may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of Nb is 0.100% or less. The content of Nb may be 0.080% or less, 0.060% or less, or 0.050% or less, as necessary. A lower limit of Nb is 0%. However, the lower limit may be 0.001% or 0.010% in order to produce the effect sufficiently.

V: 0 to 0.300%

[0048] V (vanadium) is an element that has an effect of improving strength of a steel sheet by precipitation strengthening or solid solution strengthening. Accordingly, V may be contained as necessary. However, an excessive content may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of V is 0.300% or less. The content of V may be 0.200% or less, 0.100% or less, or 0.060% or less, as necessary. A lower limit of Nb is 0%. However, the lower limit may be 0.001% or 0.010% in order to produce the effect sufficiently.

Cu: 0 to 2.00%

[0049] Cu (copper) is an element that has an effect of improving strength of a steel sheet by precipitation strengthening or solid solution strengthening. Accordingly, Cu may be contained as necessary. However, an excessive content may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of Cu is 2.00% or less. Further, a large amount of Cu content may cause a blemish due to a scale on a surface of the steel sheet. Accordingly, the content of Cu may be 1.20% or less, 0.80% or less, 0.50% or less, or 0.25% or less. A lower limit of Cu is 0%. However, the lower limit of Cu may be 0.01% in order to produce the effect sufficiently.

Ni: 0 to 2.00%

[0050] Ni (nickel) is an element that has an effect of improving strength of a steel sheet by solid solution strengthening. Accordingly, Ni may be contained as necessary. However, an excessive content may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of Ni is 2.00% or less. Further, a large amount of Ni content may cause degradation of ductility. Accordingly, the content of Ni may be 0.60% or less, 0.35% or less, or 0.20% or less. A lower limit of Ni is 0%. However, the lower limit of Ni may be 0.01% in order to produce the effect sufficiently.

Cr: 0 to 2.00%

[0051] Cr (chromium) is an element that has an effect of improving strength of a steel sheet by solid solution strengthening. Accordingly, Cr may be contained as necessary. However, an excessive content may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of Cr is 2.00% or less. To improve economy, an upper limit of Cr may be 1.00%, 0.60%, or 0.30%. A lower limit of Cr is 0%. However, the lower limit of Cr may be 0.01% in order to produce the effect sufficiently.

Mo: 0 to 1.00%

[0052] Mo (molybdenum) is an element that has an effect of improving strength of a steel sheet by precipitation strengthening or solid solution strengthening. Accordingly, Mo may be contained as necessary. However, an excessive content may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of Mo is 1.00% or less. To improve economy, an upper limit of Mo may be 0.60%, 0.30%, or 0.10%. A lower limit of Mo is 0%. However, the lower limit of Mo may be 0.005% or 0.01% in order to produce the effect sufficiently.

B: 0 to 0.0100%

[0053] B (boron) segregates at a grain boundary, and may increase grain boundary strength to improve low temperature

toughness. Accordingly, B may be contained as necessary. However, an excessive content may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of B is 0.0100% or less. Further, B is a strong quench-hardening element, and a large amount of B content may prevent ferritic transformation from sufficiently progressing during cooling and sufficient retained austenite may not be obtained. Accordingly, a content of B may be 0.0050% or less, 0.0020% or less, or 0.0015%. A lower limit of B is 0%. However, the lower limit of B may be 0.0001% or 0.0002% in order to produce the effect sufficiently.

Mg: 0 to 0.0100%

[0054] Mg (magnesium) is an element that controls a morphology of nonmetal inclusions, which may serve as an initiation point of fracture and may be a cause of degradation in workability, to improve the workability. Accordingly, Mg may be contained as necessary. However, an excessive contact may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of Mg is 0.0100% or less. A lower limit of Mg is 0%. However, the lower limit of Mg may be 0.0001% or 0.0005% in order to produce the effect sufficiently.

Ca: 0 to 0.0100%

[0055] Ca (calcium) is an element that controls a morphology of nonmetal inclusions, which may serve as an initiation point of fracture and may be a cause of degradation in workability, to improve the workability. Accordingly, Ca may be contained as necessary. However, an excessive contact may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of Ca is 0.0100% or less. A lower limit of Ca is 0%. However, the content of Ca is preferably 0.0005% or more in order to produce the effect sufficiently.

REM: 0 to 0.1000%

[0056] REM (rare earth metal) is an element that controls a morphology of nonmetal inclusions, which may serve as an initiation point of fracture and may be a cause of degradation in workability, to improve the workability. Accordingly, REM may be contained as necessary. However, an excessive contact may cause saturation of the effect, leading to a decrease in economy. Accordingly, a content of REM is 0.1000% or less. An upper limit of REM may be 0.0100% or 0.0060%, as necessary. A lower limit of REM is 0%. However, the lower limit of REM may be 0.0001% or 0.0005% in order to produce the effect sufficiently.

[0057] Here, in the present invention, REM refers to a total of 17 elements of Sc, Y and lanthanoid, and the content of REM means a total content of these elements. It is to be noted that lanthanoid is industrially added in the form of a mischmetal.

Zr: 0 to 1.000%

Co: 0 to 1.000%

Zn: 0 to 1.000%

W: 0 to 1.000%

[0058] It has been confirmed that when Zr, Co, Zn, and W are each 1.000% or less, the effect of the present invention is unimpaired even if contained. An upper limit of each of them may be 0.300% or 0.10%. A total content of Zr, Co, Zn, and W is preferably 1.000% or less, or 0.100%. These elements may not necessarily be contained, and a lower limit is 0%, although the lower limit may be 0.0001% as necessary.

Sn: 0 to 0.050%

[0059] It has been confirmed that the effect of the present invention is unimpaired if a small amount of Sn (tin) is contained. However, the content of more than 0.05% may be a cause of a flaw during hot rolling. Accordingly, a content of Sn is 0.050% or less. Sn may not necessarily be contained, and a lower limit is 0%, although the lower limit may be 0.001% as necessary.

[0060] In the chemical composition of the steel sheet of the present invention, the balance is Fe and impurities.

[0061] The "impurity" as used herein refers to a raw material such as ore and scrap and a component contained due to various factors in production processes, and one allowed to the extent that the present invention is not adversely affected.

(B) Metal Microstructure

[0062] Description will now be made as to a metal microstructure of a steel sheet of the present invention. It is to be noted that when a width and a thickness of the steel sheet in a cross section perpendicular to a rolling direction of the steel sheet are defined as W and t, respectively, a metal microstructure in the present invention refers to a microstructure that is present at a position 1/4W or 3/4W from an end face of the steel sheet and 1/4t or 3/4t from a surface of the steel sheet. Further, a symbol "%" in the following description represents "area %".

Precipitation strengthened ferrite: 5 to 70%

[0063] Ferrite resulting from precipitation-strengthened pro-eutectoid ferrite in which Ti carbides are finely dispersed (hereafter, also referred to as "precipitation strengthened ferrite"), the Ti carbides being present in the ferrite due to interphase boundary precipitation or homogeneous nucleation driven by supersaturation of the Ti carbides while fine precipitates containing Ti (carbides or the like of finely precipitated Ti, and hereafter, also referred to as "fine Ti precipitates") are subjected to γ -> α transformation during cooling after rolling. The precipitation strengthened ferrite is a microstructure necessary for achieving both excellent uniform elongation and strength.

[0064] However, when an area fraction of precipitation strengthened ferrite is less than 5%, it is difficult to achieve both the uniform elongation and strength, while on the other hand, when it is more than 70%, local ductility is degraded even through the uniform elongation is excellent. Accordingly, the area fraction of the precipitation strengthened ferrite is 5 to 70%. In view of securing a balance between the uniform elongation and the strength, the area fraction of the precipitation strengthened ferrite is preferably 7% or more, and more preferably 10% or more. In addition, the area fraction of the precipitation strengthened ferrite is preferably 65% or less, and more preferably 60% or less.

[0065] The precipitation strengthened ferrite of the present invention as used herein refers to ferrite in which a number density of fine Ti precipitates contained in the grains is 1.0×10^{16} to $50.0 \times 10^{16}/cm^3$. When the number density of fine Ti precipitates contained in the ferrite grains is less than $1.0 \times 10^{16}/cm^3$, the effect of precipitation strengthening cannot be obtained sufficiently. On the other hand, when the number density of fine Ti precipitates is more than $50.0 \times 10^{16}/cm^3$, not only the strength saturates but also the ductility decreases.

[0066] In other words, the condition that the area fraction of the precipitation strengthened ferrite is 5 to 70% means that the area fraction of ferrite is 5 to 70% and the number density of fine Ti precipitates contained in the ferrite grains is 1.0×10^{16} to $50.0 \times 10^{16}/cm^3$.

[0067] Further, an average circle-equivalent diameter of fine Ti precipitates contained in the grains of precipitation strengthened ferrite is preferably 1.00 to 3.00 nm. This is because when the average circle-equivalent diameter of fine Ti precipitates is less than 1.00 nm, the effect of precipitation strengthening is less likely to be obtained, while on the other hand, when the precipitates are coarse grained and the average circle-equivalent diameter is more than 3.00 nm, it is difficult to secure a sufficient amount of fine Ti precipitates.

Bainite: 30 to 95%

[0068] Bainite is a microstructure necessary for balancing strength and local ductility, and has an effect of inhibiting crack propagation. However, too much bainite leads to a decrease in ferrite, and uniform elongation is significantly degraded even though local ductility is excellent. Accordingly, an area fraction of bainite is 30 to 95%. The area fraction of bainite is preferably 80% or less, and more preferably 70% or less in the case in which uniform elongation is emphasized.

Retained austenite: 2% or less

[0069] A high burring steel is characterized by presence of precipitation strengthened ferrite and bainite for securing workability as well as high strength such that both strength and workability are achieved. However, presence of thermodynamically stable retained austenite, which has not been subjected to martensitic transformation, in a steel sheet indicates that the retained austenite may have high concentration of C, and hardness of martensite generated by strain induced transformation of the retained austenite during sheet metal forging may become too high, promoting void generation. Accordingly, the amount of retained austenite is preferably as small as possible, and an area fraction of the retained austenite is 2% or less. The area fraction of the retained austenite is preferably 1.5% or less, 1% or less, or 0.5% or less. It is not particularly necessary to define a lower limit, and the lower limit is 0%, which is most preferable.

Martensite: 2% or less

[0070] A high burring steel is characterized by presence of precipitation strengthened ferrite and bainite for securing workability as well as high strength such that both strength and workability are achieved. However, when an area fraction

of martensite, which is a hard microstructure, is more than 2%, a void is likely to be generated at a border between the martensite and ferrite and rupture is likely to occur as strain of a steel sheet increases by sheet metal forging. Accordingly, an area fraction of martensite is 2% or less. The area fraction of martensite is preferably 1.5% or less, 1% or less, or 0.5% or less. It is not particularly necessary to define a lower limit, and the lower limit is 0%.

Pearlite: 1% or less

[0071] Pearlite may serve as an initiation point of fracture during hole expansion, and thus an area fraction of pearlite is 1% or less. The area fraction of pearlite is preferably 0.5% or less. It is preferable to reduce the area fraction of pearlite as much as possible and the area fraction is preferably 0%.

Total of precipitation strengthened ferrite and bainite: 95% or more

[0072] A high burring steel includes precipitation strengthened ferrite providing both excellent uniform elongation and strength and bainite providing both strength and local ductility. Excellent strength, uniform elongation, and local ductility are thereby obtained. When a total area fraction of precipitation strengthened ferrite and bainite is less than 95%, these properties are degraded. Accordingly, the total area fraction of precipitation strengthened ferrite and bainite is 95% or more. The total area fraction is preferably 97% or more, and more preferably 98% or more.

[0073] Here, in the present invention, an area fraction of metal microstructures is determined as follows. A sample is taken at a position 1/4W or 3/4W from an end face of the steel sheet and 1/4t or 3/4t from a surface of the steel sheet, as described above. Then, a rolling direction cross section (so-called L-direction cross section) of the sample is observed.

[0074] Specifically, the sample is subjected to Nital etching and observed in a 300 μm $\times$ 300 μm field of view using an optical microscope after the etching. Then, a resultant microstructure photograph is subjected to image analysis to obtain an area fraction A of ferrite, an area fraction B of pearlite, and a total area fraction C of bainite, martensite, and retained austenite.

[0075] Next, the portion subjected to Nital etching is subjected to Lepera etching and observed in a 300 μm $\times$ 300 μm field of view using an optical microscope. Then, a resultant microstructure photograph is subjected to image analysis to calculate a total area fraction D of retained austenite and martensite. Further, a sample subjected to facing up to a depth of 1/4 sheet thickness from a normal direction of the sheet surface is used to determine a volume ratio of the retained austenite with X-ray diffraction measurement. Since the volume ratio is substantially equal to the area fraction, the volume ratio is defined as an area fraction E of the retained austenite. An area fraction of bainite is determined from a difference between the area fraction C and the area fraction D, and an area fraction of martensite is determined from a difference between the area fraction E and the area fraction D. In this way, the area fraction of each of ferrite, bainite, martensite, retained austenite, and pearlite can be determined.

[0076] Further, an area fraction of precipitation strengthened ferrite can be obtained with Kernel Average Misorientation (KAM) method that comes with EBSP-OIMTM (Electron Back Scatter Diffraction Pattern-Orientation Image Microscopy).

[0077] In KAM method, averages of orientation differences among 6 adjacent pixels (first approximation) of some regular-hexagonal pixels in the measurement data, 12 pixels (second approximation) outside thereof, and 18 pixels (third approximation) further outside thereof are calculated, and each averaged value is taken as a value for the center pixel to perform calculation on each pixel. A map representing an orientation variation in grains can be created by performing the calculation such that the grain boundary is not crossed. In other words, the map represents a strain distribution based on a local orientation variation in the grains.

[0078] As an analysis condition for precipitation strengthened ferrite in the present invention, an average orientation difference is calculated between adjacent pixels in the third approximation in EBSP-OIMTM, and a portion with a calculated orientation difference being 1° or less is determined to be precipitation strengthened ferrite.

[0079] A temperature range in which precipitation strengthened ferrite of the present invention is generated coincides with a temperature range in which interphase boundary precipitation or homogeneous nucleation of Ti carbides occurs in the ferrite by being driven by supersaturation of the Ti carbides while the Ti carbides are subjected to γ -> α transformation during cooling after rolling. Since polygonal pro-eutectoid ferrite subjected to transformation at a high temperature is generated in diffusion transformation, it has a small dislocation density and less strain in grains, and thus a small inter-grain difference in crystal orientation. Accordingly, the precipitation strengthened ferrite also has a small crystal orientation difference. According to the results of various studies conducted by the present inventors so far, this is because a polygonal ferrite area fraction obtained in an optical microscope observation substantially matches with an area fraction of an area obtained when an orientation difference in the third approximation as measured with the KAM method is 1° or less.

[0080] In detail, the area fraction of precipitation strengthened ferrite was measured as follows. A sample taken in the same way as described in the microstructure observation was polished with a colloidal silica abrasive for 30 to 60 minutes, and was subjected to the EBSP measurement under the following conditions: magnification of 400x, area of 160 μm $\times$

256 μm, and a measurement step of 0.5 μm. The EBSP-OIMTM method is constituted by a device and software, with which a sample highly inclined in a scanning electron microscope (SEM) is irradiated with an electron beam, and a high sensitivity camera is used to capture a Kikuchi pattern formed by backscattering, followed by computer image processing, so that a crystal orientation at the irradiated point is measured in a short time.

[0081] In the EBSP method, a quantitative analysis can be made on microscopic structures and crystal orientations on a bulk sample surface, the analyzable area is an area that can be observed by SEM, and the analysis can be made at a minimum resolution of 20 nm depending on the resolution of the SEM. The analysis takes several hours, and a region to be analyzed is mapped to tens of thousands of points in an equidistant grid. In a polycrystalline material, a crystal orientation distribution and a size of grain in the sample can be seen.

[0082] In this way, a portion with a calculated orientation difference in the third approximation being 1° or less was determined to be precipitation strengthened ferrite, an area of the precipitation strengthened ferrite was determined, and an area fraction of the precipitation strengthened ferrite was determined with respect to the measured area.

[0083] The observation of a fine Ti precipitate was conducted with a 3D Atom Probe Tomography as follows.

[0084] First, a needle-shaped sample is produced from a sample to be measured by cutting and electrolytic polishing, and as necessary, with the use of focused ion beam processing along with the electrolytic polishing. In the 3D Atom Probe Tomography, accumulated data can be reconstructed to determine an actual distribution of atoms in a real space. In the case of a fine Ti precipitate of a Na-Cl structure, since a unit cell is 4.33 angstrom, an interatomic distance between Ti and Ti was defined as $4.33 \times \sqrt{2} = 6.1$ angstrom.

[0085] Then, when a plurality of Ti atoms present at a substantially same coordinate position (7 angstrom or less), it was determined that these Ti atoms are within the same precipitate. The number of Ti atoms determined to be within the same precipitate is counted, and when the number was 50 or more, the precipitate was defined as a fine Ti precipitate.

[0086] The size of the fine Ti precipitate is a circle-equivalent diameter calculated assuming that the fine Ti precipitate is spherical from the number of Ti atoms constituting the observed fine Ti precipitate and a lattice constant of the fine Ti precipitate.

[0087] How the number of Ti atoms in a fine Ti precipitate obtained in the 3D Atom Probe Tomography is used to determine a circle-equivalent diameter R of the precipitate is as follows.

[0088] Although the 3D Atom Probe Tomography is used to measure the number N of all atoms in the sample of interest, in practice, the 3D Atom Probe Tomography cannot detect the number N of all atoms in the sample of interest. Since any device has its own detection ratio α of atoms (= number of detected atoms/total number of atoms), the number N of atoms that would have been present is calculated based on an actual measured value n. Accordingly, the total number of atoms $N = n/\alpha$. The detection ratio α of the instrument used for measurement in the present invention was 0.35.

[0089] Next, given the number N of atoms, the circle-equivalent diameter is calculated in the following formula, assuming that 8 Ti atoms are present on the unit cell in the case of a Ti precipitate of a Na-Cl structure and the lattice constant "a" of a Na-Cl structure is 4.33 angstrom:

$$\text{circle-equivalent diameter } R = \{(6/8)\cdot(1/\pi)\cdot N\cdot a^3\}^{(1/3)}$$

[0090] For example, when the number of Ti is 50, the circle-equivalent diameter is calculated as substantially 1 nm. In the present invention, circle-equivalent diameters of 30 or more fine Ti precipitates are arbitrarily measured and an average thereof is determined.

[0091] The number density of fine Ti precipitates is determined with a measurement field of view being taken as a denominator and the number of the fine Ti precipitates as a numerator. In measurement of the number density, measurement was made in 5 or more fields of view, each of the field of view being 10 nm (sheet thickness direction t) $\times$ 40 nm (sheet width direction W) $\times$ 60 nm (sheet longitudinal direction L), and an average of the number densities (number of the fine Ti precipitates/cm^3) was determined.

[0092] In the present invention, how TiN is present is defined as follows.

Average circle-equivalent diameter of TiN: 1.0 to 10.0 μm

[0093] When TiN is large, voids present at a grain boundary are likely to coalesce with each other as strain in a steel sheet due to the sheet metal forging increases, and accordingly, the average circle-equivalent diameter of TiN is 10.0 μm or less. To more reliably secure the effects, the average circle-equivalent diameter of TiN is preferably 8.0 μm or less, and more preferably 5.0 μm or less.

[0094] Th smaller TiN is, the more preferable it is, and thus it is not essentially necessary to define a lower limit for the average circle-equivalent diameter of TiN. However, in a TiN observation method described below, it is difficult to determine whether the object is TiN when the circle-equivalent diameter of TiN is less than 1.0 μm. Accordingly, in the present invention, an object that has 1.0 μm or more of the circle-equivalent diameter is to be measured as TiN. Ac-

cordingly, an average circle-equivalent diameter of TiN is 1.0 μm or more.

[0095] The average circle-equivalent diameter of TiN is determined as follows. A sample is taken at a position 1/4W or 3/4W from an end face of the steel sheet and 1/4t or 3/4t from a surface of the steel sheet, as described above. Then, a rolling direction cross section (so-called L-direction cross section) of the sample is polished and observed with the sample unetched. Specifically, an optical microscope is used to capture a microstructure photograph at a magnification of 1000x, and the microstructure photograph is observed visually or using an image processing device and the like.

[0096] In the microstructure photograph, circle-equivalent diameters are determined for possible TiN precipitates, and those that have 1.0 μm or more of the circle-equivalent diameter are determined as TiN. Then, observation is made in 20 or more fields of view, each field of view being 60 μm (rolling direction L) × 40 μm (sheet thickness direction t), and an average of all the circle-equivalent diameters of TiN precipitates is defined as the average circle-equivalent diameter of TiN.

Average of minimum distances between adjacent TiN precipitates: 10.0 μm or more

[0097] To avoid the growth of voids generated at an interface between TiN and ferrite and prevent the voids from coalescing with each other into a larger void, it is necessary to secure a certain amount of distance between TiN precipitates. Accordingly, an average of distances between adjacent TiN precipitates is 10.0 μm or more.

[0098] In view of preventing crack generation due to void growth, the average is preferably 15.0 μm or more, and more preferably 20.0 μm or more. No upper limit is particularly defined. However, precipitation of TiN is unavoidable to some extent, and accordingly, an average of minimum distances between adjacent TiN precipitates is preferably 1000 μm or less.

[0099] The average of minimum distances between adjacent TiN precipitates is determined as follows. 20 TiN precipitates are arbitrarily selected, every distances between one of the TiN precipitates and another one most adjacent to it are calculated, and an average thereof is calculated. The minimum distance between TiN precipitates is measured as with the case of the average circle-equivalent diameters.

(C) Mechanical Properties

Standard deviation of nano hardness: 1.0 GPa or less

[0100] It is possible to inhibit voids from coalescing with each other and growing into a crack by reducing a difference in deformability between a hard microstructure and a soft microstructure to reduce voids generated at an interface between the two microstructures and to create a void spacing. Accordingly, it is possible to inhibit void generation by reducing a nano hardness difference, which corresponds to the difference in deformability between a hard microstructure and a soft microstructure. In the present invention, a standard deviation of nano hardness in a sample cross section is employed as an indicator for a hardness difference between a soft microstructure and a hard microstructure.

[0101] Nano hardness can be measured with the use of, for example, TriboScope/TriboIndenter available from Hysitron. The systems can arbitrarily measure nano hardness at 100 or more points at a load of 1 mN, and calculate a standard deviation of the nano hardness from the results.

[0102] To reduce a hardness difference between a soft microstructure and a hard microstructure to inhibit void generation, a smaller standard deviation of nano hardness is preferable, and accordingly, it is 1.0 GPa or less. The standard deviation of nano hardness is preferably 0.8 GPa or less.

Tensile strength: 780 MPa or more

[0103] The steel sheet according to the present invention preferably has a tensile strength of 780 MPa or more, which is a similar level to a conventional high burring steel. It is not particularly necessary to define an upper limit to the tensile strength. However, it may be 1200 MPa, 1150 MPa, or 1000 MPa. Here, the tensile strength refers to a tensile strength in JIS Z 2241 (2011).

Product of uniform elongation and tensile strength: 7000 MPa·% or more

[0104] A small uniform elongation is likely to be a cause of sheet thickness reduction due to necking during press forming, and then a cause of press cracking. To secure press formability, it is preferable to satisfy a product of a uniform elongation (u-EL) and a tensile strength (TS): TS × u-EL ≥ 7000 MPa%. Here, in a test defined in JIS Z 2241 (2011), the uniform elongation is represented by the following formula:

$$\text{uniform elongation (u-EL)} = \ln(\varepsilon n0 + 1)$$

where in a relation between a nominal stress σn and a nominal strain εn, $\varepsilon n0$ is a nominal strain at a point where a value obtained by differentiating the nominal stress σn with the nominal strain εn is zero.

Product of hole expansion ratio and tensile strength: 50000 MPa·% or more

[0105] A poor hole expandability leads to a poor material flowability during stretch flanging, and may possibly cause a cracking. To secure hole expandability, it is preferable to satisfy a product of a hole expansion ratio (λ) and a tensile strength (TS): (TS) $\times$ (λ) $\geq$ 50000 MPa%. Here, the hole expansion ratio (λ) represents a hole expansion ratio (λ) according to a test method in conformity with JIS Z 2256 (2010).

Equivalent plastic strain: 0.9 or more

[0106] The equivalent plastic strain is converted using a relation between a shearing stress σs and a shear plastic strain εsp in a simple shear test into a relation between a tensile stress σ and a tensile strain ε in a uniaxial tensile test, which is different in deformation mode, and a constant, conversion factor (κ) is used to make a conversion, assuming an isotropic hardening rule and a plastic work conjugate relationship.

[0107] Here, the isotropic hardening rule is a work hardening rule in which it is assumed that the shape of yield curve does not change even when a strain develops (that is, it expands in a similar shape). The plastic work conjugate relationship is a relationship in which work hardening is described only as a function of a plastic work, and exhibits the same amount of work hardening given the same plastic work ($\sigma \times \varepsilon$) regardless of the deformation mode.

[0108] A shearing stress and a shear plastic strain in a simple shear test can thereby converted into a tensile stress and a tensile strain in a uniaxial tensile test. The relation is shown below.

$$\text{uniaxial tensile test tensile stress } \sigma \text{ (converted)} = \text{simple shear test shearing stress}$$

$$\sigma s \times \kappa$$

$$\text{uniaxial tensile test tensile strain } \varepsilon \text{ (converted)} = \text{simple shear test shear plastic strain } \varepsilon sp/\kappa$$

[0109] Next, conversion factor κ is determined such that a relation between a shearing stress and a shear plastic strain is similar to a relation between a tensile stress and a tensile strain. For example, the conversion factor κ can be determined in the following procedure. First, a relation between a tensile strain ε (actual value) and a tensile stress σ (actual value) in a uniaxial tensile test is determined. Then, a relation between a shearing strain εs (actual value) and a shearing stress σs (actual value) in a uniaxial shear test.

[0110] Next, "κ" is changed to determine a tensile strain ε (converted) determined from the shearing strain εs (actual value) and a tensile stress σ (converted) determined from the shearing stress σs (actual value). Then, the tensile stress σ (converted) when the tensile strain ε (converted) is from 0.2% to uniform elongation (u-EL) is determined. At this time, an error between the tensile stress σ (converted) and the tensile stress σ (actual value) is determined, and "κ" that minimizes the error is determined with the method of least squares.

[0111] An equivalent plastic strain εeq is defined as a shear plastic strain εsp (rupture) at the time of rupture in a simple shear test converted, with the use of the determined κ, into a tensile strain ε in a simple tensile test.

[0112] The steel sheet according to the present invention is characterized by good workability in a high strain domain typified by sheet metal forging, and its equivalent plastic strain εeq satisfies 0.50 or more. Since the equivalent plastic strain of a conventional TRIP steel at best on the order of 0.30, it has been confirmed that the steel sheet according to the present invention has a good sheet forgeability.

(D) Dimension

Sheet thickness: 1.0 to 4.0 mm

[0113] The steel sheet according to the present invention finds application primarily in automobiles and the like and the sheet thickness is ranging primarily from 1.0 to 4.0 mm. Accordingly, the range of sheet thickness may be from 1.0

to 4.0 mm, and, as necessary, a lower limit may be 1.2 mm, 1.4 mm, or 1.6 mm, and an upper limit may be 3.6 mm, 3.2 mm, or 2.8 mm.

(E) Production method

[0114] From studies so far, the present inventors confirmed that the hot rolled steel sheet of the present invention can be produced by conducting the following production processes (a) to (f) in this order. Description will now be made as to each of the production processes in detail.

(a) Melting process

[0115] Production methods prior to hot rolling are not particularly limited. Accordingly, subsequent to melting in a blast furnace or an electric furnace, a variety of second smelting is executed to make an adjustment for a component composition described above. Then, methods such as general continuous casting and thin slab casting may be used to produce a slab. At this time, scrap or the like may be used as raw materials provided that the material can be controlled into the component range of the present invention.

(b) Heating process

[0116] A produced slab is subjected to hot rolling into a hot rolled steel sheet. For hot rolling, the slab is heated first. In the heating process, the slab is heated to a temperature of SRTmin°C or more represented by the following formula (i), or to 1260°C or less. In the case of continuous casting, the slab may be cooled to a low temperature, and then heated again, or may be heated subsequent to the continuous casting without cooling. Here, SRTmin refers to a solution treatment temperature of TiC.

$$SRTmin = 7000/\{2.75 - \log(Ti \times C)\} - 273 \ldots (i)$$

where a symbol of an element in the above formula represents a content (in mass %) of the element in the hot rolled steel sheet and is substituted by zero when the element is not contained.

(c) Continuous hot rolling process

[0117] After heating, the slab extracted from a heating furnace is subjected to rough rolling and subsequent multi-stand finish rolling. To avoid precipitation of precipitates including Ti, an end temperature of the rough rolling is 1100°C or more. As described above, the multi-stand finish rolling is conducted by continuous rolling at multiple, three stands or more (for example, 6 or 7 stands). The multi-stand finish rolling is executed such that a cumulative strain (effective cumulative strain) of rolling at final three stands is 0.01 to 0.10.

[0118] As described above, the effective cumulative strain is an indicator that takes into consideration a grain size variation according to temperature during rolling and rolling reduction of a steel sheet by rolling and a grain size variation when grains statically recover in a time lapse after rolling. The effective cumulative strain (εeff) can be determined in the following formula:

$$\text{effective cumulative strain } (\varepsilon\text{eff}) = \Sigma\varepsilon i(ti, Ti) \ldots (iii)$$

where $\sum$ in the formula (iii) represents the sum for i = 1 to 3.
i = 1, i = 2, and i = 3 indicate a first stand of rolling from the last in the multi-stand finish rolling (that is, final stand rolling), a second stand of rolling from the last, and a third stand of rolling from the last, respectively.

[0119] Here, for each of rolling indicated by i, εi is represented by the following formula:

$$\varepsilon i(ti, Ti) = ei/\exp((ti/\tau R)^{2/3}) \ldots (iv)$$

where

ti: time (s) between i-th stand of rolling from the last and start of primary cooling
Ti: rolling temperature (K) of i-th stand of rolling from the last

ei: logarithmic strain when rolled at i-th stand of rolling from the last

$$ei = |\ln\{1 - (\text{i-th stand entry side sheet thickness} - \text{i-th stand delivery side sheet thickness})/(\text{i-th stand entry side sheet thickness})\}|$$

$$= |\ln\{(\text{i-th stand delivery side sheet thickness})/(\text{i-th stand entry side sheet thickness})\}| \ldots (v)$$

$$\tau R = \tau 0 \cdot \exp(Q/(R \cdot Ti)) \ldots (vi)$$

$$\tau 0 = 8.46 \times 10^{-9} \text{ (s)}$$

Q: constant of activation energy regarding movement of dislocations in Fe = 183200 (J/mol)
R: gas constant = 8.314 (J/(K·mol))

[0120] By the definition of the effective cumulative strain thus derived, intended microstructures are obtained and variation in nano hardness is reduced. As a result, a steel sheet with excellent sheet forgeability can be obtained, in which the void generation is inhibited at an interface between a hard microstructure and a soft microstructure and it is difficult for voids to coalesce with each other even when the voids grow, and thus sheet metal forging does not cause cracks.

[0121] An end temperature of the multi-stand finish rolling, that is, an end temperature of the continuous hot rolling process, may be satisfactory if it is Ar_3 (°C) + 30°C or more where Ar_3 is determined in the following formula (ii). This is because precipitation strengthened ferrite and bainite intended in the present invention can thus be obtained.

$$Ar_3 = 970 - 325 \times C + 33 \times Si + 287 \times P + 40 \times Al - 92 \times (Mn + Mo + Cu) - 46 \times (Cr + Ni) \ldots (ii)$$

where a symbol of an element in the above formula represents a content (in mass %) of the element in the hot rolled steel sheet and is substituted by zero when the element is not contained.

(d) First (accelerated) cooling process

[0122] After the multi-stand finish rolling is completed, cooling of the resultant hot rolled steel sheet is started after 1.00 to 5.00 seconds. Then, the sheet is cooled at an average cooling rate of 10°C/sec or more from the rolling end temperature down to a temperature of 650 to 750°C, and thereafter the sheet is held in air for 1 to 10 seconds.

[0123] When the cooling is started within less than 1.00 seconds after the completion of the continuous hot rolling process, ferritic transformation is promoted and not only an intended bainite area fraction cannot be obtained in final microstructures but also the effect of the present invention cannot be obtained because of coarsened precipitates. On the other hand, when the cooling is started beyond 5.00 seconds, ferritic transformation is delayed and an intended area fraction of precipitation strengthened ferrite cannot be obtained.

[0124] Further, when the average cooling rate of the first cooling process is less than 10°C/sec, pearlite is likely to be generated. On the other hand, an upper limit of the cooling rate is not particularly limited. However, it may be 300°C/sec or less in order to avoid supercooling. Further, when a holding temperature in air is less than 650°C, bainite is likely to be generated and the bainite area fraction increases. On the other hand, when a holding temperature in air is more than 750°C, pearlite is likely to be generated.

[0125] It is to be noted that "holding in air" as used herein includes a hot-rolled steel sheet being subjected to air cooling or minimally limited cooling in a cooling facility, and a lower limit at this time is ideally 0°C/sec and an upper limit is 8°C/sec.

(e) Second (accelerated) cooling process

[0126] After the sheet is held in air, cooling is conducted at an average cooling rate of 10°C/sec or more from a

temperature range of 600 to 740°C. A cooling starting temperature less than 600°C prevents ferritic transformation from sufficiently progressing and causes insufficient precipitation of fine Ti precipitates. On the other hand, when the cooling starting temperature is more than 740°C, ferritic transformation may excessively progress and pearlite may be generated, leading to degradation of hole expandability. In addition, fine Ti precipitates may be coarsened to reduce strength.

[0127] Further, when the average cooling rate is less than 10°C/sec, pearlite may also be generated, leading to degradation of hole expandability. An upper limit of the average cooling rate is not particularly limited. However, it may be 1000°C/sec or less because there is a concern that the steel sheet warps due to thermal strain caused by thermal deviation.

(f) Coiling process

[0128] Thereafter, the cooled hot rolled steel sheet is coiled at a coiling temperature of 450 to 650°C. Conditions after coiling process are not particularly limited.

(F) Steel Forged Part

[0129] The hot-rolled steel sheet thus obtained has excellent sheet forgeability. Accordingly, the hot-rolled steel sheet can be forged in sheet metal forging and the like to obtain forged parts in complex shapes, which are required to have high strength that could not have been achieved in the past.

[0130] The present invention will now be specifically described with reference to an example, although the present invention is not limited to the example.

EXAMPLE 1

[0131] A steel, which has a chemical composition shown in Table 1, was molten into a slab. The slab was hot rolled, cooled and then coiled under the conditions shown in Table 2 to produce a hot rolled steel sheet. Sheet thicknesses of resultant hot rolled steel sheets are shown in Table 3.

[Table 1]

[0132]

Table 1

Steel type	Chemical composition (in mass %, the balance: Fe and impurities)																			
	C	Si	Mn	Al	N	P	S	Ti	Nb	V	Cu	Ni	Cr	Mo	B	Mg	Ca	REM	Others
A	0.046	0.96	1.30	0.029	0.0018	0.008	0.003	0.149	0.040	-	-	-	-	-	-	-	-	-	-
B	0.039	0.94	1.04	0.032	0.0017	0.010	0.003	0.131	0.039	-	-	-	-	-	0.0009	-	-	-	-
C	0.040	0.06	1.50	0.337	0.0015	0.015	0.004	0.100	0.016	-	-	-	0.12	-	-	-	-	-	-
D	0.063	1.21	2.47	0.036	0.0019	0.012	0.005	0.110	0.022	-	-	-	-	-	-	-	-	-	-
E	0.041	0.99	1.25	0.037	0.0028	0.009	0.003	0.095	-	-	-	-	-	-	0.0008	-	-	-	-
F	0.040	1.00	1.30	0.030	0.0015	0.015	0.004	0.110	-	-	-	-	-	-	-	-	-	-	-
G	0.068	0.91	1.24	0.031	0.0014	0.012	0.005	0.107	-	-	-	-	-	-	-	-	-	-	-
H	0.024	1.05	1.43	0.023	0.0027	0.015	0.005	0.112	-	-	-	-	-	-	-	-	-	-	-
I	0.036	1.68	1.00	0.011	0.0024	0.012	0.003	0.098	-	-	-	-	-	-	-	-	-	-	-
J	0.051	1.28	0.63	0.013	0.0012	0.013	0.004	0.145	-	-	-	-	-	-	-	-	-	-	Zr: 0.001
K	0.045	0.93	1.08	0.031	0.0026	0.011	0.005	0.166	-	-	-	-	-	-	-	-	-	-	-
L	0.065	0.89	1.51	0.018	0.0007	0.015	0.003	0.019	-	-	-	-	-	-	-	-	-	-	-
M	0.074 *	1.03	1.23	0.030	0.0019	0.012	0.004	0.108	-	-	-	-	-	-	-	-	-	-	-
N	0.018 *	0.98	1.27	0.032	0.0018	0.008	0.003	0.111	-	-	-	-	-	-	-	-	-	-	-
O	0.038	1.73 *	1.32	0.025	0.0018	0.013	0.002	0.114	-	-	-	-	-	-	-	-	-	-	-
P	0.045	0.03 *	1.33	0.022	0.0017	0.012	0.003	0.128	-	-	-	-	-	-	-	-	-	-	-
Q	0.036	0.96	2.54 *	0.027	0.0019	0.014	0.003	0.098	-	-	-	-	-	-	-	-	-	-	-
R	0.048	1.10	0.55 *	0.025	0.0018	0.015	0.002	0.126	-	-	-	-	-	-	-	-	-	-	-
S	0.041	0.97	1.28	0.024	0.0034 *	0.013	0.002	0.110	-	-	-	-	-	-	-	-	-	-	-
T	0.044	0.94	1.25	0.023	0.0017	0.014	0.004	0.172 *	-	-	-	-	-	-	-	-	-	-	-
U	0.046	1.04	1.27	0.022	0.0016	0.014	0.004	0.011 *	-	-	-	-	-	-	-	-	-	-	-
V	0.045	0.97	1.25	0.036	0.0020	0.013	0.002	0.088	-	0.080	-	-	-	-	-	-	-	-	-
W	0.042	0.99	1.35	0.037	0.0018	0.014	0.004	0.096	-	-	0.20	-	-	-	-	-	-	-	-
X	0.043	0.96	1.22	0.030	0.0023	0.014	0.004	0.103	-	-	-	0.10	-	-	-	-	-	-	-
Y	0.036	0.95	1.31	0.031	0.0022	0.010	0.002	0.107	-	-	-	-	0.10	-	-	-	-	-	-
Z	0.033	1.02	1.33	0.025	0.0018	0.012	0.003	0.096	-	-	-	-	-	0.15	-	-	-	-	-
a	0.037	1.03	1.27	0.022	0.0020	0.015	0.005	0.112	-	-	-	-	-	-	-	0.0006	-	-	-
b	0.039	1.00	1.29	0.027	0.0022	0.018	0.004	0.108	-	-	-	-	-	-	-	-	0.0010	-	-
c	0.040	1.04	1.30	0.025	0.0021	0.016	0.005	0.102	-	-	-	-	-	-	-	-	-	0.0005	-

* indicates out of the definition of the present invention

[Table 2]

[0133]

Table 2

Test No.	Steel type	SRTmin (°C)	Ar_3 (°C)	Heating temperature (°C)	Rough rolling	Finish rolling		First cooling		Air cooling		Second cooling		Coiling
					End temperature (°C)	End temperature (°C)	Cumulative strain at final three stands	Time before start of cooling (s)	Average cooling rate (°C/s)	Start temperature (°C)	Time (s)	Start temperature (°C)	Cooling rate (°C/s)	Coiling temperature (°C)
1	A	1151	871	1200	1120	930	0.062	1.23	44	700	2	690	14	550
2	A	1151	871	1120	1120	930	0.062	1.23	44	700	2	690	14	550
3	A	1151	871	1200	1050	930	0.062	1.23	44	700	2	690	14	550
4	A	1151	871	1200	1120	880	0.140	1.23	34	700	2	690	14	550
5	A	1151	871	1200	1120	900	0.173	1.06	44	700	2	690	16	550
6	A	1151	871	1200	1120	1000	0.003	2.00	32	730	1	725	10	550
7	A	1151	871	1200	1120	900	0.011	5.15	10	680	1	675	10	450
8	A	1151	871	1200	1120	930	0.086	0.94	58	700	1	695	18	550
9	A	1151	871	1200	1120	930	0.007	4.09	9	750	2	740	10	450
10	A	1151	871	1200	1120	930	0.062	1.23	32	760	4	740	18	550
11	A	1151	871	1200	1120	930	0.062	1.23	56	640	1	635	13	500
12	A	1151	871	1200	1120	930	0.062	1.23	35	745	11	690	18	500
13	A	1151	871	1200	1120	930	0.062	1.23	52	660	0	660	11	550
14	A	1151	871	1200	1120	930	0.062	1.23	34	750	1	745	19	550
15	A	1151	871	1200	1120	930	0.062	1.23	54	650	10	590	14	450
16	A	1151	871	1200	1120	930	0.062	1.23	49	675	8	635	8	550
17	A	1151	871	1200	1120	930	0.062	1.23	35	745	1	740	10	700
18	A	1151	871	1200	1120	930	0.062	1.23	49	675	4	655	25	400
19	B	1115	897	1230	1150	950	0.042	1.60	30	745	1	740	10	600
20	C	1087	833	1230	1150	950	0.042	1.60	30	745	1	740	10	600

(continued)

Test No.	Steel type	SRTmin (°C)	Ar_3 (°C)	Heating temperature (°C)	Rough rolling	Finish rolling		First cooling		Air cooling		Second cooling		Coiling
					End temperature (°C)	End temperature (°C)	Cumulative strain at final three stands	Time before start of cooling (s)	Average cooling rate (°C/s)	Start temperature (°C)	Time (s)	Start temperature (°C)	Cooling rate (°C/s)	Coiling temperature (°C)
21	D	1153	767	1230	1150	950	0.042	1.60	30	745	1	740	10	600
22	E	1084	878	1230	1150	950	0.042	1.60	30	745	1	740	10	600
23	F	1098	876	1230	1150	950	0.042	1.60	30	745	1	740	10	600
24	G	1159	869	1230	1150	950	0.042	1.60	30	745	1	740	10	600
25	H	1043	871	1230	1150	950	0.035	1.77	27	745	1	740	11	580
26	I	1073	926	1230	1150	960	0.027	1.77	28	745	1	740	11	580
27	J	1161	942	1230	1150	975	0.018	1.77	30	745	1	740	11	580
28	K	1162	891	1230	1150	950	0.035	1.77	27	745	1	740	11	580
29	L	964	844	1230	1150	950	0.035	1.77	27	745	1	740	11	580
30	M *	1171	871	1230	1150	930	0.062	1.23	44	700	2	690	14	550
31	N *	1012	883	1230	1150	930	0.062	1.23	44	700	2	690	14	550
32	O *	1096	898	1230	1150	930	0.062	1.23	44	700	2	690	14	550
33	P *	1130	838	1230	1150	930	0.062	1.23	44	700	2	690	14	550
34	Q *	1073	761	1230	Inapplicable to rolling due to slab cracking									
35	R *	1136	945	1230	1150	930	0.062	1.23	44	700	2	690	14	550
36	S *	1101	876	1230	1150	930	0.062	1.23	44	700	2	690	14	550
37	T *	1164	877	1230	1150	930	0.062	1.23	44	700	2	690	14	550
38	U *	885	877	1230	1150	930	0.062	1.23	44	700	2	690	14	550
39	V	1086	878	1230	1150	915	0.098	1.33	42	680	5	655	14	500
40	W	1088	852	1230	1150	915	0.098	1.33	42	680	5	655	14	500
41	X	1099	876	1230	1150	915	0.098	1.33	42	680	5	655	14	500

(continued)

Test No.	Steel type	SRTmin (°C)	Ar_3 (°C)	Heating temperature (°C)	Rough rolling	Finish rolling		First cooling		Air cooling		Second cooling		Coiling
					End temperature (°C)	End temperature (°C)	Cumulative strain at final three stands	Time before start of cooling (s)	Average cooling rate (°C/s)	Start temperature (°C)	Time (s)	Start temperature (°C)	Cooling rate (°C/s)	Coiling temperature (°C)
42	Y	1082	869	1230	1150	915	0.098	1.33	42	680	5	655	14	500
43	Z	1061	861	1230	1150	915	0.098	1.33	42	680	5	655	14	500
44	a	1091	880	1230	1150	915	0.098	1.33	42	680	5	655	14	500
45	b	1093	878	1230	1150	915	0.098	1.33	42	680	5	655	14	500
46	c	1089	877	1230	1150	915	0.098	1.33	42	680	5	655	14	500
* indicates out of the definition of the present invention														

[Table 3]

[0134]

Table 3

Test No.	Steel type	Sheet thickness (mm)	Metal microstructures											Mechanical properties				
			Pearlite (area %)	Ferrite (area %)	Bainite (area %)	Ferrite + bainite (area%)	Martensite (area %)	Retained γ (area %)	TiN average circle-equivalent diameter (mm)	TiN average minimum distance (mm)	Average circle-equivalent diameter of microscopic Ti precipitates (nm)	Number density of microscopic Ti precipitates ($\times 10^{16}$/cm^3)	Nano hardness standard deviation (GPa)	TS (MPa)	TS×u-EL (MPa×%)	TS×λ (MPa×%)	Equivalent plastic strain	
1	A	29	0	35	65	100	0	0	6.0	11.0	2.4	10.0	0.95	790	7900	63990	0.98	Inv. example

(continued)

Test No.	Steel type	Sheet thickness (mm)	Metal microstructures											Mechanical properties				
			Pearlite (area %)	Ferrite (area %)	Bainite (area %)	Ferrite + bainite (area%)	Martensite (area %)	Retained γ (area %)	TiN average circle-equivalent diameter (mm)	TiN average minimum distance (mm)	Average circle-equivalent diameter of microscopic Ti precipitates (nm)	Number density of microscopic Ti precipitates ($\times 10^{16}$/cm^3)	Nano hardness standard deviation (GPa)	TS (MPa)	TS×u-EL (MPa×%)	TS×λ (MPa×%)	Equivalent plastic strain	
2	A	29	0	40	60	100	0	0	10.0	10.0	2.9	0.4 *	0.90	710	8520	68160	0.99	Comparative example
3	A	29	0	40	60	100	0	0	9.0	10.0	6.0	0.7 *	0.91	761	7991	66207	0.98	
4	A	29	0	75 *	25 *	100	0	0	6.5	11.5	2.7	7.0	1.05 *	820	9020	49200	0.87	
5	A	29	0	80 *	20 *	100	0	0	6.3	10.2	2.9	5.5	1.10 *	781	10153	42955	0.86	
6	A	29	0	3 *	97 *	100	0	0	6.4	10.3	-	- *	0.90	841	5887	89146	0.90	
7	A	29	0	4 *	96 *	100	0	0	6.2	10.1	-	- *	0.89	836	6270	83600	0.90	
8	A	29	0	71 *	29 *	100	0	0	6.3	10.4	3.0	5.0	1.03 *	784	9408	39200	0.88	
9	A	29	5 *	90 *	5 *	95	0	0	6.1	10.8	3.0	5.1	1.11 *	780	4680	31200	0.85	
10	A	29	8 *	89 *	3 *	92 *	0	0	6.0	10.8	3.0	4.8	1.12 *	781	5077	35145	0.84	
11	A	29	0	4 *	96 *	100	0	0	6.2	10.9	-	- *	0.87	844	5908	97060	0.91	
12	A	29	6 *	92 *	2 *	94 *	0	0	6.3	10.5	4.0	0.8 *	1.09 *	763	5341	30520	0.82	
13	A	29	0	0 *	98 *	98	2	0	6.4	10.3	-	- *	0.89	866	4763	108250	0.94	
14	A	29	6 *	90 *	4 *	94 *	0	0	6.2	10.1	3.6	1.0	1.08 *	760	5928	34200	0.81	
15	A	29	0	4 *	96 *	100	0	0	6.3	10.4	-	- *	0.90	851	5021	102120	0.95	
16	A	29	3 *	41	56	97	0	0	6.1	10.8	4.0	2.0	1.11 *	730	5110	40150	0.84	
17	A	29	4 *	40	56	96	0	0	6.0	10.8	7.0	0.4 *	1.12 *	722	5054	36100	0.82	
18	A	29	0	37	57	94 *	3 *	3 *	6.2	10.4	1.5	0.1 *	1.24 *	741	10004	33345	0.61	

(continued)

Test No.	Steel type	Sheet thickness (mm)	Metal microstructures											Mechanical properties				
			Pearlite (area %)	Ferrite (area %)	Bainite (area %)	Ferrite + bainite (area%)	Martensite (area %)	Retained γ (area %)	TiN average circle-equivalent diameter (mm)	TiN average minimum distance (mm)	Average circle-equivalent diameter of microscopic Ti precipitates (nm)	Number density of microscopic Ti precipitates ($\times 10^{16}$/cm^3)	Nano hardness standard deviation (GPa)	TS (MPa)	TS×u-EL (MPa×%)	TS×λ (MPa×%)	Equivalent plastic strain	
19	B	3.1	0	15	85	100	0	0	5.8	10.8	2.5	7.8	0.97	785	7850	51025	0.91	Inventive example
20	C	3.1	0	10	90	100	0	0	5.1	11.6	2.0	12.0	0.99	786	8253	66810	0.96	
21	D	3.1	0	8	92	100	0	0	5.5	12.3	1.5	30.0	0.98	1016	9144	66040	0.90	
22	E	3.1	0	20	80	100	0	0	5.4	13.1	2.0	11.0	0.94	820	8200	57400	0.94	
23	F	3.1	0	70	30	100	0	0	5.6	12.0	1.8	18.0	0.95	825	9075	70125	0.97	
24	G	3.1	0	15	85	100	0	0	6.1	10.8	1.6	25.0	0.94	881	7929	66075	0.91	
25	H	3.1	0	70	30	100	0	0	6.4	10.2	2.2	9.5	0.91	781	8591	66385	0.95	
26	I	3.1	0	69	31	100	0	0	5.9	10.9	1.4	34.0	0.98	830	8300	53950	0.92	
27	J	3.1	0	65	35	100	0	0	6.0	11.0	2.3	11.0	0.95	786	8646	70740	0.94	
28	K	3.1	0	55	45	100	0	0	9.8	10.0	2.4	11.0	0.95	800	7600	68000	0.94	
29	L	3.1	0	20	80	100	0	0	40	15.0	1.1	2.0	0.96	812	7308	67396	0.91	

(continued)

Test No.	Steel type	Sheet thickness (mm)	Metal microstructures											Mechanical properties				
			Pearlite (area %)	Ferrite (area %)	Bainite (area %)	Ferrite + bainite (area%)	Martensite (area %)	Retained γ (area %)	TiN average circle-equivalent diameter (mm)	TiN average minimum distance (mm)	Average circle-equivalent diameter of microscopic Ti precipitates (nm)	Number density of microscopic Ti precipitates ($\times 10^{16}$/cm^3)	Nano hardness standard deviation (GPa)	TS (MPa)	TS×u-EL (MPa×%)	TS×λ (MPa×%)	Equivalent plastic strain	
30	M *	29	0	3 *	97 *	100	0	0	5.8	11.0	-	- *	0.95	896	6272	62720	0.90	Comparative example
31	N *	29	0	95 *	5 *	100	0	0	5.5	11.5	16.0	0.01 *	0.99	478	10516	56404	1.00	
32	O *	29	0	68	30	98	2	0	5.8	10.9	2.0	17.0	1.00	850	8330	55250	0.90	
33	P *	29	0	0 *	98 *	98	2	0	5.6	11.0	-	- *	0.92	546	10374	51870	1.00	
34	Q *	Inapplicable to rolling due to slab cracking																
35	R *	29	0	85 *	15 *	100	0	0	6.7	11.0	3.0	1.6	0.92	588	8232	55860	0.91	
36	S *	29	0	32	68	100	0	0	10.5 *	8.9 *	2.0	5.0	0.97	782	8211	66470	0.79	
37	T *	29	0	55	45	100	0	0	10.2 *	6.8 *	3.0	41.0	0.99	803	7869	60225	0.71	
38	U *	29	0	30	70	100	0	0	3.6	16.0	1.0	0.03 *	0.95	696	9744	62640	1.00	
39	V	25	0	30	70	100	0	0	5.8	10.2	1.3	38.0	0.98	825	8085	66000	0.96	Inventive example
40	W	25	0	32	68	100	0	0	6.1	11.6	2.0	11.0	0.95	836	7858	68552	0.95	
41	X	25	0	25	75	100	0	0	6.3	11.8	1.6	24.0	0.96	844	7680	71740	0.96	
42	Y	25	0	38	62	100	0	0	6.2	12.0	1.5	30.0	0.93	811	7867	65691	0.98	
43	Z	25	0	29	71	100	0	0	5.9	10.6	2.0	10.0	0.95	854	7686	70882	0.97	
44	a	2.5	0	40	60	100	0	0	5.5	10.8	2.2	9.0	0.93	782	8055	66470	1.00	
45	b	25	0	38	62	100	0	0	4.8	10.5	2.4	7.0	0.95	788	8038	69344	0.99	
46	c	25	0	37	63	100	0	0	5.1	10.2	2.0	12.0	0.94	796	8438	68456	0.99	

* indicates out of the definition of the present invention

[Metal microstructure]

[0135] The present invention observed metal microstructures of the resultant hot rolled steel sheet and measured the area fraction of each of the microstructures. Specifically, when a width and a thickness of the steel sheet in a cross section perpendicular to a rolling direction of the steel sheet are defined as W and t, respectively, a specimen for metal microstructure observation was cut out at a position 1/4W from an end face of the steel sheet and 1/4t from a surface of the steel sheet.

[0136] Then, a rolling direction cross section (so-called L-direction cross section) of the specimen was subjected to Nital etching, and observed in a 300 μm × 300 μm field of view using an optical microscope after the etching. Then, a resultant microstructure photograph was subjected to image analysis to determine an area fraction A of ferrite, an area fraction B of pearlite, and a total area fraction C of bainite, martensite, and retained austenite.

[0137] Next, the portion subjected to Nital etching was subjected to Lepera etching and observed in a 300 μm × 300 μm field of view using an optical microscope. Then, a resultant microstructure photograph was subjected to image analysis to calculate a total area fraction D of retained austenite and martensite. Further, a sample subjected to facing up to a depth of 1/4 sheet thickness from a normal direction of the sheet surface was used to determine a volume ratio of the retained austenite with X-ray diffraction measurement. Since the volume ratio is substantially equal to the area fraction, the volume ratio was defined as an area fraction E of the retained austenite. An area fraction of bainite was determined from a difference between the area fraction C and the area fraction D, and an area fraction of martensite was determined from a difference between the area fraction E and the area fraction D. In this way, the area fraction of each of ferrite, bainite, martensite, retained austenite, and pearlite was determined.

[0138] The area fraction of precipitation strengthened ferrite was determined in KAM method after the specimen was polished with a colloidal silica abrasive and subjected to the EBSP measurement under the following conditions: magnification of 400x, 160 × 256 μm field of view, and a measurement step of 0.5 μm, as described above.

[0139] For fine Ti precipitates, the specimen was subjected to electrolytic polishing and measured in the 3D Atom Probe Tomography to determine the circle-equivalent diameter and the number density, as described above.

[0140] For TiN, the specimen was observed at a magnification of 1000x in 20 fields of view, each field of view being 60 × 40 μm, and the average circle-equivalent diameter of TiN was determined by image processing, as described above. Further, the minimum distance between TiN precipitates was determined by observing the same portions as in the microstructure studies using a metallurgical microscope at a magnification of 500x.

[Mechanical properties]

[0141] Among mechanical properties, tensile strength properties (tensile strength (TS), uniform elongation (u-EL), and hole expansion ratio (λ)) were evaluated in conformity with JIS Z 2241 (2011) using a JIS Z 2241 (2011) No. 5 specimen, which was taken at a position 1/4W or 3/4W from one end of the sheet in a sheet width direction when a sheet width is defined as W with a direction (width direction) perpendicular to a rolling direction being a longitudinal direction. The hole expansion ratio was evaluated in conformity with a test method described in JIS Z 2256 2010 using a specimen taken at a position similar to the position where the tensile test specimen was taken.

[0142] Further, the present inventors conducted a simple shear test in a procedure described below, and determined the equivalent plastic strain based on the results.

[0143] A specimen for the simple shear test is taken at a position 1/4W or 3/4W from one end of the sheet in a sheet width direction when a sheet width is defined as W with a direction (width direction) perpendicular to a rolling direction being a longitudinal direction. Figure 1(a) illustrates an example of the specimen. The specimen for the simple shear test illustrated in Figure 1(a) was processed into a rectangular specimen of 23 mm in the width direction of the steel sheet and 38 mm in the rolling direction of the steel sheet in such a way that both sides were uniformly polished to a sheet thickness of 2.0 mm for uniform sheet thickness.

[0144] Chucks were applied to opposite chucking portions 2 on long sides (rolling direction) of the specimen, each chucking portion having 10 mm along a short side direction (width direction), so that a shear width (shear deformation generation portion 1) of 3 mm is provided in the center of the specimen. In the case in which the sheet thickness is less than 2.0 mm, the test was conducted with the sheet thickness being left intact without polishing. Further, the center of the specimen was marked with a straight line in the short side direction (width direction) with a pen or the like.

[0145] Then, the chucked long sides were moved in opposite directions along the long side direction (rolling direction) so that the specimen was subjected to shear deformation by loading the specimen with a shearing stress σs. Figure 1(b) illustrates an example of the specimen subjected to shear deformation. The shearing stress σs is a nominal stress as determined in the following formula:

$$\text{shearing stress } \sigma s = \text{shear force}/(\text{length of specimen in rolling direction of steel sheet} \times \text{sheet thickness of specimen})$$

[0146] Since the length and the sheet thickness are invariable in the shear test, it can be considered that the shear nominal stress is nearly equal to the shear true stress. During the shear test, a CCD camera was used to capture the straight line drawn in the center of the specimen and the inclination θ of the line was measured (see Figure 1(b)). From the inclination θ, a shear strain εs, which was generated due to the shear deformation, was determined using the following formula:

$$\text{shear strain } \varepsilon s = \tan(\theta)$$

[0147] For the simple shear test, a simple shear tester (maximum displacement 8 mm) was used. Accordingly, there is a limitation to the stroke (displacement) of the tester. Further, since cracks may be generated on an end or a chucked portion of the specimen, only one shear test may not complete the test until the specimen ruptures in some cases. As such, a "multi-stage shear test" method, in which a series of operations including application of a shear test load, removal of the load, cutting of an end of a chucked portion of the specimen in a straight line, and reapplication of a load were repeated, was applied as described above.

[0148] To evaluate a one continuous simple shear test result by connecting results of these multi-stage shear test in series, a shear plastic strain (εsp) was determined as follows by subtracting an elastic shear strain (εse) taking an elastic shear modulus into consideration from a shear strain (εs) obtained in each stage of the shear test, such that the shear plastic strains (εs) in every stages were connected into one:

$$\text{shear plastic strain } \varepsilon sp = \text{shear strain } \varepsilon s - \text{elastic shear strain } \varepsilon se$$

$$\text{elastic shear strain } \varepsilon se = \sigma s/G$$

where

σs: elastic shearing stress
G: shear modulus
Here, G = E/2(1 + v) was nearly equal to 78000 (MPa).
E (Young's modulus (modulus of longitudinal elasticity)) = 206000 (MPa)
Poisson's ratio (v) = 0.3

[0149] The simple shear test was conducted until the specimen ruptures. In this way, it is possible to trace a relation between the shearing stress σs and the shear plastic strain εsp. Then, a shear plastic strain when the specimen ruptures is εspf.

[0150] From a relation between the shearing stress σs obtained in the simple shear test and the shear plastic strain εspf when the specimen ruptures, a conversion factor κ is used to determine the equivalent plastic strain εeq in the above-described method.

[0151] Next, the standard deviation of nano hardness was measured. The specimen for the metal microstructure observation was polished again. The specimen was measured in measurement areas of 25 μm $\times$ 25 μm each at an interval of 5 μm at a 1/4 depth position (1/4t portion) of sheet thickness t from a steel sheet surface in a cross section in parallel to the rolling direction under a load of 1 mN (loading 10s and unloading 10s). From the results, an average nano hardness value and a standard deviation of nano hardness were calculated. The nano hardness was measured with the use of TriboScope/TriboIndenter available from Hysitron.

[0152] The measurement results are also shown in Table 3.

[0153] As can be clearly seen from Table 3, according to the hot rolled steel sheet according to the present invention, a hot-rolled steel sheet having balanced properties, which has a tensile strength (TS) of 780 MPa or more, a product (TS $\times$ u-EL) of a uniform elongation u-EL and the tensile strength TS being equal to 7000 MPa·% or more, and a product (TS $\times$ λ) of a hole expansion ratio λ and the tensile strength TS being equal to 50000 MPa·% or more, can be obtained. Further, the hot rolled steel sheet according to the present invention has an equivalent plastic strain of more than 0.90 (90%), and it has been confirmed that the steel sheet can endure in high strain range working such as sheet metal forging.

INDUSTRIAL APPLICABILITY

[0154] According to the present invention, a hot rolled steel sheet with excellent sheet forgeability maintaining a good hole expandability, which is a basic feature as a high burring steel, can be obtained. Accordingly, the hot rolled steel sheet according to the present invention can find broad application in machine parts and the like. In particular, when it is applied to working on steel sheets including working in a high strain range such as sheet metal forging, remarkable effects thereof can be achieved.

REFERENCE SIGNS LIST

[0155]

1 shear deformation generation portion

2 chucking portions

Claims

1. A hot rolled steel sheet having a chemical composition consisting of, in mass %,
C: 0.020 to 0.070%,
Si: 0.05 to 1.70%,
Mn: 0.60 to 2.50%,
Al: 0.010 to 1.000%,
N: more than 0% to 0.0030% or less,
P: 0.050% or less,
S: 0.005% or less,
Ti: 0.015 to 0.170%,
Nb: 0 to 0.100%,
V: 0 to 0.300%,
Cu: 0 to 2.00%,
Ni: 0 to 2.00%,
Cr: 0 to 2.00%,
Mo: 0 to 1.00%,
B: 0 to 0.0100%,
Mg: 0 to 0.0100%,
Ca: 0 to 0.0100%,
REM: 0 to 0.1000%,
Zr: 0 to 1.000%,
Co: 0 to 1.000%,
Zn: 0 to 1.000%,
W: 0 to 1.000%,
Sn: 0 to 0.050%, and,
the balance: Fe and impurities, wherein
when a width and a thickness of the steel sheet in a cross section perpendicular to a rolling direction of the steel sheet are defined as W and t, respectively, a metal microstructure includes, in area %, at a position 1/4W or 3/4W from an end face of the steel sheet and 1/4t or 3/4t from a surface of the steel sheet,
ferrite: 5 to 70%,
bainite: 30 to 95%,
retained austenite: 2% or less,
martensite: 2% or less, and,
pearlite: 1% or less, and
a total of ferrite and bainite: 95% or more, and wherein
the ferrite contains precipitates including Ti in grains of the ferrite,
a number density of the precipitates including Ti is 1.0×10^{16} to $50.0 \times 10^{16}/cm^3$, the steel sheet includes TiN precipitates,
an average circle-equivalent diameter of the TiN precipitates is 1.0 to 10.0 μm,
an average of minimum distances between adjacent TiN precipitates is 10.0 μm or more, and

a standard deviation of nano hardness is 1.00 GPa or less.

2. The hot rolled steel sheet according to claim 1, wherein
an average circle-equivalent diameter of the precipitates including Ti is 1.00 to 3.00 nm.

3. The hot rolled steel sheet according to claim 1 or 2, wherein
a tensile strength is 780 MPa or more,
a product of a uniform elongation and a tensile strength is 7000 MPa·% or more, and
a product of a hole expansion ratio and a tensile strength is 50000 MPa·% or more.

4. A method of producing the hot rolled steel sheet according to any of claims 1 to 3, comprising:

subjecting a slab having the chemical composition according to claim 1 to a heating process, a continuous hot rolling process, a first cooling process, a second cooling process, and a coiling process, in this order, wherein in the heating process, the slab is heated to a temperature of SRTmin°C or more to 1260°C or less, the SRTmin being represented by a formula (i) below,
the continuous hot rolling process includes rough rolling and multi-stand finish rolling of three stands or more,
an end temperature of the rough rolling is 1100°C or more,
a cumulative strain of rolling at final three stands in the multi-stand finish rolling is 0.01 to 0.10,
a rolling end temperature of the multi-stand finish rolling is at a temperature of Ar_3 + 30°C or more, the Ar_3 being determined by a formula (ii) below,
in the first cooling process, cooling is started after 1.00 to 5.00 seconds after completion of the multi-stand finish rolling, the cooling is continued at an average cooling rate of 10°C/sec or more from the rolling end temperature down to a temperature range of 650 to 750°C, and thereafter the sheet is held in air for 1 to 10 seconds,
in the second cooling process, after the sheet is held in air, cooling is conducted at an average cooling rate of 10°C/sec or more from a temperature range of 600 to 740°C, and
in the coiling process, the sheet is coiled at a coiling temperature of 450 to 650°C:

$$SRTmin = 7000/\{2.75 - \log(Ti \times C)\} - 273 \ldots (i)$$

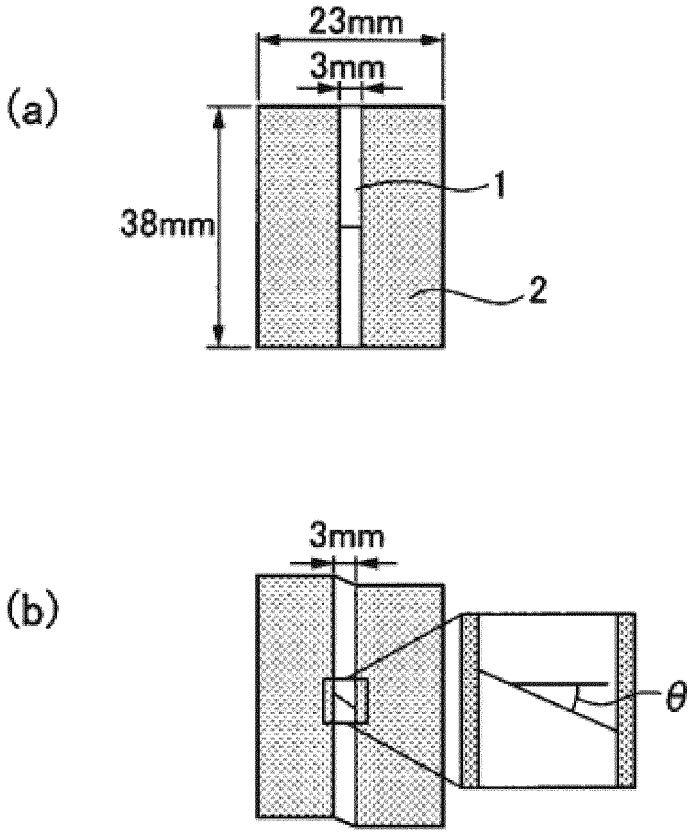
$$Ar_3 = 970 - 325 \times C + 33 \times Si + 287 \times P + 40 \times Al - 92 \times (Mn + Mo + Cu) - 46 \times (Cr + Ni) \ldots (ii)$$

where a symbol of an element in the above formula represents a content (in mass %) of the element in the hot rolled steel sheet and is substituted by zero when the element is not contained.

5. A steel forged part obtained from the hot rolled steel sheet according to any of claims 1 to 3.

6. A method of producing a steel forged part, wherein the hot rolled steel sheet according to any of claims 1 to 3 is subjected to at least forging.

FIGURE 1



INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2017/013747

A. CLASSIFICATION OF SUBJECT MATTER
C22C38/00(2006.01)i, *C21D9/46*(2006.01)i, *C22C38/58*(2006.01)i

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, C21D9/46, C22C38/58

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 1996-2017
Kokai Jitsuyo Shinan Koho 1971-2017 Toroku Jitsuyo Shinan Koho 1994-2017

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.
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A	JP 2011-12308 A (Nippon Steel Corp.), 20 January 2011 (20.01.2011), (Family: none)	1-6
A	WO 2011/135997 A1 (Sumitomo Metal Industries, Ltd.), 03 November 2011 (03.11.2011), & US 2013/0098515 A1 & EP 2565288 A1 & CN 102959119 A & KR 10-2013-0008622 A	1-6
A	JP 2002-161340 A (Nippon Steel Corp.), 04 June 2002 (04.06.2002), (Family: none)	1-6

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

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Date of the actual completion of the international search	Date of mailing of the international search report
23 June 2017 (23.06.17)	04 July 2017 (04.07.17)
Name and mailing address of the ISA/ Japan Patent Office 3-4-3,Kasumigaseki,Chiyoda-ku, Tokyo 100-8915,Japan	Authorized officer Telephone No.

INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2017/013747

C (Continuation). DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
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Form PCT/ISA/210 (continuation of second sheet) (January 2015)

REFERENCES CITED IN THE DESCRIPTION

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- JP 2002322540 A **[0004]**
- JP 2002322541 A **[0004]**