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(54) HIGH-STRENGTH LINE PIPE STEEL MATERIAL HAVING EXCELLENT FRACTURE TOUGHNESS IN HYDROGEN, METHOD FOR MANUFACTURING SAME, STEEL TUBE FOR HIGH-STRENGTH LINE PIPES, AND METHOD FOR MANUFACTURING SAME

(57) To provide a steel material for a high-strength line pipe with high fracture toughness in hydrogen in a high-pressure hydrogen gas environment, a method for producing the steel material, a steel pipe for a high-strength line pipe, and a method for producing the steel pipe, suitable for a steel structure used in a high-pressure hydrogen gas environment, such as a line pipe for 100% hydrogen gas or a natural gas containing hydrogen at a hydrogen partial pressure of 1 MPa or more (natural gas is a gas containing hydrocarbons, such as methane and ethane, as main components).

The steel material for a high-strength line pipe with high fracture toughness in hydrogen has a specific composition and a specific microstructure, has a tensile strength of 520 MPa or more, and has a hydrogen-induced crack growth threshold K_{IH} of 80 MPa $\cdot m^{1/2}$ or more in a high-pressure hydrogen gas environment of 1 MPa or more.

EP 4 578 981 A1

Description

Technical Field

[0001] The present invention relates to a steel material for a high-strength line pipe with high fracture toughness in hydrogen in a high-pressure hydrogen gas environment of 1 MPa or more, a method for producing the steel material, a steel pipe for a high-strength line pipe, and a method for producing the steel pipe, suitable for applications, such as a line pipe for transporting hydrogen gas.

Background Art

[0002] As an existing energy infrastructure, there is a line pipe for transporting crude oil, natural gas, or the like. Such a steel structure is used in an atmosphere containing hydrogen sulfide, and the occurrence of hydrogen embrittlement, such as hydrogen-induced cracking (HIC) or sulfide stress corrosion cracking (SSCC), has become a safety problem, and suppression thereof has been required. To prevent the occurrence of hydrogen embrittlement, such as hydrogen-induced cracking or sulfide stress corrosion cracking, various measures have been taken, such as reducing amount of MnS as an origin of a crack in a steel material, suppressing accumulation of carbonitride or oxide of Ti or Nb, or suppressing segregation of hardened phase of center segregation. From the perspective of suppressing the occurrence of origin of a crack by improving the corrosion resistance of a steel material, addition of Sn or Sb to the steel material has been proposed (for example, Patent Literature 1 and Patent Literature 2).

[0003] In recent years, utilization of hydrogen has been promoted as a clean energy source for the purpose of building a decarbonizing society. Thus, for the purpose of transporting a large amount of hydrogen gas, construction of a hydrogen gas transportation network that pressure-feeds natural gas mixed with a certain percentage of hydrogen or hydrogen gas as an alternative through a natural gas line pipe has been studied. The transport pressure in such a pipeline operation is assumed to be a high pressure in the range of 1 to 40 MPa, and a line pipe is exposed to a high-pressure hydrogen gas environment. A steel material used in such an environment needs to have hydrogen resistance required in a hydrogen gas environment in addition to characteristics required in an existing sour environment, that is, suppression of occurrence of corrosion on an inner surface of a steel pipe and reduction of accumulation of hydrogen in the material.

[0004] Austenite stainless steel, such as SUS 316L, which exhibits fracture toughness in hydrogen, is used for a steel structure used in a high-pressure hydrogen gas environment. However, such a steel material is expensive, has low strength, and when designed to be able to withstand a high hydrogen pressure, has a large wall thickness, resulting in a very expensive line pipe and, therefore, not suitable for laying a pipeline. Thus, there has been a demand for a steel material that can withstand a high-pressure hydrogen gas environment at a lower cost for use in a hydrogen line pipe.

[0005] To solve the above problems, for example, Patent Literature 3 proposes an austenitic steel material with a high Mn content. According to the technique described in Patent Literature 3, it is possible to provide a steel material at lower cost than austenite stainless steel, but the steel material is austenitic and is therefore higher in cost than low-alloy steel. Furthermore, suppression of pitting corrosion as an origin of a hydrogen-induced, such as HIC resistance or SSCC resistance, is not considered.

[0006] Furthermore, in a pipeline, the operation is repeatedly started and shut down, applying repeated stress to a line pipe. Thus, when designing a steel structure, such as a pipeline, it is essential to consider fatigue fracture. The breaking point of fatigue fracture of a steel structure used in a high-pressure hydrogen gas environment corresponds to the critical crack length calculated from the operating conditions of the pipeline and the hydrogen-induced crack growth threshold K_{IH} corresponding to the fracture toughness value of the steel material in hydrogen gas. From the perspective of extending the life and improving the safety of a structure for hydrogen, increasing the K_{IH} of a steel material is considered to be an effective guide.

[0007] A hydrogen pipeline is assumed to use a line pipe with a weld, that is, a welded metal portion, and a heat-affected zone [HAZ]. Patent Literature 4 proposes a method for producing a steel material with high K_{IH} but does not refer to the characteristics of a weld. In general, a weld is more susceptible to characteristic degradation due to hydrogen than a base material. Thus, it is important to improve the K_{IH} also for a weld.

[0008] To increase the K_{IH} of a steel material, for example, it is better to decrease upper bainite containing a coarse carbide.

Citation List

Patent Literature

[0009]

PTL 1: Japanese Unexamined Patent Application Publication No. 2011-26695
PTL 2: Japanese Unexamined Patent Application Publication No. 2010-209461
PTL 3: PCT Japanese Translation Patent Publication No. 2019-505675
PTL 4: International Publication No. WO 2017/047099 Non Patent Literature

[0010] NPL 1: (written by) The Japan Society for Heat Treatment, Introduction: Microstructure and Properties of Metallic Materials - Heat Treatment and Microstructure Controlling for Materials, 2004

Summary of Invention

Technical Problem

[0011] The present invention has been made to solve these known problems and aims to provide a steel material for a high-strength line pipe with high fracture toughness in hydrogen in a high-pressure hydrogen gas environment, a method for producing the steel material, a steel pipe for a high-strength line pipe, and a method for producing the steel pipe, suitable for a steel structure used in a high-pressure hydrogen gas environment, such as a line pipe for 100% hydrogen gas or a natural gas containing hydrogen at a hydrogen partial pressure of 1 MPa or more (natural gas is a gas containing hydrocarbons, such as methane and ethane, as main components). The high-pressure hydrogen gas environment is assumed to be a high-pressure hydrogen gas of 1 MPa or more or an environment containing 0.2% or more of hydrogen gas.

[0012] The phrase "high fracture toughness in hydrogen in a high-pressure hydrogen gas environment", as used herein, refers to a hydrogen-induced crack growth threshold K_{IH} of 80 MPa·$m^{1/2}$ or more, as determined by a fracture toughness test at room temperature (20°C ± 10°C) in both environments of hydrogen gas with a pressure of 1 MPa or more and a natural gas (the main components are hydrocarbons, such as methane and ethane) mixed atmosphere containing hydrogen at a hydrogen partial pressure of 1 MPa or more. The fracture toughness value refers to a value determined by a fracture toughness test according to ASTM E399, ASTM E1820, and ASTM E1681. The natural gas containing hydrogen with a hydrogen partial pressure of 1 MPa or more, for example, has a hydrogen concentration of 30% or less by volume and a pressure of the entire gas of 30 MPa or less.

[0013] The term "steel material", as used herein, includes a steel sheet, a steel plate, a seamless steel pipe, an electric-resistance-welded steel pipe, a shaped steel, a steel bar, and the like.

Solution to Problem

[0014] For the purpose of suppressing hydrogen absorption into a steel material, which is a fundamental factor of hydrogen embrittlement, the present inventors have made technical studies on conditions to be satisfied by the steel material to produce a steel material for a high-strength line pipe and a steel pipe for a high-strength line pipe with high fracture toughness in hydrogen in a high-pressure hydrogen gas environment. As a result, it has been found that the hydrogen-induced crack growth threshold K_{IH} of a steel material or a steel pipe is improved in a metallic microstructure in which the number of inclusions with an aspect ratio of 2.0 or more and a length of 10 μm or more is 15 pieces/100 mm^2 or less and the bainite in the range from a surface of the steel material or the steel pipe to the middle of the sheet thickness has a maximum grain size of 25 μm or less. It has also been found that the hydrogen-induced crack growth threshold K_{IH} of a steel material is further improved when an area fraction of retained austenite is 0% to 3% and an area fraction of bainite in the range from a surface of the steel material or a steel pipe to the middle of the sheet thickness is 90% or more. To realize such a steel microstructure, it is necessary to strictly control the rolling conditions in a hot rolling step and the cooling conditions after rolling, and these conditions have been successfully discovered. The present invention is based on these findings. In the present invention, the term "high strength" refers to a tensile strength of 520 MPa or more.

[0015] Thus, the gist of the present invention is as follows:

[1] A steel material for a high-strength line pipe with high fracture toughness in hydrogen, the steel material including: a chemical composition containing:

on a mass percent basis,
C: 0.02% to 0.15%,
Si: 0.01% to 2.0%,
Mn: 0.5% to 1.5%,
P: 0.0001% to 0.015%,
S: 0.0002% to 0.0015%,
Al: 0.005% to 0.15%,

O: 0.01% or less,
N: 0.010% or less,
Nb: 0.10% or less, and
H: 0.02 ppm or less, and
optionally at least one selected from
Ca: 0% to 0.005%,
Ni: 0% to 2.0%,
Ti: 0% to 0.1%,
Cu: 0% to 1.0%,
Cr: 0% to 1.0%,
Mo: 0% to 0.60%,
W: 0% to 1.0%,
V: 0% to 0.10%,
Zr: 0% to 0.050%,
Mg: 0% to 0.01%,
REM: 0% to 0.01%,
B: 0% to 0.0020%,
Ta: 0% to 0.2%,
Hf: 0% to 0.2%,
Re: 0% to 0.005%,
Sn: 0% to 0.3%, and
Sb: 0% to 0.3%,
the remainder being Fe and an incidental impurity element,
a metallic microstructure containing bainite and 15 pieces/100 mm^2 or less of inclusions having an aspect ratio of 2.0 or more and a length of 10 μm or more, the bainite in a range from a surface to a middle of a thickness of the steel material having a maximum grain size of 25 μm or less,
tensile strength of 520 MPa or more, and
a hydrogen-induced crack growth threshold K_{IH} in a high-pressure hydrogen gas environment of 1 MPa or more of 80 $MPa \cdot m^{1/2}$ or more.

[2] The steel material for a high-strength line pipe with high fracture toughness in hydrogen according to [1], wherein the chemical composition contains, on a mass percent basis,

Ca: 0.0001% to 0.005%,
Ni: 0.01% to 2.0%,
Ti: 0.005% to 0.1%,
Cu: 0.01% to 1.0%,
Cr: 0.01% to 1.0%,
Mo: 0.01% to 0.60%,
W: 0.01% to 1.0%,
V: 0.01% to 0.10%,
Zr: 0.0001% to 0.050%,
Mg: 0.0001% to 0.01%,
REM: 0.0001% to 0.01%,
B: 0.0001% to 0.0020%,
Ta: 0.0001% to 0.2%,
Hf: 0.0001% to 0.2%,
Re: 0.0001% to 0.005%,
Sn: 0.0001% to 0.3%, and
Sb: 0.0001% to 0.3%.

[3] The steel material for a high-strength line pipe with high fracture toughness in hydrogen according to [1] or [2], wherein an area fraction of retained austenite is 0% to 3% by area, and an area fraction of the bainite in the range from the surface to the middle of the thickness of the steel material is 90% or more.

[4] A method for producing a steel material for a high-strength line pipe with high fracture toughness in hydrogen, the method including:

a heating step of heating a cast steel having the chemical composition according to [1] or [2] at 1000°C to 1250°C;

a hot rolling step of rolling the cast steel heated in the heating step under conditions in which a total rolling reduction in a recrystallization temperature range is 35% or more and 55% or less, a rolling reduction in a final rolling pass in the recrystallization temperature range is 10% or more, a rolling reduction in a final rolling pass at (recrystallization temperature - 80°C) or more is 15% or more, and a finish rolling temperature is an Ar_3 transformation point or higher in terms of a temperature at a surface of a steel sheet; and
a controlled cooling step of cooling a hot-rolled steel sheet produced in the hot rolling step under conditions in which a cooling start temperature is the Ar_3 transformation point or higher in terms of a temperature at the surface of the hot-rolled steel sheet, a cooling start time difference between a front end and a rear end of the hot-rolled steel sheet is 50 seconds or less, an average cooling rate from 750°C to 550°C ranges from 15°C/s to 50°C/s in terms of a temperature at a middle of a thickness of the steel sheet, and a cooling stop temperature ranges from 250°C to 650°C.

[5] A steel pipe for a high-strength line pipe with high fracture toughness in hydrogen, the steel pipe including:
a chemical composition containing:

on a mass percent basis,
C: 0.02% to 0.15%,
Si: 0.01% to 2.0%,
Mn: 0.5% to 1.5%,
P: 0.0001% to 0.015%,
S: 0.0002% to 0.0015%,
Al: 0.005% to 0.15%,
O: 0.01% or less,
N: 0.010% or less,
Nb: 0.10% or less, and
H: 0.02 ppm or less, and
optionally at least one selected from
Ca: 0% to 0.005%,
Ni: 0% to 2.0%,
Ti: 0% to 0.1%,
Cu: 0% to 1.0%,
Cr: 0% to 1.0%,
Mo: 0% to 0.60%,
W: 0% to 1.0%,
V: 0% to 0.10%,
Zr: 0% to 0.050%,
Mg: 0% to 0.01%,
REM: 0% to 0.01%,
B: 0% to 0.0020%,
Ta: 0% to 0.2%,
Hf: 0% to 0.2%,
Re: 0% to 0.005%,
Sn: 0% to 0.3%, and
Sb: 0% to 0.3%,
the remainder being Fe and an incidental impurity element,
a metallic microstructure containing bainite and 15 pieces/100 mm^2 or less of inclusions having an aspect ratio of 2.0 or more and a length of 10 μm or more, the bainite in a range from an inner surface to a middle of a thickness of the steel pipe having a maximum grain size of 25 μm or less,
tensile strength of 520 MPa or more, and
a hydrogen-induced crack growth threshold K_{IH} in a high-pressure hydrogen gas environment of 1 MPa or more of 80 $MPa \cdot m^{1/2}$ or more.

[6] The steel pipe for a high-strength line pipe with high fracture toughness in hydrogen according to [5], wherein the chemical composition contains, on a mass percent basis,

Ca: 0.0001% to 0.005%,
Ni: 0.01% to 2.0%,
Ti: 0.005% to 0.1%,

Cu: 0.01% to 1.0%,
Cr: 0.01% to 1.0%,
Mo: 0.01% to 0.60%,
W: 0.01% to 1.0%,
V: 0.01% to 0.10%,
Zr: 0.0001% to 0.050%,
Mg: 0.0001% to 0.01%,
REM: 0.0001% to 0.01%,
B: 0.0001% to 0.0020%,
Ta: 0.0001% to 0.2%,
Hf: 0.0001% to 0.2%,
Re: 0.0001% to 0.005%,
Sn: 0.0001% to 0.3%, and
Sb: 0.0001% to 0.3%.

[7] The steel pipe for a high-strength line pipe with high fracture toughness in hydrogen according to [5] or [6], wherein an area fraction of retained austenite is 0% to 3%, and an area fraction of the bainite in the range from the inner surface to the middle of the thickness of the steel pipe is 90% or more.

[8] A method for producing a steel pipe for a high-strength line pipe with high fracture toughness in hydrogen, the method including:

a heating step of heating a cast steel having the chemical composition according to [5] or [6] at 1000°C to 1250°C;
a hot rolling step of rolling the cast steel heated in the heating step under conditions in which a total rolling reduction in a recrystallization temperature range is 35% or more and 55% or less, a rolling reduction in a final rolling pass in the recrystallization temperature range is 10% or more, a rolling reduction in a final rolling pass at (recrystallization temperature - 80°C) or more is 15% or more, and a finish rolling temperature is an Ar_3 transformation point or higher in terms of a temperature of a surface of a steel sheet;
a controlled cooling step of cooling a hot-rolled steel sheet produced in the hot rolling step under conditions in which a cooling start temperature is the Ar_3 transformation point or higher in terms of a temperature at a surface of the hot-rolled steel sheet, a cooling start time difference between a front end and a rear end of the hot-rolled steel sheet is 50 seconds or less, an average cooling rate from 750°C to 550°C ranges from 15°C/s to 50°C/s in terms of a temperature at a middle of a thickness of the steel sheet, and a cooling stop temperature ranges from 250°C to 650°C; and
any one of a pipe production step of bending the hot-rolled steel sheet and butt-welding both end portions thereof after the controlled cooling step and a pipe production step of forming the hot-rolled steel sheet into a cylindrical shape by cold roll forming and subjecting both circumferential end portions of the cylindrical shape to butt electric resistance welding after the controlled cooling step.

Advantageous Effects of Invention

[0016] The present invention can easily and simply produce a steel material with considerably improved fracture toughness in hydrogen in a high-pressure hydrogen gas environment and exhibits industrially significant effects. The present invention also has the effects of significantly improving the hydrogen absorption resistance of a steel structure, such as a high-pressure hydrogen gas line pipe, and greatly contributing to the improvement of the safety of the steel structure.

Description of Embodiments

[0017] Next, a method for implementing the present invention is more specifically described.

[0018] A steel material is more specifically described as a first embodiment, a UOE steel pipe as an example of a steel pipe according to the present invention is more specifically described as a second embodiment, and an electric-resistance-welded steel pipe as an example of a steel pipe according to the present invention is more specifically described as a third embodiment.

First Embodiment

[Chemical Composition]

[0019] The reasons for limiting base material components in a steel material according to the present invention are described below. Unless otherwise specified, the unit "%" in the following description refers to "% by mass".

C: 0.02% to 0.15%

[0020] C effectively contributes to the improvement of strength, but the strength cannot be sufficient at a C content of less than 0.02%, so that the C content is 0.02% or more. Preferably, the C content is 0.03% or more. More preferably, the C content is 0.035% or more. Still more preferably, the C content is 0.04% or more. On the other hand, more than 0.15% results in low weldability. Thus, the C content is limited to 0.15% or less. Preferably, the C content is 0.10% or less. More than 0.08% may result in a decrease in SSCC resistance and HIC resistance due to an increase in the hardness of a surface layer portion or a center segregation zone during controlled cooling. Furthermore, toughness also deteriorates. Thus, the C content is more preferably 0.08% or less. Still more preferably, the C content is 0.06% or less.

Si: 0.01% to 2.0%

[0021] Si is contained for deoxidization, but the deoxidation effect is not sufficient at a content of less than 0.01%, so that the Si content is 0.01% or more. The Si content is preferably 0.02% or more. More preferably, the Si content is 0.05% or more. Still more preferably, the Si content is 0.08% or more. The effect is observed up to 2.0%, and the Si content is therefore 2.0% or less. The Si content is preferably 1.8% or less, more preferably 1.5% or less. The Si content is still more preferably 1.0% or less. However, more than 0.5% sometimes results in lower toughness or weldability, and the Si content is therefore most preferably 0.5% or less.

Mn: 0.5% to 1.5%

[0022] Mn effectively contributes to the improvement of strength and toughness, but the effect of containing Mn is insufficient at a content of less than 0.5%, so that the Mn content is 0.5% or more. The Mn content is preferably 0.6% or more, more preferably 0.8% or more. Still more preferably, the Mn content is 1.0% or more. On the other hand, more than 1.5% results in lower SSCC resistance and HIC resistance due to an increase in the hardness of a surface layer portion or a center segregation zone during controlled cooling. Furthermore, weldability also deteriorates. Thus, the Mn content is limited to 1.5% or less. Preferably, the Mn content is 1.4% or less. The Mn content is more preferably 1.3% or less, still more preferably 1.2% or less.

P: 0.0001% to 0.015%

[0023] P is an incidental impurity element, reduces weldability, and reduces the HIC resistance due to an increase in the hardness of a center segregation zone. This tendency becomes remarkable at more than 0.015%, and the P content is therefore limited to 0.015% or less. The P content is preferably 0.012% or less, more preferably 0.010% or less. Still more preferably, the P content is 0.008% or less. Although a lower P content is better, from the perspective of refining costs, the P content is 0.0001% or more.

S: 0.0002% to 0.0015%

[0024] S is an incidental impurity element, forms a MnS inclusion in steel, and reduces the HIC resistance, so that a lower S content is preferred, but 0.0015% or less is allowable. Thus, the S content is 0.0015% or less. The S content is preferably 0.0010% or less, more preferably 0.0008% or less. Although a lower S content is better, from the perspective of refining costs, the S content is 0.0002% or more.

Al: 0.005% to 0.15%

[0025] Al is added as a deoxidizing agent, and the effect of containing Al is insufficient at less than 0.005%, so that the Al content is 0.005% or more. On the other hand, more than 0.15% results in steel with lower cleanliness and toughness, so that the Al content is 0.15% or less. The Al content is preferably 0.12% or less, more preferably 0.10% or less. Still more preferably, the Al content is 0.08% or less.

O: 0.01% or less

[0026] O causes the formation of an oxide inclusion, and the O content is therefore preferably as small as possible. This influence does not become a problem at an O content of 0.01% or less, and the O content is therefore 0.01% or less. The O content is preferably 0.0080% or less. More preferably, the O content is less than 0.0030%. The lower limit may be, but is not limited to, 0.0005% or more.

N: 0.010% or less

[0027] N effectively contributes to the improvement of the strength, but a content of more than 0.010% results in an increase in the hardness during controlled cooling and lower toughness. Thus, the N content is 0.010% or less. The N content is preferably 0.008% or less, more preferably 0.006% or less, still more preferably 0.004% or less. However, sufficient strength cannot be ensured at less than 0.00001%, and an excessive decrease increases the steelmaking cost. Thus, the content is preferably 0.00001% or more. More preferably, the N content is 0.002% or more.

Nb: 0.10% or less

[0028] Nb is an element effective in increasing the strength and toughness of a steel material. The effects of containing Nb are insufficient at a content of less than 0.001%, and 0.001% or more is therefore preferred. On the other hand, more than 0.10% results in a weld with lower toughness, and the Nb content is therefore 0.10% or less. The Nb content is preferably 0.095% or less. The Nb content is more preferably 0.090% or less, still more preferably 0.085% or less. The Nb content is most preferably 0.080% or less.

H: 0.02 ppm or less

[0029] H may be introduced into a steel material in various steps during production, and a large amount of H introduced may increase the risk of cracking after solidification and significantly reduce the K_{IH}. These effects do not cause a problem at 0.02 ppm or less, and the H content is therefore 0.02 ppm or less. The H content is preferably 0.015 ppm or less, more preferably 0.008 ppm or less. The H content is still more preferably 0.005 ppm or less, most preferably less than 0.002 ppm. The lower limit is preferably, but not limited to, 0.0008 ppm or more from the perspective of production cost. The H content is more preferably 0.001 ppm or more. The amount of hydrogen is the amount of residual hydrogen after forming of a steel material, a steel pipe, UOE, or the like.

[0030] The chemical composition in the present disclosure may optionally contain at least one selected from Ca, Ni, Ti, Cu, Cr, Mo, W, V, Zr, Mg, REM, B, Ta, Hf, Re, Sn, and Sb in the following range.

Ca: 0% to 0.005%

[0031] Ca is an element effective in improving the HIC resistance by the shape control of a sulfide inclusion, and when Ca is contained, the Ca content may be 0% or more, but the effect of addition is insufficient at less than 0.0001%. Thus, when Ca is contained, the Ca content is 0.0001% or more, more preferably 0.0005% or more. On the other hand, at more than 0.005%, not only the effect is saturated but also the HIC resistance decreases due to a decrease in the cleanliness of steel, so that when Ca is contained the Ca content is limited to 0.005% or less. The Ca content is preferably 0.004% or less. The Ca content is more preferably 0.002% or less, still more preferably 0.0008% or less.

Ni: 0% to 2.0%

[0032] Ni is an element effective in improving the toughness and increasing the strength, and when Ni is contained, the Ni content may be 0% or more, but to achieve these effects, it is preferable to contain 0.01% or more. The Ni content is more preferably 0.1% or more. On the other hand, to reduce the cost, when Ni is contained, the Ni content is 2.0% or less. The Ni content is preferably 1.8% or less. The Ni content is more preferably 1.4% or less, still more preferably 0.8% or less.

Ti: 0% to 0.1%

[0033] Ti contributes to an increase in the strength of a steel material, and when Ti is contained, the Ti content may be 0% or more. To achieve the effect, when Ti is contained, the content is preferably 0.005% or more, more preferably 0.008% or more. On the other hand, a content of more than 0.1% results in saturation of the effect and causes an increase in cost, so that when Ti is contained the Ti content is 0.1% or less. The Ti content is preferably 0.08% or less, more preferably 0.06% or less. To reduce the cost, the Ti content is still more preferably 0.05% or less. The Ti content is most preferably 0.04% or less.

Cu: 0% to 1.0%

[0034] Cu is an element effective in improving the toughness and increasing the strength, and when Cu is contained, the Cu content may be 0% or more, but to achieve these effects, it is preferable to contain 0.01% or more, more preferably 0.05% or more. On the other hand, an excessively high content results in lower weldability, and when Cu is contained, the Cu content is 1.0% or less. The Cu content is preferably 0.95% or less, more preferably 0.9% or less. Still more preferably, the Cu content is 0.85% or less. Most preferably, the Cu content is 0.5% or less.

Cr: 0% to 1.0%

[0035] Like Mn, Cr is an element effective in obtaining sufficient strength even at a low C content, and when Cr is contained, the Cr content may be 0% or more, but to obtain this effect, it is preferable to contain 0.01% or more, more preferably 0.05% or more. On the other hand, an excessively high content results in lower SSCC resistance due to excessive hardenability. Furthermore, weldability also deteriorates. Thus, when Cr is contained, the Cr content is 1.0% or less. The Cr content is preferably 0.95% or less. The Cr content is more preferably 0.9% or less, still more preferably 0.85% or less.

Mo: 0% to 0.60%

[0036] Mo is an element effective in improving the toughness and increasing the strength and is an element effective in improving the SSCC resistance and the HIC resistance. When Mo is contained, the Mo content may be 0% or more and is preferably 0.01% or more, more preferably 0.10% or more, to achieve the above effects. On the other hand, an excessively high content results in lower SSCC resistance due to excessive hardenability. Furthermore, weldability also deteriorates. Thus, when Mo is contained, the Mo content is 0.60% or less. The Mo content is preferably 0.50% or less, more preferably 0.40% or less, still more preferably 0.35% or less.

W: 0% to 1.0%

[0037] W contributes to an increase in the strength of a steel material. When W is contained, the W content may be 0% or more and is preferably 0.01% or more to achieve the above effect. On the other hand, a W content of more than 1.0% results in saturation of the effect and causes an increase in cost, so that when W is contained, the W content is 1.0% or less. The W content is preferably 0.9% or less, more preferably 0.8% or less. To reduce the cost, 0.5% or less is still more preferred.

V: 0% to 0.10%, Zr: 0% to 0.050%, Mg: 0% to 0.01%, and REM: 0% to 0.01%

[0038] V is an element that can be optionally contained to increase the strength and toughness of a steel material. When V is contained, the V content may be 0% or more, but the effects of containing V are not sufficient at a V content of less than 0.01%, so that the V content is preferably 0.01% or more. The V content is more preferably 0.03% or more. On the other hand, more than 0.10% results in a weld with lower toughness, so that when V is contained, 0.10% or less is preferred. The V content is preferably 0.09% or less. The V content is more preferably 0.07% or less, still more preferably 0.06% or less.

[0039] Zr, Mg, and REM are elements that can be optionally added to increase the toughness through grain refinement or to increase cracking resistance through the control of inclusion properties. When these elements are contained, each content may be 0% or more, but the effects of containing these elements are insufficient at a content of less than 0.0001%, so that each content is preferably 0.0001% or more, more preferably 0.0005% or more. More specifically, the Zr content is preferably 0.0001% or more. The Zr content is more preferably 0.0005% or more. The REM content is preferably 0.0001% or more. The REM content is more preferably 0.0005% or more. The Mg content is preferably 0.0001% or more. The Mg content is more preferably 0.0005% or more.

[0040] On the other hand, when the Zr content is more than 0.050%, and the Mg and REM contents are more than 0.01%, the effects are saturated, so that when these are contained, the Zr content is 0.050% or less, and the Mg and REM contents are 0.01% or less. More specifically, when Zr is contained, the Zr content is 0.050% or less. The Zr content is preferably 0.040% or less. The Zr content is more preferably 0.020% or less. When REM is contained, the REM content is 0.01% or less. The REM content is preferably 0.009% or less. The REM content is more preferably 0.008% or less. When Mg is contained, the Mg content is 0.01% or less. The Mg content is preferably 0.009% or less. The Mg content is more preferably 0.008% or less.

B: 0% to 0.0020%

[0041] B is an element that improves hardenability, and contributes to an increase in the strength of a steel material, suppresses coarsening of prior-austenite grains, and improves various characteristics of the material. When B is contained, the B content may be 0% or more and is preferably 0.0001% or more, more preferably 0.0008% or more, to achieve the above effects. On the other hand, a B content of more than 0.0020% results in saturation of the effect and causes an increase in cost, so that when B is contained the B content is 0.0020% or less. The B content is preferably 0.0014% or less. The B content is more preferably 0.0012% or less. To reduce the cost, 0.0010% or less is still more preferred.

Ta: 0% to 0.2%

[0042] Ta is an element that forms a carbide or a nitride and contributes to an improvement in the strength. When Ta is contained, the Ta content may be 0% or more and is preferably 0.0001% or more to achieve the above effect. More preferably, the Ta content is 0.0008% or more. On the other hand, a content of more than 0.2% sometimes results in lower toughness, so that when Ta is contained the Ta content is 0.2% or less. Ta is preferably 0.16% or less. Ta is more preferably 0.12% or less, still more preferably 0.10% or less.

Hf: 0% to 0.2%, Re: 0% to 0.005%

[0043] These elements contribute to an increase in the strength of a steel material. To achieve the above effect, when these elements are contained, each content is preferably 0.0001% or more, preferably 0.0010% or more. More specifically, when Hf is contained, the Hf content is preferably 0.0001% or more. The Hf content is more preferably 0.0010% or more. When Re is contained, the Re content is preferably 0.0001% or more. The Re content is preferably 0.001% or more. On the other hand, when these elements are contained, a Hf content of more than 0.2% or a Re content of more than 0.005% results in an increase in an oxide, and aggregation reduces the hydrogen resistance, so that the Hf content is 0.2% or less, and the Re content is 0.005% or less. More specifically, when Hf is contained, the Hf content is 0.2% or less. The Hf content is preferably 0.18% or less, more preferably 0.12% or less. When Re is contained, the Re content is 0.005% or less. The Re content is preferably 0.004% or less, more preferably 0.003% or less.

Sn: 0% to 0.3%, Sb: 0% to 0.3%

[0044] These elements contribute to an increase in the strength and an improvement in the hardenability of a steel material. When Sn and Sb are contained, the Sn and Sb contents may be 0% or more and are each preferably 0.0001% or more to achieve the effects. Preferably, it is 0.001% or more. More specifically, when Sn is contained, the Sn content may be 0% or more and is preferably 0.0001% or more. The Sn content is more preferably 0.001% or more. When Sb is contained, the Sb content may be 0% or more and is preferably 0.0001% or more. The Sb content is more preferably 0.001% or more. On the other hand, each content of more than 0.3% results in saturation of the effects and an increase in cost, so that when Sn or Sb is contained the Sn or Sb content is 0.3% or less. To reduce the cost, 0.01% or less is preferred. Thus, when Sn is contained, the Sn content is 0.3% or less. The Sn content is preferably 0.2% or less. The Sn content is more preferably 0.1% or less. The Sn content is still more preferably 0.01% or less. When Sb is contained, the Sb content is 0.3% or less. The Sb content is preferably 0.2% or less. The Sb content is more preferably 0.1% or less. The Sb content is still more preferably 0.01% or less.

[0045] In the chemical composition of a steel material, the remainder other than these components (elements) is composed of Fe and an incidental impurity element.

[0046] The metallic microstructure of a steel material according to the present invention is described below.

Metallic Microstructure

[0047] Inclusions having aspect ratio of 2.0 or more and length of 10 μm or more: 15 pieces/100 mm^2 or less

An inclusion in a material is, for example, elongated MnS or cementite. These act as a hydrogen accumulation source and cause a significant decrease in the HIC resistance and a decrease in the hydrogen-induced crack growth threshold K_{IH}. Thus, the number of inclusions with an aspect ratio of 2.0 or more and a length of 10 μm or more is 15 pieces/100 mm^2 or less. The number density of the inclusions is preferably 10 pieces/100 mm^2 or less. The lower limit may be, but is not limited to, 0 pieces/100 mm^2.

Retained austenite: 0% to 3% (preferred)

[0048] Retained austenite remaining in a steel material microstructure may act as a hydrogen trap site, increases the amount of hydrogen in the steel, and increases hydrogen embrittlement sensitivity. Furthermore, when a steel material or a steel pipe is used as a steel structure, retained austenite is transformed into martensite due to stress loading during use. Martensite, which is very hard, serves as a source or a propagation path of HIC, and may significantly reduce the K_{IH}. In the present invention, a content of retained austenite of 3% or less improves the K_{IH}. Thus, a content of retained austenite is preferably 3% or less. The content of retained austenite is more preferably 2% or less, still more preferably 1% or less. The content of retained austenite may be 0%.

[0049] Area fraction of bainite in range from surface of steel material (for steel pipe, inner surface of steel pipe) to middle of sheet thickness: 90% or more (preferred)

[0050] As a material suitable for a line pipe, the steel microstructure of a steel material is required to be a bainite microstructure to increase the tensile strength to 520 MPa or more. The bainite microstructure includes bainitic ferrite or granular bainite that transforms during or after accelerated cooling contributing to transformation strengthening, and also includes tempered bainite. A different microstructure, such as ferrite, martensite, pearlite, a martensite-austenite constituent (MA), or retained austenite, in the bainite microstructure reduces the strength, the toughness under normal conditions (in the atmospheric environment), and the K_{IH}. Furthermore, presence of steel microstructures with different hardnesses cause stress distribution in the steel material at the time of stress loading during use, acts as a hydrogen accumulation source due to stress-induced diffusion, and reduces the HIC resistance. Thus, an area fraction of bainite is preferably 90% or more. The area fraction of bainite is more preferably 92% or more, still more preferably 95% or more. The upper limit may be, but is not limited to, 100%.

[0051] Maximum grain size in range from surface of steel material (for steel pipe, inner surface of steel pipe) to middle of sheet thickness: 25 μm or less

[0052] Although the average grain size is reduced to improve the toughness, the reduction of the average grain size is limited when cooling is started at the Ar_3 point or higher. In the present disclosure, it is important to suppress the formation of coarse crystal grains. Crystal grains with a large maximum grain size, if present, induce occurrence of nonuniform strain in the material, promote accumulation of hydrogen, and therefore reduce the fracture toughness in a hydrogen gas environment. In particular, crystal grains with a maximum grain size of more than 25 μm in the range from the inner surface of a steel material to the middle of the sheet thickness are likely to accumulate strain around the grains, easily act as an origin of hydrogen fracture and a propagation path, and significantly reduce the K_{IH}. Thus, the maximum grain size in the range from the inner surface of a steel material to the middle of the sheet thickness should be 25 μm or less. The maximum grain size in the range from the inner surface of a steel material to the middle of the sheet thickness is preferably 24 μm or less, more preferably 22 μm or less, still more preferably 20 μm or less. Although the lower limit is not particularly limited, the maximum grain size is preferably 4 μm or more. The grain size was measured in an area of 1 mm x 1 mm, and the grain size was defined as an area grain size (a weighted average when a boundary with an orientation difference of 15 degrees or more is defined as a grain boundary).

Hydrogen-induced crack growth threshold K_{IH} in high-pressure hydrogen gas environment of 1 MPa or more: 80 $MPa \cdot m^{1/2}$ or more

[0053] For safe operation of a steel structure in an environment containing hydrogen, a high-strength steel material according to the present disclosure has a hydrogen-induced crack growth threshold K_{IH} of 80 $MPa \cdot m^{1/2}$ or more in a high-pressure hydrogen gas environment of 1 MPa or more. Although the upper limit is not particularly limited, the hydrogen-induced crack growth threshold K_{IH} of the steel material is preferably 120 $MPa \cdot m^{1/2}$ or less, more preferably 100 $Pa \cdot m^{1/2}$ or less. The hydrogen-induced crack growth threshold K_{IH} refers to the plane-strain fracture toughness K_{IC} or its provisional value determined in accordance with ASTM E399 and ASTM E1820 in a high-pressure hydrogen gas of 1 MPa or more, or the crack growth threshold or its provisional value determined in accordance with ASTM E1681.

[0054] The sheet thickness of a steel material is preferably, but not limited to, 5 mm or more. The sheet thickness is preferably 30 mm or less.

[0055] The chemical composition and metallic microstructure described above in the present invention can provide a high hydrogen-induced crack threshold K_{IH} in a high-pressure hydrogen gas, and the present invention can be applied to a hydrogen line pipe.

[0056] Furthermore, a high-strength steel material for a line pipe according to the present invention can be produced by specifying the following production conditions, and the production method and conditions are more specifically described below.

Molten steel step [average cooling rate of molten steel: 50°C/min or more (suitable conditions)]

[0057] To reduce inclusions, it is also effective to reduce the S or O content. Because the inclusions limited in the present invention aggregates in a cooling process of molten steel, it is also effective to increase the average cooling rate of the molten steel. Thus, the average cooling rate in the temperature range of 1500°C to 1000°C is preferably 50°C/min or more. The average cooling rate is more preferably 60°C/min or more, still more preferably 70°C/min or more. The upper limit is preferably, but not limited to, 90°C/min or less.

Heating Step

[Heating temperature of cast steel: 1000°C to 1250°C]

[0058] A heating temperature of a cast steel, such as a billet or a slab, lower than 1000°C results in insufficient diffusion of microsegregated impurity elements, such as C, P, or S and an inhomogeneous material, and causes an increase in the number of inclusions, nonuniform precipitations, and lower toughness. Thus, the heating temperature of a cast steel is 1000°C or more. The heating temperature of a cast steel is preferably 1050°C or more, more preferably 1100°C or more. On the other hand, more than 1250°C results in excessively coarse crystal grains and lower toughness. Thus, the heating temperature of a cast steel is 1250°C or less. The heating temperature of a cast steel is preferably 1200°C or less, more preferably 1150°C or less.

Rolling Step

[Total rolling reduction in recrystallization temperature range after heating cast steel: 35% or more and 55% or less]

[0059] To reduce the maximum grain size of bainite, it is necessary to promote the recrystallization of crystal grains and suppress the formation of coarse grains in hot rolling in a recrystallization temperature range after heating a cast steel. When the total rolling reduction in the recrystallization temperature range is less than 35%, recrystallization is insufficient, and coarse grains remain. Thus, the total rolling reduction in the recrystallization temperature range is 35% or more, preferably 38% or more. The total rolling reduction in the recrystallization temperature range is more preferably 40% or more, still more preferably 43% or more. On the other hand, when the total rolling reduction in the recrystallization temperature range is more than 55%, the coarsening of crystal grains can be suppressed, but the rolling reduction in the non-recrystallization region is insufficient, and the crystal grains in the final product cannot be refined. Thus, the total rolling reduction in the recrystallization temperature range is 55% or less, preferably 52% or less. The total rolling reduction in the recrystallization temperature range is more preferably 50% or less, still more preferably 48% or less. The lower limit temperature Tnr of recrystallization can be determined, for example, from the components of steel using the following formula. The surface temperature of a steel sheet can be measured with a radiation thermometer or the like. The total rolling reduction in the recrystallization temperature range refers to a total rolling reduction at a temperature equal to or higher than the lower limit temperature Tnr of recrystallization determined using the following formula.

$$Tnr\ (^{\circ}C) = 174 \times \log\ [\%Nb][\%C + 12/14\%N] + 1444$$

[%X] represents the element X content (% by mass) of the steel.

[Rolling reduction in final rolling pass in recrystallization temperature range: 10% or more]

[0060] In addition to setting the total rolling reduction in the recrystallization temperature range to 35% or more and 55% or less, it is necessary to sufficiently ensure the rolling reduction in the final rolling pass in the recrystallization temperature range and sufficiently promote recrystallization, thereby starting rolling in a partial recrystallization range in a state of uniform grains without coarse grains. When the rolling reduction in the final rolling pass in the recrystallization temperature range is less than 10%, recrystallization is insufficient, and coarse grains grow during the holding time from rough rolling to the start of finish rolling. Thus, the rolling reduction in the final rolling pass in the recrystallization temperature range is 10% or more. The rolling reduction in the final rolling pass in the recrystallization temperature range is preferably 11% or more. The rolling reduction in the final rolling pass in the recrystallization temperature range is more preferably 13% or more, still more preferably 15% or more. Although the upper limit of the rolling reduction in the final rolling pass in the recrystallization temperature range is not particularly limited, a higher rolling reduction is more preferred. However, a rolling reduction of more than 70% results in a significant decrease in productivity, so that the rolling reduction is preferably 70% or less.

[Rolling reduction in final rolling pass at (recrystallization temperature - 80°C) or more: 15% or more]

[0061] Since recrystallization partially occurs even after completion of rolling in the recrystallization range, a further increase in the rolling reduction can promote recrystallization and is effective in refining the top 20% of the grain size. Thus, the rolling reduction in the final rolling pass at (recrystallization temperature - 80°C) or more is 15% or more. The rolling reduction in the final rolling pass at (recrystallization temperature - 80°C) or more is preferably 16% or more. The rolling reduction in the final rolling pass at (recrystallization temperature - 80°C) or more is more preferably 18% or more, still more preferably 20% or more. Although the upper limit of the rolling reduction in the final rolling pass at (recrystallization temperature - 80°C) or more is not particularly limited, a higher rolling reduction is more preferred. However, a rolling reduction of more than 40% results in a significant decrease in productivity, so that the rolling reduction is preferably 40% or less.

[0062] Rolling at a temperature lower than (recrystallization temperature - 80°C) is effective for grain refinement because rolling at a low temperature introduces a large amount of strain. Thus, rolling is preferably performed at a low temperature within the range in which the cooling start temperature of controlled cooling can be complied with.

[0063] In the hot rolling step, although the finish rolling temperature is preferably as low as possible to reduce the grain size, from the perspective of ensuring HISC resistance in a high-pressure hydrogen environment, the finish rolling temperature should be set so that the cooling start temperature of controlled cooling can be the Ar_3 point or higher in terms of a surface temperature of the hot-rolled steel sheet. The term "Ar_3 point", as used herein, refers to a ferrite transformation start temperature during cooling and can be determined, for example, from the components of steel using the following formula. The surface temperature of a hot-rolled steel sheet can be measured with a radiation thermometer or the like.

Ar_3 (°C) = 910 - 310[%C] - 80[%Mn] - 20[%Cu] - 15[%Cr] - 55[%Ni] - 80[%Mo]

[%X] represents the element X content (% by mass) of the steel.

Cooling Step after Rolling (Controlled Cooling Step)

[Cooling start temperature of controlled cooling: Ar_3 transformation point or higher in terms of surface temperature of hot-rolled steel sheet]

[0064] When the steel sheet surface temperature at the start of cooling is lower than the Ar_3 transformation point (the Ar_3 point), ferrite is formed before controlled cooling and greatly decreases the strength. Thus, the surface temperature of a hot-rolled steel sheet at the start of cooling is the Ar_3 transformation point or higher. The surface temperature of the hot-rolled steel sheet at the start of cooling is preferably the Ar_3 transformation point + 20°C or more, more preferably the Ar_3 transformation point + 50°C or more. The surface temperature of a hot-rolled steel sheet at the start of cooling is the temperature of the rear end of the hot-rolled steel sheet at which the cooling start temperature is lowest. The surface temperature of the hot-rolled steel sheet at the start of cooling is preferably the Ar_3 transformation point + 120°C or less, more preferably the Ar_3 transformation point + 80°C or less.

[Cooling start time difference between front end and rear end of hot-rolled steel sheet in controlled cooling: 50 seconds or less]

[0065] When the time difference between a front end and a rear end of a hot-rolled steel sheet in the rolling direction at the start of cooling is more than 50 seconds (s), the temperature difference between the front end and the rear end at the start of cooling increases, resulting in a larger temperature variation at the cooling stop, a larger variation in Vickers hardness at 0.25 mm from a surface of a steel material (for a steel pipe, the inner surface of the steel pipe), and causes lower HISC resistance. Thus, the cooling start time difference between a front end and a rear end of a hot-rolled steel sheet is 50 seconds or less. The cooling start time difference is preferably 45 seconds or less. The cooling start time difference is more preferably 40 seconds or less, still more preferably 32 seconds or less. Although the hot-rolled steel sheet length can be shortened to reduce the cooling start time difference, it reduces the productivity. The cooling start time difference is therefore preferably reduced by increasing the hot-rolled steel sheet line speed. The cooling start time difference may be 0 seconds but is preferably 20 seconds or more from the perspective of productivity.

[Average Cooling Rate of Controlled Cooling]

[0066] Average cooling rate from 750°C to 550°C at middle of sheet thickness: 15°C/s to 50°C/s
When the average cooling rate from 750°C to 550°C at the middle of the sheet thickness is less than 15°C/s, a predetermined bainite microstructure containing granular bainite is not formed, and the strength decreases. Thus, the

average cooling rate at the middle of the sheet thickness is 15°C/s or more. From the perspective of reducing variations in microstructure, the average cooling rate at the middle of the sheet thickness is preferably 17°C/s or more. The average cooling rate at the middle of the sheet thickness is preferably 20°C/s or more, more preferably 25°C/s or more. On the other hand, to suppress variations in the grain size of the bainite microstructure, the average cooling rate is 50°C/s or less. The average cooling rate is preferably 48°C/s or less, more preferably 45°C/s or less. The average cooling rate is still more preferably 42°C/s or less, most preferably 38°C/s or less. Further cooling of a hot-rolled steel sheet temperature at the middle of the sheet thickness to 550°C or less is not particularly limited. However, from the perspective of reducing variations in the microstructure and grain size, the average cooling rate is preferably 15°C/s or more and 50°C/s or less. That is, for further cooling to 550°C or less, the average cooling rate is preferably 15°C/s or more. The average cooling rate is more preferably 30°C/s or more, still more preferably 35°C/s or more. For further cooling to 550°C or less, the average cooling rate is preferably 50°C/s or less. The average cooling rate is more preferably 48°C/s or less, still more preferably 42°C/s or less. The average cooling rate to 550°C or less is an average value of the cooling rates from 550°C to 250°C.

[Cooling Stop Temperature: 250°C to 650°C]

[0067] When the cooling stop temperature at the middle of the sheet thickness after hot rolling is more than 650°C, the material strength decreases greatly. Furthermore, from the perspective of obtaining a uniform bainite microstructure, the cooling stop temperature at the middle of the sheet thickness is 650°C or less. The cooling stop temperature at the middle of the sheet thickness is preferably 620°C or less, more preferably 615°C or less, still more preferably 600°C or less. On the other hand, when the cooling stop temperature at the middle of the sheet thickness is less than 250°C, a quenching crack is likely to occur during cooling. Furthermore, to form a uniform bainite microstructure, the cooling stop temperature is 250°C or more. The cooling stop temperature at the middle of the sheet thickness is preferably 300°C or more, more preferably 350°C or more, still more preferably 380°C or more. From the perspective of reducing the amount of hydrogen in the steel, the cooling stop temperature should be a predetermined temperature or higher. More specifically, hydrogen in the steel is gradually released during cooling and this effect increases with the temperature. However, an excessively low cooling stop temperature results in supercooling and hydrogen remaining in the steel. Moreover, an excessively low cooling stop temperature tends to result in the formation of retained austenite, which stores a larger amount of hydrogen than other phases. Thus, the cooling stop temperature should be 250°C or more to decrease the amount of hydrogen in the steel. After the cooling is stopped, the steel may be allowed to cool and, to promote the formation of bainite, is preferably gradually cooled until the temperature is lowered by approximately 50°C from the cooling stop temperature.

[Dehydrogenation Treatment (Suitable Conditions)]

[0068] Hydrogen originally present in a steel material increases the acceleration of fatigue crack growth and decreases the fatigue life. Thus, dehydrogenation treatment is preferably performed to release hydrogen remaining after production. In the dehydrogenation treatment, the amount of hydrogen in the steel can be reduced by holding the steel at a high temperature for a certain period before using the product. The dehydrogenation treatment can also be achieved by holding for an extended period even at room temperature. For holding at room temperature, the holding time is prolonged and is preferably 96 hours or more. Furthermore, the scale on the steel surface inhibits dehydrogenation and is therefore preferably removed before dehydrogenation treatment. The holding time R (s) is preferably determined from the sheet thickness or the wall thickness t (mm) of a steel sheet or a steel pipe and the hydrogen diffusion coefficient D ($mm \cdot s^{-1}$) in the steel at room temperature using the following formula (A).

$$R \geq t^2/D \quad (A)$$

[0069] The hydrogen diffusion coefficient varies depending on components contained and the metallic microstructure and may range from, for example, 1×10^{-5} to 5×10^{-3} mm^2/s, more preferably 5×10^{-4} mm^2/s or less. The dehydrogenation treatment step is performed before pipe production or welding for connecting steel pipes. The dehydrogenation treatment is preferably performed at a high temperature because the hydrogen diffusion coefficient D at a high temperature is small and hydrogen is released quickly. At a high temperature, the calculation may be performed using a diffusion coefficient D' (diffusion coefficient at each temperature) at a temperature at which the value of D in the formula (A) is held. On the other hand, an excessively high temperature T in the dehydrogenation step results in a significant decrease in the material strength, and the dehydrogenation treatment temperature is preferably 550°C or less. The dehydrogenation treatment temperature T is more preferably 500°C or less. The dehydrogenation treatment temperature T is still more preferably 400°C or less, most preferably 300°C or less. Furthermore, the dehydrogenation treatment temperature T is preferably room temperature or higher for the reason that the dehydrogenation treatment at a temperature lower than room temperature increases the treatment time and cost. The dehydrogenation treatment temperature T is more preferably 50°C or more. The dehydrogenation treatment temperature T is still more preferably 100°C or more, most preferably 150°C

or more. The dehydrogenation treatment temperature T herein is the temperature of the atmosphere in the dehydrogenation treatment step. The room temperature refers to 20°C ± 10°C.

[0070] In particular, when heating, it takes time for the temperature Tc at the middle of the sheet thickness of a steel material or a steel pipe to reach the temperature of the ambient in the dehydrogenation treatment step (dehydrogenation treatment temperature T). Therefore, even if the holding time R (s) satisfies at the ambient temperature, the dehydrogenation treatment may be insufficient if the dehydrogenation treatment temperature T (ambient temperature) has not been reached at the middle of the sheet thickness. Thus, it is preferable to hold for R (s) or more after the temperature Tc at the middle of the sheet thickness reaches the target dehydrogenation treatment temperature T. Furthermore, to achieve a predetermined fracture toughness in hydrogen in hydrogen gas, it is necessary to appropriately adjust the amount of hydrogen in a steel material in a surface layer portion and at the middle of the sheet thickness. For this purpose, it is preferable to hold the steel material at the dehydrogenation treatment temperature T (ambient temperature) for R (s) or more defined by the formula (A), and it is further preferable to hold the steel material for the holding time R (s) or more after the temperature Tc at the middle of the sheet thickness reaches the target dehydrogenation treatment temperature T. In other words, at least the former can appropriately control the amount of hydrogen in the steel material in the surface layer portion of the steel material or the steel pipe, and when the latter is also performed, the amount of hydrogen in the steel material from the surface layer portion to the middle of the sheet thickness of the steel material or the steel pipe can be appropriately controlled. The temperature Tc at the middle of the sheet thickness may be actually measured with a thermocouple or the like or may be predicted using a finite element method or the like.

[0071] The time and temperature in the dehydrogenation treatment step may include the temperature and time applied at the time of heating in the pipe production step of an electric-resistance-welded pipe, UOE, or the like, as described later. Furthermore, the scale on the steel surface inhibits dehydrogenation and is therefore preferably removed before dehydrogenation treatment. The removal method may be, for example, but is not limited to, physical cleaning by high-pressure cleaning or a chemical method using a scale remover. If the scale is removed by approximately 100 μm in thickness, the effects of scale removal can be obtained.

Second Embodiment

[0072] Furthermore, a UOE steel pipe as an example of a steel pipe for a high-strength line pipe can be produced by specifying the following production conditions, and the production method and conditions are more specifically described below. The chemical composition, the metallic microstructure, and the hydrogen-induced crack growth threshold K_{IH} of a UOE steel pipe are the same as those described for the steel sheet of the first embodiment. Further, the molten steel step, the heating step, the hot rolling step, the controlled cooling step after hot rolling, and the dehydrogenation treatment step in the production method are performed in the same manner as described for the steel material. The pipe production step after rolling is more specifically described below.

Pipe Production Step

[0073] A UOE steel pipe is produced by bending a hot-rolled steel sheet, more specifically, groove-cutting an end portion of the hot-rolled steel sheet, forming the steel sheet into a steel pipe shape by C-press, U-press, and O-press, seam-welding a butt joint by inner surface welding and outer surface welding, and performing an expansion step if necessary. The welding method may be any method that can achieve sufficient joint strength and joint toughness and, from the perspective of good weld quality and production efficiency, submerged arc welding is preferably used. Furthermore, a steel pipe produced by press bending into a pipe shape and then seam-welding a butt joint can also be subjected to expansion. Furthermore, when the inclusions are present in a weld heat-affected zone after pipe production, the inclusions act as a hydrogen accumulation source in the same manner as in the base metal zone and reduce the HIC resistance and K_{IH}. To reduce inclusions in a weld, it is also effective to reduce the S or O content. Thus, the average cooling rate in the temperature range of 1500°C to 1000°C in a steel pipe after welding is preferably 50°C/min or more. The average cooling rate is more preferably 55°C/min or more, still more preferably 60°C/min or more. Although the upper limit is not particularly limited, the average cooling rate is preferably 100°C/min or less.

Third Embodiment

[0074] Furthermore, an electric-resistance-welded steel pipe as an example of a steel pipe for a high-strength line pipe according to the present invention can be produced by specifying the following production conditions, and the production method and conditions are more specifically described below. The chemical composition, the metallic microstructure, and the hydrogen-induced crack growth threshold K_{IH} of the steel material are the same as those described for the steel material of the first embodiment. Further, the steps other than the cooling step after rolling and the pipe production step (the molten steel step, the heating step, the hot rolling step, and the dehydrogenation treatment step) in the production method

are performed in the same manner as described for the steel material.

Cooling Step after Rolling (Controlled Cooling Step)

[0075] The cooling start temperature of the controlled cooling and the average cooling rate of the controlled cooling are the same as those described in the first embodiment.

[Cooling Stop Temperature: 250°C to 650°C]

[0076] When the cooling stop temperature at the middle of the sheet thickness after hot rolling is more than 650°C, the material strength decreases greatly, and from the perspective of obtaining a uniform bainite microstructure, the cooling stop temperature at the middle of the sheet thickness is 650°C or less. The cooling stop temperature at the middle of the sheet thickness is preferably 620°C or less, more preferably 615°C or less, still more preferably 600°C or less. On the other hand, when the cooling stop temperature at the middle of the sheet thickness is less than 250°C, a quenching crack is likely to occur during cooling. Thus, the cooling stop temperature at the middle of the sheet thickness is 250°C or more. The cooling stop temperature at the middle of the sheet thickness is preferably 300°C or more, more preferably 350°C or more, still more preferably 380°C or more. To reliably suppress the formation of hard microstructures on the surface of a steel sheet, the cooling stop temperature at the middle of the sheet thickness is most preferably 450°C or more. After the cooling is stopped, the steel may be allowed to cool and, to promote the formation of bainite, is preferably gradually cooled until the temperature is lowered by approximately 50°C from the cooling stop temperature.

[0077] A hot-rolled steel sheet thus produced is then coiled. The coiling temperature is preferably 650°C or less. The coiling temperature is more preferably 620°C or less, still more preferably 615°C or less, still more preferably 600°C or less. The lower limit of the coiling temperature is preferably 250°C or more, more preferably 300°C or more, still more preferably 350°C or more, most preferably 380°C or more.

Pipe Production Step

[0078] An electric-resistance-welded steel pipe as an example of the present invention is produced by forming a cylindrical shape by cold roll forming and butt-welding both circumferential end portions of the cylindrical shape. An electric-resistance-welded steel pipe may also be produced by forming an electric-resistance-welded steel pipe material (electric-resistance-welded steel pipe) using a sizing roll satisfying the following formula (1) (a sizing step) and applying an internal pressure p (MPa) satisfying the following formula (2) to the inner surface of the electric-resistance-welded steel pipe material (an internal pressure applying step). The term "cylindrical shape" means that the cross section of the pipe has a "C" shape.

Diameter (mm) of sizing roll ≥ Thickness (mm) of hot-rolled steel sheet/0.020 (1) (1)

[0079] The thickness of a hot-rolled steel sheet refers to the thickness of the hot-rolled steel sheet before the sizing step.

$$X < p \leq X \times 1.5 \quad (2)$$

[0080] X = (wall thickness (mm) of electric-resistance-welded steel pipe material/radius (mm) of electric-resistance-welded steel pipe material) x yield strength (MPa) of electric-resistance-welded steel pipe material

[0081] The internal pressure can be applied, for example, by sealing a pipe end with a packing made of a rubber material and applying water pressure to the inside of the pipe. To stabilize the shape, if necessary, a die with a desired diameter may be used as an outer frame.

[0082] An electric-resistance-welded steel pipe material as an example of a steel pipe according to the present invention preferably has a wall thickness of 5 mm or more. The electric-resistance-welded steel pipe material preferably has a wall thickness of 30 mm or less. Although the radius of the electric-resistance-welded steel pipe material may have any upper limit, the load on the facilities increases with the radius, and the electric-resistance-welded pipe material therefore preferably has a radius of 400 mm or less. The electric-resistance-welded pipe material preferably has a radius of 200 mm or more. The electric-resistance-welded steel pipe material preferably has a yield strength of 480 MPa or more to withstand pipeline operation gas pressures. The yield strength is more preferably 500 MPa or more. On the other hand, to avoid an increase in hydrogen embrittlement sensitivity, the yield strength is preferably 560 MPa or less. The yield strength is more preferably 550 MPa or less.

[0083] In the sizing step, passage through rolls causes bending deformation along the roll shape in the pipe axis direction and generates residual stress in the pipe axis direction. The absolute value of the residual stress in the pipe axis

direction increases with the bending strain in the bending deformation. The bending strain increases as the diameter of the sizing roll decreases and as the thickness of the hot-rolled steel sheet increases. Thus, in the present invention, from the perspective of reducing the shear residual stress, the diameter of the sizing roll satisfies the formula (1) to reduce the absolute value of the residual stress in the pipe axis direction. When the sizing roll has a diameter smaller than the right side of the formula (1), the shear residual stress intended in the present invention cannot be obtained. Although the diameter of the sizing roll may have any upper limit, the load on the facilities increases with the sizing roll, and the sizing roll therefore preferably has a diameter of 2000 mm or less.

[0084] In the internal pressure applying step, the electric-resistance-welded steel pipe material is expanded to generate tensile stress in the circumferential direction of the pipe and reduce the absolute value of residual stress in the circumferential direction of the pipe. As the internal pressure p (MPa) in the internal pressure applying step increases, the absolute value of the residual stress in the circumferential direction of the pipe decreases. The tensile stress generated in the circumferential direction of the pipe increases as the radius of the steel pipe increases and as the wall thickness of the steel pipe decreases.

[0085] The left side (X) of the formula (2) corresponds to the internal pressure p when the tensile stress generated in the circumferential direction of the pipe is equal to the yield stress of the electric-resistance-welded steel pipe material. In the present invention, from the perspective of reducing the shear residual stress, the internal pressure p is larger than the left side (X) of the formula (2) to expand the electric-resistance-welded steel pipe material to the plastic region in order to reduce the absolute value of the residual stress in the pipe axis direction. On the other hand, when the internal pressure p exceeds the right side (X x 1.5) of the formula (2), the absolute value of the residual stress in the circumferential direction of the pipe decreases, but the amount of work hardening due to expansion increases excessively, the dislocation density on the pipe surface increases, and the fracture toughness in hydrogen decreases.

[0086] As partially described above, regarding a high-strength steel pipe, a high-strength steel pipe for a line pipe for sour gas service (a UOE steel pipe, an electric-resistance-welded steel pipe, a spiral steel pipe, or the like) with high material uniformity in the steel sheet suitable for transportation of crude oil or natural gas can be produced by forming a high-strength steel material according to the present disclosure into a tubular shape by press bending, roll forming, UOE forming, or the like and then welding a butt joint. Furthermore, a high-strength steel sheet according to the present disclosure can be used for a steel pipe to produce a steel pipe with high HISC resistance even when a high hardness region of a weld is present.

EXAMPLE 1

[0087] The present invention is more specifically described in the following examples. The examples are preferred examples of the present invention, and the present invention is not limited to these examples.

[0088] Slabs with the chemical compositions shown in Tables 1-1 and 1-2 were prepared, were hot-rolled, and were subjected to controlled cooling and dehydrogenation treatment to produce steel materials. The steel materials were formed into steel pipes. The production conditions are shown in Tables 2-1 and 2-2. For Nos. 2 to 6, 12 to 22, 35, and 37, steel pipes were formed by the pipe production step of bending each steel material (hot-rolled steel sheet) and butt-welding both end portions thereof. For Nos. 7 to 11, 23 to 33, 36, and 38, steel pipes were formed by the pipe production step of forming each steel material (hot-rolled steel sheet) into a cylindrical shape by cold roll forming and subjecting both circumferential end portions of the cylindrical shape to butt electric resistance welding. In Nos. 1 and 34, the steel materials were used as they were. Tables 3-1 and 3-2 show the evaluation results of the metallic microstructure and the material quality of each of the steel materials and steel pipes thus produced. The evaluation method is described below.

Retained Austenite Measurement

[0089] A sample for metallic microstructure observation was taken from a central portion of the sheet width in a central portion in the longitudinal direction of each of the steel materials and the steel pipes thus produced. A cross section parallel to the longitudinal direction was buffed as an observation surface. The surface layer was then removed by chemical polishing using picric acid etching, and X-ray diffractometry was performed. More specifically, a Co-Kα radiation source was used for an incident X-ray, and the area fraction of retained austenite was calculated from the intensity ratios of the (200), (211), and (220) planes of ferrite to the (200), (220), and (311) planes of austenite.

Calculation of Maximum Grain Size and Area Fraction of Bainite

[0090] A sample for metallic microstructure observation was taken from a central portion of the sheet width of each of the steel materials and the steel pipes thus produced, and a cross section of the sample parallel to the rolling longitudinal direction was used as a surface to be observed. The surface to be observed was mirror-polished and etched with colloidal silica, and crystal data were collected by an electron backscatter diffraction (EBSD) method in a visual field of 1 mm x 1 mm

at the center of the sample (measuring step: 0.8 μm). The grain size was defined as an area grain size (a weighted average when a boundary with an orientation difference of 15 degrees or more is defined as a grain boundary). Each grain size was determined from the crystal data to determine the maximum grain size. For the area fraction, the surface to be observed was etched with a 3% by volume nital solution, and a scanning electron microscope photograph was taken at an appropriate magnification in the range of 1000 to 5000 times to observe bainite. The bainite was visually identified by comparison with the microstructure photograph of Non Patent Literature 1, and the microstructure fraction was determined as an area fraction of bainite by binarizing the bainite and the other region in the SEM photograph based on the above identification and determining the area fraction by image analysis.

Observation of Inclusions and Calculation of Number Density

[0091] A sample for metallic microstructure observation was taken from a central portion of the sheet width in a central portion in the longitudinal direction of each of the steel materials and the steel pipes thus produced, and a cross section of the sample parallel to the rolling longitudinal direction was used as a surface to be observed. The surface to be observed was mirror-polished, was then etched with colloidal silica, and was observed with a scanning electron microscope (SEM) in a visual field of 10 mm x 10 mm at the center of the sample. The observation magnification ranges from 2000 to 5000 times, and the average of three visual fields was taken as the number density of inclusions.

Tensile Strength (TS)

[0092] JIS No. 14 proportional test pieces (parallel portion diameter: 7 mm, gauge length: 35 mm) were taken in accordance with JIS Z 2201 from the steel materials and the steel pipes thus produced, and the tensile strength was measured.

Hydrogen Temperature-Programmed Analysis

[0093] The amount of hydrogen remaining in the steel was measured by thermal desorption spectrometry using a low-temperature programmed hydrogen analyzer <gas chromatograph type> (JTF-20AL). The thermal desorption spectrometry was performed in the temperature range of room temperature to 400°C at a heating rate of 200°C/h, and the sum total thereof was taken as the amount of hydrogen. The specimen has a cylindrical shape with 30 mm in length and 7Φ in diameter in the longitudinal direction of the steel pipe at the quarter thickness position of the steel material and at the quarter thickness position from the inner surface of the steel pipe. The amount of hydrogen is the amount of H shown in Tables 1-1 and 1-2 before being subjected to a high-pressure hydrogen fatigue test as explained in the item described later.

Fracture Toughness Test in High-Pressure Hydrogen Gas

[0094] The test was performed in accordance with ASTM E1820 in a hydrogen gas (including 100% hydrogen) at room temperature (20°C ± 10°C) with a pressure of 25 MPa or in a natural gas (the main components are hydrocarbons, such as methane and ethane) mixed atmosphere having the above temperature and pressure and containing hydrogen at a hydrogen partial pressure of 1 MPa or more. A CT test specimen (thickness: 12.7 mm, width: 25.4 mm) was used as a test specimen and was taken in a direction in which the machine notch introduction direction was parallel to the rolling direction of the steel material. A fatigue precrack was introduced in the atmosphere, and the conditions included frequency: 1 Hz, cyclic loading waveform: sine wave, control method: K-value control, and stress ratio R: 0.1. The atmosphere was then changed to hydrogen gas or a hydrogen gas + natural gas mixed atmosphere. A fracture toughness test was performed by an unloading-elastic compliance method using a single test specimen. The crosshead displacement speed at the time of loading was 0.002 mm/s.

[0095] The evaluation results are shown in Tables 3-1 and 3-2. All of the steel materials and steel pipes satisfying Inventive examples of the present invention had high fracture toughness resistance in hydrogen with a hydrogen-induced crack growth threshold K_{IH} of 80 MPa·$m^{1/2}$ or more and had a tensile strength of 520 MPa or more. The steel pipes in Tables 3-1 and 3-2 also showed the same results as the steel materials.

[Table 1-1]

Steel No.	Chemical composition (% by mass) *1																										Ar3 point (°C)	Notes
	C	Si	Mn	P	S	Al	O	N	Nb	H	Ca	Ni	Ti	Cu	Cr	Mo	W	V	Zr	Mg	REM	B	Ta	Hf	Re	Sn	Sb	
1	0.07	0.99	1.0	0.01050	0.0012	0.0520	0.0070	0.003	0.0520	0.0000016																	808	Conforming steel
2	0.09	0.90	0.8	0.00810	0.0002	0.0830	0.0060	0.003	0.0010	0.0000018																	818	Conforming steel
3	0.11	0.64	1.2	0.00550	0.0007	0.0890	0.0050	0.004	0.1000	0.0000018																	780	Conforming steel
4	0.06	0.98	0.9	0.00620	0.0008	0.0420	0.0080	0.002	0.0340	0.0000013																	819	Conforming steel
5	0.07	2.00	1.1	0.00370	0.0010	0.0940	0.0060	0.005	0.0170	0.0000020																	800	Conforming steel
6	0.07	1.89	1.4	0.00630	0.0010	0.0640	0.0100	0.005	0.0420	0.0000019																	776	Conforming steel
7	0.05	1.43	1.4	0.00550	0.0003	0.0710	0.0080	0.005	0.0750	0.0000020																	783	Conforming steel
8	0.04	0.35	1.2	0.00990	0.0011	0.0420	0.0090	0.002	0.0900	0.0000016																	802	Conforming steel
9	0.03	1.43	1.3	0.00320	0.0007	0.1180	0.0010	0.004	0.0440	0.0000018																	797	Conforming steel
10	0.04	0.27	1.4	0.01070	0.0003	0.1110	0.0090	0.005	0.0720	0.0000018																	786	Conforming steel
11	0.08	1.58	1.3	0.00910	0.0003	0.0920	0.0060	0.003	0.0340	0.0000013																	781	Conforming steel
12	0.11	1.50	0.8	0.00930	0.0007	0.0600	0.0060	0.003	0.0760	0.0000016	0.00020	0.95	0.100	0.590	0.650	0.210	0.110	0.01	0.0290	0.0067	0.0011	0.00100	0.0120	0.0094	0.0018	0.130	0.256	721	Conforming steel
13	0.06	1.63	0.8	0.00620	0.0007	0.0250	0.0040	0.003	0.0130	0.0000018	0.00230	0.60	0.099	0.510	0.920	0.160	0.750	0.08	0.0084	0.0081	0.0006	0.00080	0.0607	0.0021	0.0005	0.291	0.117	758	Conforming steel
14	0.08	1.90	1.1	0.00560	0.0008	0.0080	0.0070	0.005	0.0190	0.0000018	0.00250	0.33	0.027	0.010	0.790	0.220	0.060	0.06	0.0031	0.0095	0.0033	0.00070	0.0346	0.0220	0.0018	0.192	0.003	749	Conforming steel
15	0.03	0.48	0.6	0.01010	0.0011	0.0950	0.0090	0.004	0.0040	0.0000013	0.00090	0.71	0.029	1.000	0.670	0.520	0.260	0.02	0.0270	0.0058	0.0052	0.00110	0.0635	0.1856	0.0048	0.244	0.213	742	Conforming steel
16	0.11	0.85	0.5	0.01270	0.0009	0.0680	0.0030	0.002	0.0170	0.0000016	0.00460	0.08	0.016	0.410	0.505	0.170	0.840	0.08	0.0320	0.0063	0.0017	0.00150	0.0437	0.0234	0.0005	0.104	0.137	802	Conforming steel
17	0.05	0.97	0.7	0.00480	0.0008	0.0670	0.0090	0.001	0.0030	0.0000017	0.00230	0.31	0.043	0.380	0.220	0.390	0.680	0.01	0.0230	0.0073	0.0039	0.00170	0.0731	0.1639	0.0040	0.159	0.046	779	Conforming steel
18	0.04	1.57	1.0	0.01240	0.0007	0.0240	0.0030	0.001	0.0190	0.0000018	0.00100	0.55	0.093	0.820	1.000	0.600	0.370	0.04	0.0430	0.0053	0.0005	0.00180	0.1383	0.0707	0.0018	0.286	0.057	708	Conforming steel
19	0.04	1.51	1.4	0.00500	0.0009	0.0960	0.0030	0.003	0.0860	0.0000013	0.00110	0.76	0.014	0.030	0.210	0.305	0.220	0.05	0.0013	0.0041	0.0005	0.00170	0.0846	0.0092	0.0004	0.199	0.206	716	Conforming steel
20	0.05	1.08	1.0	0.00900	0.0003	0.0070	0.0020	0.003	0.0500	0.0000016	0.00080	1.65	0.021	0.720	0.510	0.010	0.930	0.04	0.0045	0.0023	0.0004	0.00150	0.1721	0.0649	0.0031	0.076	0.280	701	Conforming steel

*1 The remainder is composed of Fe and incidental impurities
Blank: no intended addition.
Underline: outside the scope of the present invention.

[Table 1-2]

Steel No.	Chemical composition (% by mass) *1																										Ar3 point (°C)	Notes
	C	Si	Mn	P	S	Al	O	N	Nb	H	Ca	Ni	Ti	Cu	Cr	Mo	W	V	Zr	Mg	REM	B	Ta	Hf	Re	Sn	Sb	
21	0.08	0.77	1.0	0.00980	0.0004	0.1220	0.0060	0.003	0.0910	0.0000018	0.00020	0.57	0.099	0.530	0.270	0.310	0.460	0.02	0.0120	0.0058	0.0067	0.00150	0.0523	0.0167	0.0039	0.088	0.280	734	Conforming steel
22	0.06	0.81	1.0	0.00960	0.0015	0.0680	0.0090	0.001	0.0480	0.0000018	0.00160	1.71	0.011	0.190	0.390	0.060	0.505	0.06	0.0200	0.0053	0.0007	0.00150	0.1794	0.0325	0.0013	0.034	0.050	703	Conforming steel
23	0.08	1.98	0.6	0.00680	0.0008	0.1100	0.0090	0.004	0.0590	0.0000013	0.00340	1.57	0.018	0.180	0.240	0.240	0.220	0.03	0.0340	0.0045	0.0084	0.00120	0.1882	0.0552	0.0009	0.104	0.171	724	Conforming steel
24	0.10	0.15	1.4	0.00900	0.0008	0.0790	0.0050	0.003	0.0230	0.0000019	0.00440	0.50	0.026	0.970	0.530	0.300	1.000	0.07	0.0360	0.0098	0.0053	0.00120	0.1635	0.1357	0.0009	0.029	0.076	688	Conforming steel
25	0.03	0.93	1.3	0.01380	0.0010	0.0370	0.0050	0.003	0.0960	0.0000016	0.00340	0.38	0.085	0.620	0.980	0.160	0.750	0.06	0.0490	0.0018	0.0003	0.00120	0.0981	0.1010	0.0002	0.254	0.196	736	Conforming steel
26	0.13	1.75	1.0	0.00970	0.0007	0.0540	0.0030	0.003	0.0700	0.0000018	0.00080	0.21	0.027	0.840	0.510	0.030	0.450	0.01	0.0290	0.0027	0.0016	0.00150	0.0095	0.0263	0.0043	0.162	0.248	751	Conforming steel
27	0.09	0.62	1.0	0.01020	0.0013	0.1270	0.0040	0.002	0.0590	0.0000018	0.00090	0.13	0.020	0.750	0.030	0.020	0.190	0.10	0.0360	0.0066	0.0011	0.00110	0.0987	0.1240	0.0007	0.087	0.296	778	Conforming steel
28	0.03	0.64	1.1	0.00560	0.0014	0.0320	0.0040	0.005	0.0740	0.0000013	0.00070	0.22	0.034	0.090	0.950	0.470	0.410	0.01	0.0260	0.0036	0.0008	0.00170	0.0104	0.1006	0.0020	0.155	0.208	747	Conforming steel
29	0.14	1.43	1.1	0.00810	0.0010	0.0460	0.0060	0.002	0.0050	0.0000016	0.00080	0.31	0.085	0.620	0.120	0.200	0.270	0.08	0.0140	0.0045	0.0070	0.00160	0.1946	0.0117	0.0026	0.176	0.111	731	Conforming steel
30	0.14	0.57	1.0	0.00760	0.0009	0.0050	0.0040	0.001	0.0670	0.0000018	0.00240	0.12	0.068	0.830	0.470	0.210	0.860	0.03	0.0500	0.0053	0.0084	0.00140	0.1778	0.1977	0.0011	0.155	0.033	740	Conforming steel
31	0.04	1.51	1.2	0.00780	0.0013	0.0960	0.0090	0.005	0.0800	0.0000011	0.00320	0.22	0.071	0.440	0.420	0.390	0.880	0.09	0.0220	0.0049	0.0059	0.00190	0.1870	0.1876	0.0033	0.221	0.221	743	Conforming steel
32	0.07	0.28	1.1	0.00090	0.0012	0.1270	0.0030	0.005	0.0500	0.0000018	0.00330	0.16	0.065	0.420	0.330	0.280	0.570	0.06	0.0033	0.0035	0.0072	0.00110	0.0534	0.1242	0.0034	0.132	0.192	756	Conforming steel
33	0.04	1.57	1.0	0.01240	0.0007	0.0240	0.0030	0.001	0.0190	0.0000013	0.00100	0.12	0.093	0.820	0.270	0.600	0.460	0.02	0.0103	0.0058	0.0067	0.00120	0.0523	0.1668	0.0039	0.088	0.280	743	Conforming steel
34	0.05	1.51	1.4	0.00620	0.0007	0.0250	0.0070	0.005	0.0480	0.0000034	0.00160	0.32	0.053	0.230	0.170	0.420	0.880	0.08	0.0240	0.0087	0.0014	0.00140	0.1874	0.1630	0.0046	0.022	0.132	724	Comparative steel
35	0.03	1.58	1.5	0.00940	0.0012	0.1150	0.0400	0.003	0.0610	0.0000014	0.00310	0.22	0.043	0.640	0.750	0.010	0.890	0.09	0.0310	0.0053	0.0098	0.00180	0.1224	0.1916	0.0004	0.150	0.269	744	Comparative steel
36	0.05	1.89	1.1	0.00430	0.0021	0.0130	0.0050	0.002	0.0540	0.0000014	0.00040	0.45	0.081	0.590	0.660	0.380	0.770	0.05	0.0320	0.0027	0.0030	0.00130	0.0298	0.1537	0.0001	0.040	0.168	730	Comparative steel
37	0.06	1.63	0.8	0.00620	0.0007	0.0920	0.0060	0.002	0.0170	0.0000016		0.71																788	Conforming steel
38	0.03	0.71	1.2	0.00830	0.0003	0.0880	0.0080	0.001	0.0920	0.0000013	0.00030	1.20		0.730	0.420	0.330												691	Conforming steel

*1 The remainder is composed of Fe and incidental impurities
Blank: no intended addition.
Underline: outside the scope of the present invention.

[Table 2-1]

Steel pipe No.	Steel material No.	Steel No.	Thickness (mm)	Cooling step of molten steel	Heating step	Hot rolling step				Cooling step							Dehydrogenation treatment step		Notes
				Molten steel average cooling rate (°C/min)	Heating temperature (°C)	Rolling in recrystallization region		Rolling reduction in final rolling pass at (recrystallization temperature - 80°C) or more (%)	Finish rolling temperature (°C)	Start of cooling			750 → 550°C cooling rate (0.25 mm below surface of steel sheet) (°C/s)	750 → 550°C cooling rate (middle of sheet thickness) (°C/s)	550°C or less cooling rate (0.25 mm below surface of steel sheet) (°C/s)	550°C or less cooling rate (middle of sheet thickness) (°C/s)	Cooling stop temperature (°C)	Temperature (°C)	Time (h)	
						Total rolling reduction (%)	Rolling reduction in final rolling pass (%)			Surface temperature at rear end of steel sheet (°C)	Temperature - Ar_3 (°C)	Cooling start time difference between front and rear ends of steel sheet (s)								
-	1	1	35	50	1030	37	12	17	895	810	2	41	18	16	170	16	450	300	100	Inventive example
2	2	2	30	56	1060	39	11	16	820	819	1	35	21	18	190	18	410	200	150	Inventive example
3	3	3	18	51	990	49	11	17	855	785	5	19	35	31	230	33	500	500	100	Comparative example
4	4	4	27	59	1090	40	10	15	829	822	3	32	24	21	200	22	430	50	150	Inventive example
5	5	5	39	50	1000	35	12	17	821	801	1	50	17	15	150	15	440	50	200	Inventive example
6	6	6	21	50	1290	46	13	20	842	794	18	20	30	26	220	27	550	50	200	Comparative example
7	7	7	12	55	1250	33	13	20	890	790	8	10	50	44	250	47	500	25	200	Comparative example
8	8	8	24	59	1100	43	12	16	816	803	1	25	26	23	210	24	500	25	200	Inventive example
9	9	9	15	61	1200	52	10	18	883	802	5	15	41	36	240	38	450	25	200	Inventive example
10	10	10	27	50	1080	39	13	19	833	798	12	29	25	22	210	23	470	200	150	Inventive example
11	11	11	18	55	1150	48	10	17	851	783	2	18	34	30	230	32	480	200	150	Inventive example
12	12	12	30	55	1050	58	13	20	820	781	60	34	20	17	180	17	500	200	150	Comparative example
13	13	13	15	59	1190	51	12	18	847	765	7	14	39	35	230	37	440	200	150	Inventive example
14	14	14	12	63	1240	54	12	20	884	786	37	11	49	44	250	47	550	500	200	Inventive example
15	15	15	21	60	1120	45	13	20	845	795	53	21	31	27	220	28	440	25	200	Inventive example
16	16	16	18	60	1170	48	9	18	864	810	8	17	36	32	220	34	500	80	200	Comparative example
17	17	17	39	69	1010	36	15	20	812	792	10	45	17	15	160	15	490	80	200	Inventive example
18	18	18	35	50	1040	37	14	20	811	787	79	40	19	17	170	17	440	50	250	Inventive example
19	19	19	24	55	1110	42	12	18	842	772	56	24	25	22	200	23	550	50	250	Inventive example
20	20	20	15	53	1210	52	12	19	861	775	74	16	40	35	240	37	500	50	250	Inventive example

Underline: outside the scope of the present invention.

[Table 2-2]

Steel pipe No.	Steel material No.	Steel No.	Thickness (mm)	Cooling step of molten steel	Heating step	Hot rolling step				Cooling step								Dehydrogenation treatment step		Notes
				Molten steel average cooling rate (°C/min)	Heating temperature (°C)	Rolling in recrystallization region		Rolling reduction in final rolling pass at (recrystallization temperature - 80°C) or more (%)	Finish rolling temperature (°C)	Start of cooling			750 → 550°C cooling rate (below surface of steel sheet 0.25 mm) (°C/s)	750 → 550°C cooling rate (middle of sheet thickness) (°C/s)	550°C or less cooling rate (below surface of steel sheet 0.25 mm) (°C/s)	550°C or less cooling rate (middle of sheet thickness) (°C/s)	Cooling stop temperature (°C)	Temperature (°C)	Time (h)	
						Total rolling reduction (%)	Rolling reduction in final rolling pass (%)			Surface temperature at rear end of steel sheet (°C)	Temperature - -Ar3 (°C)	Cooling start time difference between front and rear ends of steel sheet (s)								
21	21	21	35	55	1100	38	10	15	806	780	46	39	20	17	160	17	450	50	250	Inventive example
22	22	22	27	56	1240	41	12	20	829	759	56	27	23	20	200	21	480	80	200	Inventive example
23	23	23	39	65	1020	33	11	16	800	760	36	47	18	16	150	16	500	80	200	Inventive example
24	24	24	35	60	1020	57	7	17	808	762	74	42	17	15	160	15	500	80	200	Comparative example
25	25	25	30	54	1070	37	13	17	814	769	33	36	22	19	190	19	480	500	100	Inventive example
26	26	26	35	53	1030	36	10	20	819	770	19	41	19	17	180	17	480	500	100	Inventive example
27	27	27	30	55	1050	38	11	15	820	761	-17	35	20	17	180	17	470	500	100	Comparative example
28	28	28	24	58	1090	44	11	14	817	778	31	30	27	24	210	25	450	500	100	Comparative example
29	29	29	39	51	1010	35	12	18	821	769	37	34	14	12	160	12	500	500	100	Comparative example
30	30	30	15	57	1200	53	11	18	848	767	27	15	57	51	240	53	500	500	100	Comparative example
31	31	31	29	60	1200	40	10	16	885	799	56	30	15	19	220	33	490	-	-	Inventive example
32	32	32	35	61	1020	38	10	16	820	793	37	40	20	17	170	17	450	25	24	Inventive example
33	33	33	38	32	1050	38	10	15	806	780	37	40	17	15	160	15	500	500	100	Inventive example
-	34	34	30	54	1070	49	13	17	814	782	58	36	22	19	190	19	240	500	100	Comparative example
35	35	35	12	55	1250	52	12	19	861	775	31	16	40	35	240	37	490	500	100	Comparative example
36	36	36	21	60	1120	45	13	20	845	770	40	21	31	27	220	28	460	500	100	Comparative example
37	37	37	24	55	1200	40	12	18	840	819	89	39	25	22	210	23	390	200	200	Inventive example
38	38	38	29	60	1180	38	14	20	829	770	40	42	31	27	220	19	400	250	120	Inventive example

Underline: outside the scope of the present invention.

[Table 3-1]

Steel pipe No.	Steel material No.	Steel No.	Area fraction of retained austenite (%)	B fraction (%)	Maximum grain size of bainite (μm)	Number density of inclusions (/100 mm^2)	Tensile strength (MPa)	Hydrogen-induced crack growth threshold KIH (MPa √m)	Notes
								Base metal zone	
1	1	1	0.0	98.6	22	15	699	129	Inventive example
2	2	2	0.2	92.5	23	12	561	159	Inventive example
3	3	3	0.5	93.4	19	33	700	75	Comparative example
4	4	4	0.0	98.1	21	12	552	170	Inventive example
5	5	5	0.9	95.9	18	15	590	129	Inventive example
6	6	6	1.0	95.3	55	14	530	62	Comparative example
7	7	7	0.3	98.1	51	12	702	60	Comparative example
8	8	8	0.0	90.6	19	9	662	100	Inventive example
9	9	9	0.8	92.0	19	15	543	167	Inventive example
10	10	10	0.9	93.4	22	9	559	131	Inventive example

(continued)

Steel pipe No.	Steel material No.	Steel No.	Area fraction of retained austenite (%)	B fraction (%)	Maximum grain size of bainite (μm)	Number density of inclusions (/100 mm^2)	Tensile strength (MPa)	Hydrogen-induced crack growth threshold KIH (MPa √m)	Notes
								Base metal zone	
11	11	11	0.4	92.0	22	10	576	119	Inventive example
12	12	12	2.5	99.9	39	9	699	59	Comparative example
13	13	13	0.0	91.8	24	11	653	99	Inventive example
14	14	14	0.0	97.1	22	9	626	100	Inventive example
15	15	15	0.0	93.6	14	10	564	137	Inventive example
16	16	16	2.1	98.9	41	13	756	59	Comparative example
17	17	17	0.0	94.5	25	15	534	153	Inventive example
18	18	18	0.5	96.1	18	15	545	157	Inventive example
19	19	19	0.0	90.5	21	12	623	122	Inventive example
20	20	20	0.0	93.7	19	11	647	120	Inventive example

Underline: outside the scope of the present invention.
B: bainite

[Table 3-2]

Steel pipe No.	Steel material No.	Steel No.	Area fraction of retained austenite (%)	B fraction (%)	Maximum grain size of bainite (μm)	Number density of inclusions (/100 mm^2)	Tensile strength (MPa)	Hydrogen-induced crack growth threshold KIH (MPa √m)	Notes
								Base metal zone	
21	21	21	0.0	90.1	17	9	586	140	Inventive example
22	22	22	0.2	94.0	24	13	525	160	Inventive example
23	23	23	0.0	97.0	23	9	548	149	Inventive example

(continued)

Steel pipe No.	Steel material No.	Steel No.	Area fraction of retained austenite (%)	B fraction (%)	Maximum grain size of bainite (μm)	Number density of inclusions (/100 mm^2)	Tensile strength (MPa)	Hydrogen-induced crack growth threshold KIH (MPa √m)	Notes
								Base metal zone	
24	24	24	1.5	92.9	40	14	530	58	Comparative example
25	25	25	0.0	90.2	20	10	587	124	Inventive example
26	26	26	0.0	92.8	19	11	554	137	Inventive example
27	27	27	0.0	92.7	17	9	503	90	Comparative example
28	28	28	1.2	91.1	29	8	530	70	Comparative example
29	29	29	1.1	82.1	26	13	501	69	Comparative example
30	30	30	0.0	93.0	30	14	556	70	Comparative example
31	31	31	0.0	90.9	18	10	520	82	Inventive example
32	32	32	1.7	93.6	24	15	700	81	Inventive example
33	33	33	0.3	99.3	19	15	579	81	Inventive example
34	34	34	3.9	90.1	17	12	593	61	Comparative example
35	35	35	0.0	90.9	17	25	609	69	Comparative example
36	36	36	0.2	94.0	19	27	525	59	Comparative example
37	37	37	0.9	93.2	21	9	582	102	Inventive example
38	38	38	1.0	90.4	18	13	601	99	Inventive example

Underline: outside the scope of the present invention.
B: bainite

EXAMPLE 2

[0096] Examples that have verified the advantages of the present invention are described below. In the following Examples, steel materials and steel pipes were produced under the following production conditions and were characterized. The steel Nos. 2, 4, 8, 14, 22, and 33 shown in Tables 1-1 and 1-2 used in Example 1 were used, up to the controlled cooling step was performed under the same conditions as the steel pipes 2, 4, 8, 14, 22, and 33 shown in Example 1 (Tables 2-1 and 2-2). Steel pipe forming was also performed under the same conditions as Example 1, and the characteristics were

evaluated while the dehydrogenation treatment conditions were changed. Table 4 shows the results.

[0097] The dehydrogenation treatment of the steel pipes Nos. 2, 4, 8, 14, 22, and 33 in Example 1 was performed at a dehydrogenation treatment temperature T (ambient temperature) and time shown in Tables 2-1 and 2-2. As shown in Table 4, the dehydrogenation holding time t corresponds to Y and the retention time tc at the temperature Tc at the middle of the sheet thickness corresponds to N, respectively in Table 4.

[0098] For the steel pipes Nos. 2A, 4A, 8A, 14A, 22A, and 33A, the dehydrogenation treatment temperature T was the temperature shown in Table 4, and the holding time tc after the temperature Tc at the middle of the sheet thickness reached the dehydrogenation treatment temperature T shown in Table 4 satisfied the formula (A).

[0099] For the steel pipes Nos. 2B, 4B, 8B, 14B, 22B, and 33B, the dehydrogenation treatment temperature T is the temperature shown in Table 4, but neither the holding time t at the ambient temperature nor the holding time tc after the temperature Tc at the middle of the sheet thickness reaches the dehydrogenation treatment temperature T satisfy the formula (A).

[0100] In Table 4, "Dehydrogenation holding time t is Y" means that the dehydrogenation treatment temperature T (ambient temperature) is a predetermined temperature and the holding time t satisfies the formula (A), and "Dehydrogenation holding time t is N" means that the dehydrogenation treatment temperature T (ambient temperature) is a predetermined temperature, but the holding time t does not satisfy the formula (A). Furthermore, "Holding time tc at steel material center temperature Tc is Y" means that the holding time tc after the temperature Tc at the middle of the sheet thickness reaches a predetermined temperature satisfies the formula (A), and "Holding time tc at steel material center temperature Tc is N" means that the temperature Tc at the middle of the sheet thickness reaches a predetermined temperature, but the holding time tc after Tc reaches a predetermined temperature does not satisfy the formula (A).

[0101] The examination of the fracture toughness in hydrogen and the tensile strength and the evaluation of the microstructure and inclusions were performed in the same manner as in Example 1.

[0102] All of the examples of the present invention satisfied the conditions of a hydrogen-induced crack growth threshold K_{IH} of 80 MPa·$m^{1/2}$ or more and a tensile strength of 520 MPa or more. Among them, the fracture toughness resistance in hydrogen was better when the dehydrogenation treatment was performed under more suitable conditions.

[0103] The steel pipes in Table 4 also showed the same results as the steel materials.

[Table 4]

Steel pipe No.	Steel material No.	Steel No.	Dehydrogenation treatment temperature T (°C)	Dehydrogenation holding time t	Holding time tc at steel material center temperature T_c	Area fraction of retained austenite (%)	B fraction % ()	Maximum grain size of bainite (μm)	Number density of inclusions (/100 mm^2)	Tensile strength (MPa)	Hydrogen-induced crack growth threshold KIH (MPa √m) base metal zone	Notes
2A	2A	2	200	Y	Y	0.2	92.5	23	12	540	199	Inventive example
2	2	2	200	Y	N	0.2	92.5	23	12	561	159	Inventive example
2B	2B	2	200	N	N	0.2	92.5	23	12	570	134	Inventive example
4A	4A	4	50	Y	Y	0.0	98.1	21	12	552	188	Inventive example
4	4	4	50	Y	N	0.0	98.1	21	12	552	170	Inventive example
4B	4B	4	50	N	N	0.0	98.1	21	12	552	132	Inventive example
8A	8A	8	25	Y	Y	0.0	90.6	19	9	662	151	Inventive example
8	8	8	25	Y	N	0.0	90.6	19	9	662	100	Inventive example
8B	8B	8	25	N	N	0.0	90.6	19	9	662	88	Inventive example
14A	14A	14	500	Y	Y	0.0	97.1	22	9	589	141	Inventive example
14	14	14	500	Y	N	0.0	97.1	22	9	626	100	Inventive example
14B	14B	14	500	N	N	0.0	97.1	22	9	641	89	Inventive example

(continued)

Steel pipe No.	Steel material No.	Steel No.	Dehydrogenation treatment temperature T (°C)	Dehydrogenation holding time t	Holding time tc at steel material center temperature T_c	Area fraction of retained austenite (%)	B fraction % ()	Maximum grain size of bainite (μm)	Number density of inclusions (/100 mm^2)	Tensile strength (MPa)	Hydrogen-induced crack growth threshold KIH (MPa √m) base metal zone	Notes
22A	22A	22	80	Y	Y	0.2	94.0	24	13	525	179	Inventive example
22	22	22	80	Y	N	0.2	94.0	24	13	525	160	Inventive example
22B	22B	22	80	N	N	0.2	94.0	24	13	525	120	Inventive example
33A	33A	33	500	Y	Y	0.3	99.3	19	15	540	139	Inventive example
33	33	33	500	Y	N	0.3	99.3	19	15	579	81	Inventive example
33B	33B	33	500	N	N	0.3	99.3	19	15	588	80	Inventive example
B: bainite												

Claims

1. A steel material for a high-strength line pipe with high fracture toughness in hydrogen, the steel material comprising:
a chemical composition containing:

on a mass percent basis,
C: 0.02% to 0.15%,
Si: 0.01% to 2.0%,
Mn: 0.5% to 1.5%,
P: 0.0001% to 0.015%,
S: 0.0002% to 0.0015%,
Al: 0.005% to 0.15%,
O: 0.01% or less,
N: 0.010% or less,
Nb: 0.10% or less, and
H: 0.02 ppm or less, and
optionally at least one selected from
Ca: 0% to 0.005%,
Ni: 0% to 2.0%,
Ti: 0% to 0.1%,
Cu: 0% to 1.0%,
Cr: 0% to 1.0%,
Mo: 0% to 0.60%,
W: 0% to 1.0%,
V: 0% to 0.10%,
Zr: 0% to 0.050%,
Mg: 0% to 0.01%,
REM: 0% to 0.01%,
B: 0% to 0.0020%,
Ta: 0% to 0.2%,
Hf: 0% to 0.2%,
Re: 0% to 0.005%,
Sn: 0% to 0.3%, and
Sb: 0% to 0.3%,
the remainder being Fe and an incidental impurity element,
a metallic microstructure containing bainite and 15 pieces/100 mm^2 or less of inclusions having an aspect ratio of 2.0 or more and a length of 10 μm or more, the bainite in a range from a surface to a middle of a thickness of the steel material having a maximum grain size of 25 μm or less,
tensile strength of 520 MPa or more, and
a hydrogen-induced crack growth threshold K_{IH} in a high-pressure hydrogen gas environment of 1 MPa or more of 80 MPa $\cdot m^{1/2}$ or more.

2. The steel material for a high-strength line pipe with high fracture toughness in hydrogen according to Claim 1, wherein the chemical composition contains, on a mass percent basis,

Ca: 0.0001% to 0.005%,
Ni: 0.01% to 2.0%,
Ti: 0.005% to 0.1%,
Cu: 0.01% to 1.0%,
Cr: 0.01% to 1.0%,
Mo: 0.01% to 0.60%,
W: 0.01% to 1.0%,
V: 0.01% to 0.10%,
Zr: 0.0001% to 0.050%,
Mg: 0.0001% to 0.01%,
REM: 0.0001% to 0.01%,
B: 0.0001% to 0.0020%,
Ta: 0.0001% to 0.2%,

Hf: 0.0001% to 0.2%,
Re: 0.0001% to 0.005%,
Sn: 0.0001% to 0.3%, and
Sb: 0.0001% to 0.3%.

3. The steel material for a high-strength line pipe with high fracture toughness in hydrogen according to Claim 1 or 2, wherein an area fraction of retained austenite is 0% to 3% by area, and an area fraction of the bainite in the range from the surface to the middle of the thickness of the steel material is 90% or more.

4. A method for producing a steel material for a high-strength line pipe with high fracture toughness in hydrogen, the method comprising:

a heating step of heating a cast steel having the chemical composition according to Claim 1 or 2 at 1000°C to 1250°C;
a hot rolling step of rolling the cast steel heated in the heating step under conditions in which a total rolling reduction in a recrystallization temperature range is 35% or more and 55% or less, a rolling reduction in a final rolling pass in the recrystallization temperature range is 10% or more, a rolling reduction in a final rolling pass at (recrystallization temperature - 80°C) or more is 15% or more, and a finish rolling temperature is an Ar_3 transformation point or higher in terms of a temperature at a surface of a steel sheet; and
a controlled cooling step of cooling a hot-rolled steel sheet produced in the hot rolling step under conditions in which a cooling start temperature is the Ar_3 transformation point or higher in terms of a temperature at the surface of the hot-rolled steel sheet, a cooling start time difference between a front end and a rear end of the hot-rolled steel sheet is 50 seconds or less, an average cooling rate from 750°C to 550°C ranges from 15°C/s to 50°C/s in terms of a temperature at a middle of a thickness of the steel sheet, and a cooling stop temperature ranges from 250°C to 650°C.

5. A steel pipe for a high-strength line pipe with high fracture toughness in hydrogen, the steel pipe comprising:
a chemical composition containing:

on a mass percent basis,
C: 0.02% to 0.15%,
Si: 0.01% to 2.0%,
Mn: 0.5% to 1.5%,
P: 0.0001% to 0.015%,
S: 0.0002% to 0.0015%,
Al: 0.005% to 0.15%,
O: 0.01% or less,
N: 0.010% or less,
Nb: 0.10% or less, and
H: 0.02 ppm or less, and
optionally at least one selected from
Ca: 0% to 0.005%,
Ni: 0% to 2.0%,
Ti: 0% to 0.1%,
Cu: 0% to 1.0%,
Cr: 0% to 1.0%,
Mo: 0% to 0.60%,
W: 0% to 1.0%,
V: 0% to 0.10%,
Zr: 0% to 0.050%,
Mg: 0% to 0.01%,
REM: 0% to 0.01%,
B: 0% to 0.0020%,
Ta: 0% to 0.2%,
Hf: 0% to 0.2%,
Re: 0% to 0.005%,
Sn: 0% to 0.3%, and
Sb: 0% to 0.3%,

the remainder being Fe and an incidental impurity element,
a metallic microstructure containing bainite and 15 pieces/100 mm^2 or less of inclusions having an aspect ratio of 2.0 or more and a length of 10 μm or more, the bainite in a range from an inner surface to a middle of a thickness of the steel pipe having a maximum grain size of 25 μm or less,
tensile strength of 520 MPa or more, and
a hydrogen-induced crack growth threshold K_{IH} in a high-pressure hydrogen gas environment of 1 MPa or more of 80 $MP_a \cdot m^{1/2}$ or more.

6. The steel pipe for a high-strength line pipe with high fracture toughness in hydrogen according to Claim 5, wherein the chemical composition contains, on a mass percent basis,

Ca: 0.0001% to 0.005%,
Ni: 0.01% to 2.0%,
Ti: 0.005% to 0.1%,
Cu: 0.01% to 1.0%,
Cr: 0.01% to 1.0%,
Mo: 0.01% to 0.60%,
W: 0.01% to 1.0%,
V: 0.01% to 0.10%,
Zr: 0.0001% to 0.050%,
Mg: 0.0001% to 0.01%,
REM: 0.0001% to 0.01%,
B: 0.0001% to 0.0020%,
Ta: 0.0001% to 0.2%,
Hf: 0.0001% to 0.2%,
Re: 0.0001% to 0.005%,
Sn: 0.0001% to 0.3%, and
Sb: 0.0001% to 0.3%.

7. The steel pipe for a high-strength line pipe with high fracture toughness in hydrogen according to Claim 5 or 6, wherein an area fraction of retained austenite is 0% to 3%, and an area fraction of the bainite in the range from the inner surface to the middle of the thickness of the steel pipe is 90% or more.

8. A method for producing a steel pipe for a high-strength line pipe with high fracture toughness in hydrogen, the method comprising:

a heating step of heating a cast steel having the chemical composition according to Claim 5 or 6 at 1000°C to 1250°C;
a hot rolling step of rolling the cast steel heated in the heating step under conditions in which a total rolling reduction in a recrystallization temperature range is 35% or more and 55% or less, a rolling reduction in a final rolling pass in the recrystallization temperature range is 10% or more, a rolling reduction in a final rolling pass at (recrystallization temperature - 80°C) or more is 15% or more, and a finish rolling temperature is an Ar_3 transformation point or higher in terms of a temperature of a surface of a steel sheet;
a controlled cooling step of cooling a hot-rolled steel sheet produced in the hot rolling step under conditions in which a cooling start temperature is the Ar_3 transformation point or higher in terms of a temperature at a surface of the hot-rolled steel sheet, a cooling start time difference between a front end and a rear end of the hot-rolled steel sheet is 50 seconds or less, an average cooling rate from 750°C to 550°C ranges from 15°C/s to 50°C/s in terms of a temperature at a middle of a thickness of the steel sheet, and a cooling stop temperature ranges from 250°C to 650°C; and
any one of a pipe production step of bending the hot-rolled steel sheet and butt-welding both end portions thereof after the controlled cooling step and a pipe production step of forming the hot-rolled steel sheet into a cylindrical shape by cold roll forming and subjecting both circumferential end portions of the cylindrical shape to butt electric resistance welding after the controlled cooling step.

INTERNATIONAL SEARCH REPORT

International application No.
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A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/00(2006.01)i; ***C21D 8/02***(2006.01)i; ***C21D 8/10***(2006.01)i; ***C22C 38/60***(2006.01)i
FI: C22C38/00 301F; C22C38/00 301Z; C22C38/60; C21D8/02 C; C21D8/10 C

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)

C22C38/00-38/60; C21D8/02; C21D8/10

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

Published examined utility model applications of Japan 1922-1996
Published unexamined utility model applications of Japan 1971-2023
Registered utility model specifications of Japan 1996-2023
Published registered utility model applications of Japan 1994-2023

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.
A	WO 2020/137812 A1 (JFE STEEL CORPORATION) 02 July 2020 (2020-07-02) claims	1-8
A	JP 2022-68942 A (JFE STEEL CORPORATION) 11 May 2022 (2022-05-11) claims	1-8
A	WO 2014/156187 A1 (JFE STEEL CORPORATION) 02 October 2014 (2014-10-02) claims	1-8
A	JP 2012-122103 A (SUMITOMO METAL IND., LIMITED) 28 June 2012 (2012-06-28) claims	1-8
A	WO 2022/030818 A1 (POSCO) 10 February 2022 (2022-02-10) claims	1-8
A	JP 2018-12855 A (NIPPON STEEL & SUMITOMO METAL CORPORATION) 25 January 2018 (2018-01-25) claims	1-8

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

* Special categories of cited documents:
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"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention
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"&" document member of the same patent family

Date of the actual completion of the international search	Date of mailing of the international search report
07 December 2023	**19 December 2023**
Name and mailing address of the ISA/JP **Japan Patent Office (ISA/JP)** **3-4-3 Kasumigaseki, Chiyoda-ku, Tokyo 100-8915** **Japan**	Authorized officer Telephone No.

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INTERNATIONAL SEARCH REPORT

International application No.
PCT/JP2023/035560

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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Information on patent family members

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