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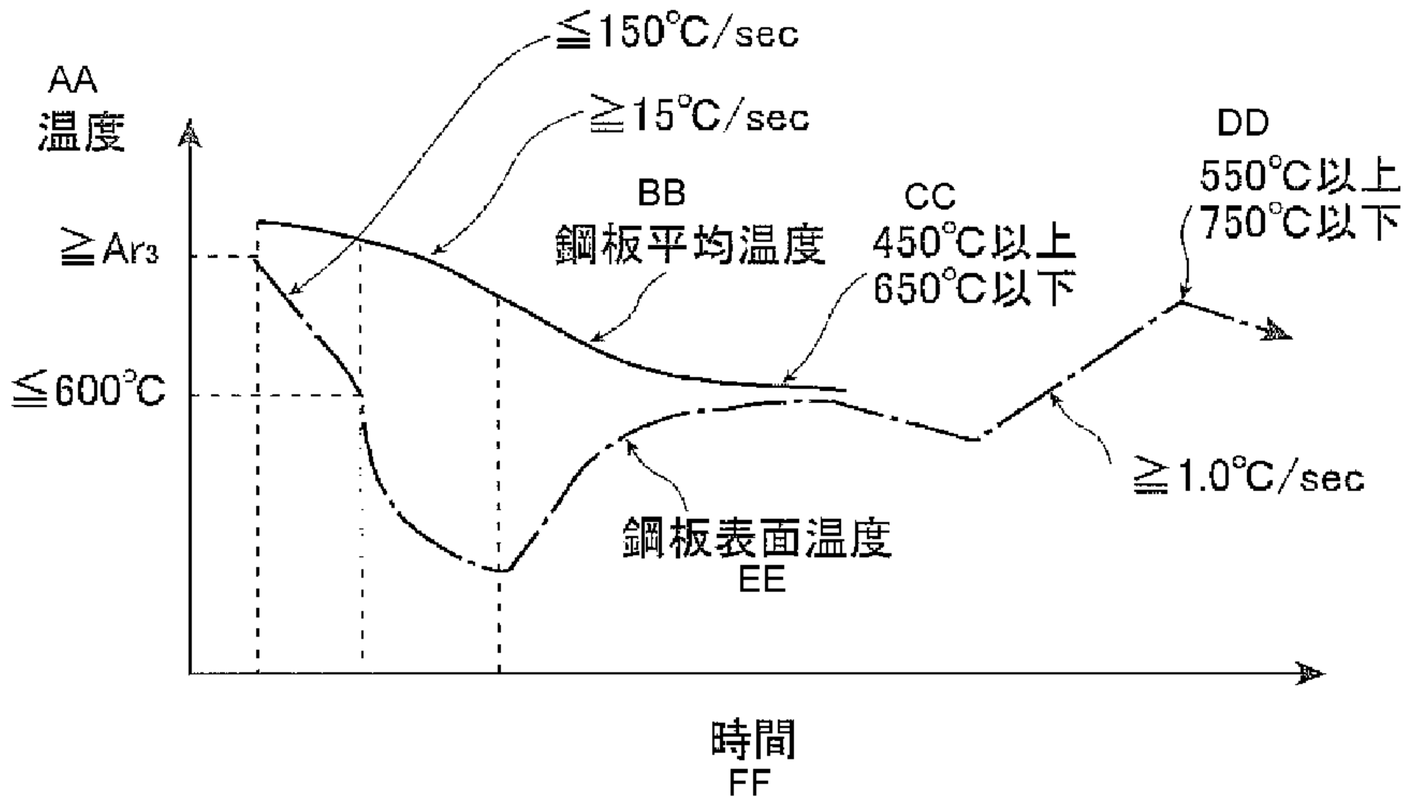
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(54) Titre : PLAQUE D'ACIER A HAUTE RESISTANCE, A FAIBLE RAPPORT D'ELASTICITE, AYANT UNE RESISTANCE SUPERIEURE AU VIEILLISSEMENT APRES DEFORMATION, SON PROCEDE DE FABRICATION, ET TUYAU EN ACIER SOUDE A HAUTE RESISTANCE UTILISANT CETTE PLAQUE

(54) Title: HIGH STRENGTH STEEL PLATE HAVING LOW YIELD RATIO EXCELLENT IN TERMS OF STRAIN AGEING RESISTANCE, METHOD FOR MANUFACTURING THE SAME AND HIGH STRENGTH WELDED STEEL PIPE MADE OF THE SAME



AA Temperature
BB Average temperature of steel plate
CC 450° C to 650° C
DD 550° C to 750° C
EE Surface temperature of steel plate
FF Time

(57) Abrégé/Abstract:
The present invention provides a low yield ratio high-strength steel plate that is API 5L X70 grade or lower, a production method for said steel plate, and a high-strength welded steel pipe that employs the steel plate. The low yield ratio high-strength steel plate of

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(13) A1

(57) Abrégé(suite)/Abstract(continued):
the present invention is characterized in that: said steel plate has a predetermined component composition; the metal structure of the steel plate is a two-phase structure comprising bainite and island-shaped martensite; the island-shaped martensite has an area fraction of 3-15% and an equivalent circular diameter of 3.0 μm or less, and the remaining portion is bainite; the hardness variation in the plate thickness direction is ΔHV30 or less; the hardness variation in the plate width direction is ΔHV30 or less; the maximum hardness of the surface layer of the steel plate is HV230 or less; and in that the yield ratio is 85% or less and elongation is 22% or more according to a full thickness tensile test using a test fragment having a shape defined by the GOST standard.

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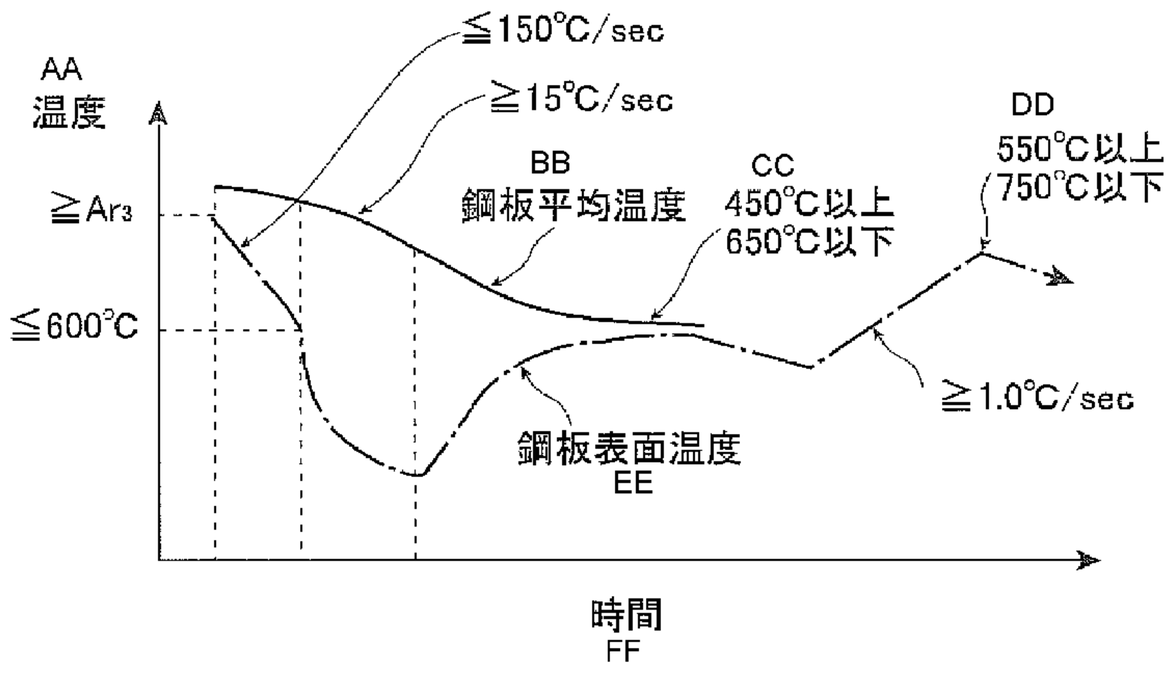
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[続葉有]

(54) **Title**: LOW YIELD RATIO HIGH-STRENGTH STEEL PLATE HAVING SUPERIOR STRAIN AGING RESISTANCE, PRODUCTION METHOD THEREFOR, AND HIGH-STRENGTH WELDED STEEL PIPE USING SAME

(54) 発明の名称 : 耐歪時効特性に優れた低降伏比高強度鋼板およびその製造方法並びにそれを用いた高強度溶接鋼管



AA Temperature
BB Average temperature of steel plate
CC 450° C to 650° C
DD 550° C to 750° C
EE Surface temperature of steel plate
FF Time

(57) **Abstract**: The present invention provides a low yield ratio high-strength steel plate that is API 5L X70 grade or lower, a production method for said steel plate, and a high-strength welded steel pipe that employs the steel plate. The low yield ratio high-strength steel plate of the present invention is characterized in that: said steel plate has a predetermined component composition; the metal structure of the steel plate is a two-phase structure comprising bainite and island-shaped martensite; the island-shaped martensite has an area fraction of 3-15% and an equivalent circular diameter of 3.0 μm or less, and the remaining portion is bainite; the hardness variation in the plate thickness direction is ΔHV30 or less; the hardness variation in the plate width direction is ΔHV30 or less; the maximum hardness of the surface layer of the steel plate is HV230 or less; and in that the yield ratio is 85% or less and elongation is 22% or more according to a full thickness tensile test using a test fragment having a shape defined by the GOST standard.

(57) 要約:

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本発明は、ＡＰＩ ５Ｌ Ｘ７０グレード以下の低降伏比高強度鋼板及びその製造方法並びにその鋼板を用いた高強度溶接鋼管を提供するものである。 本発明の低降伏比高強度鋼板は、その成分組成を所定のものとし、その金属組織を、ベイナイトと島状マルテンサイトとの２相組織として、前記島状マルテンサイトの面積分率は３～１５％で円相当径は３．０μｍ以下、残部は前記のベイナイトとし、板厚方向の硬さのばらつきはΔＨＶ３０以下、板幅方向の硬さのばらつきはΔＨＶ３０以下、鋼板表層部の最高硬さはＨＶ２３０以下とし、ＧＯＳＴ規格で定められた形状の試験片を用いた全厚引張試験における降伏比は８５％以下で伸びは２２％以上としたことを特徴とするものである。

2014-30/228

DESCRIPTION

[Title of Invention]

HIGH STRENGTH STEEL PLATE HAVING LOW YIELD RATIO EXCELLENT IN TERMS OF STRAIN AGEING RESISTANCE, METHOD FOR MANUFACTURING THE SAME AND HIGH STRENGTH WELDED STEEL PIPE MADE OF THE SAME

[Technical Field]

[0001]

The present invention relates to a steel plate having a low yield ratio, high strength and high toughness which is preferably used mainly in a linepipe field, a method for manufacturing the steel plate and a high strength steel pipe. In particular, the present invention relates to a steel plate having a low yield ratio, high strength and high toughness excellent in terms of strain ageing resistance, a method for manufacturing the steel plate and a high strength welded steel pipe excellent in terms of buckling resistance and ductility which is made of the steel plate.

[Background Art]

[0002]

Nowadays, steel materials for welded structures are required to have not only high strength and high toughness but also a low yield ratio and high uniform elongation from the viewpoint of earthquake resistance. Generally, it is known that, by forming a metallographic structure of a steel

material in which hard phases such as a bainite phase (hereinafter, also referred to as β) and a martensite phase (hereinafter, also referred to as M) are appropriately dispersed in a ferrite phase (hereinafter, referred to as α) which is a soft phase, it is possible to achieve a decrease in the yield ratio of the steel material and an increase in the uniform elongation of the steel material.

[0003]

As an example of manufacturing methods for forming a microstructure in which hard phases are appropriately dispersed in a soft phase as described above, a certain method is described in Patent Literature 1. That is, Patent Literature 1 discloses a heat treatment method in which, as an intermediate treatment between quenching (hereinafter, also referred to as Q) and tempering (hereinafter, also referred to as T), quenching starting from a temperature range for forming a dual phase consisting of a ferrite phase and an austenite phase (hereinafter, also referred to as γ) is performed (hereinafter, also referred to as Q').

[0004]

Patent Literature 2 discloses an example of a method which does not require an additional manufacturing process, in which, after rolling has been finished at a temperature equal to or higher than the Ar_3 transformation point, the start of accelerated cooling is delayed until the steel

material has a temperature equal to or lower than the Ar_3 transformation point at which a ferrite phase is formed.

[0005]

As an example of a technique with which it is possible to achieve a decrease in yield ratio without performing complicated heat treatments as disclosed in Patent Literature 1 and Patent Literature 2, Patent Literature 3 discloses a method with which a decrease in yield ratio is achieved by finishing rolling of a steel material at a temperature equal to or higher than the Ar_3 transformation point, and then by controlling an accelerated cooling rate and a cooling stop temperature to form a dual-phase structure consisting of an acicular ferrite phase and a martensite phase.

[0006]

Moreover, Patent Literature 4 discloses an example of a technique with which it is possible to achieve a low yield ratio and excellent toughness in a weld heat affected zone without significantly increasing the contents of alloying chemical elements of a steel material, in which, by controlling Ti/N and a Ca-O-S balance, a three-phase structure consisting of a ferrite phase, a bainite phase and island martensite (hereinafter, also referred to as MA) is formed.

[0007]

In addition, Patent Literature 5 discloses a technique in which a decrease in yield ratio and an increase in uniform elongation are achieved by adding alloying chemical elements such as Cu, Ni and Mo.

[0008]

On the other hand, in the case of a welded steel pipe, which is used to form a linepipe, such as a UOE steel pipe or an electric resistance welded steel pipe, after a steel plate has been subjected to cold forming into a cylindrical shape and the butted portions have been welded, usually, a coating treatment such as polyethylene coating or powder epoxy coating is performed on the outer surface of the resultant steel pipe from the viewpoint of, for example, corrosion protection. Therefore, there is a problem in that, since strain ageing occurs due to working strain applied when pipe forming is performed and due to heating when the coating treatment is performed, there is an increase in yield stress, which results in the yield ratio of the steel pipe being larger than that of the steel plate.

[0009]

In order to solve the problem described above, for example, Patent Literatures 6 and 7 disclose steel pipes having a low yield ratio, high strength and high toughness excellent in terms of strain ageing resistance and methods for manufacturing the steel pipes utilizing the fine

precipitations of complex carbides containing Ti and Mo or the fine precipitations of complex carbides containing two or all of Ti, Nb, and V.

[Citation List]

[Patent Literature]

[0010]

[PTL 1] Japanese Unexamined Patent Application Publication No. 55-97425

[PTL 2] Japanese Unexamined Patent Application Publication No. 55-41927

[PTL 3] Japanese Unexamined Patent Application Publication No. 1-176027

[PTL 4] Japanese Patent No. 4066905

[PTL 5] Japanese Unexamined Patent Application Publication No. 2008-248328

[PTL 6] Japanese Unexamined Patent Application Publication No. 2005-60839

[PTL 7] Japanese Unexamined Patent Application Publication No. 2005-60840

[Summary of Invention]

[Technical Problem]

[0011]

In the case of the heat treatment method according to Patent Literature 1, by appropriately selecting a quenching

temperature in a range for forming a dual phase, it is possible to achieve a decrease in yield ratio, but there is a problem in that there is a decrease in productivity and there is an increase in manufacturing cost due to an increase in the number of heat treatment processes.

In addition, in the case of the technique according to Patent Literature 2, there is a problem in that, since it is necessary to perform cooling at a cooling rate equivalent to that of a natural cooling in a temperature range from a rolling finish temperature to an accelerated cooling start temperature, there is a significant decrease in productivity.

[0012]

Moreover, in the case of the technique according to Patent Literature 3, as indicated by the examples in the literature, there is a problem in that, since the carbon content or the contents of other alloying chemical elements of a steel plate are increased in order to obtain a steel material having a tensile strength of 490 N/mm^2 (50 kg/mm^2) or more, there is an increase in material cost, and, in addition to that, there is a decrease in toughness in a weld heat affected zone.

In addition, in the case of the technique according to Patent Literature 4, the influence of, for example, a microstructure on uniform elongation which is required in use for, for example, a linepipe has not necessarily been

clarified. In addition, since the low-temperature toughness of a base metal was evaluated only at a temperature of -10°C, it is not clear for a possibility of a new use application in which toughness at a lower temperature is required.

[0013]

In the case of the technique according to Patent Literature 5, since it is necessary that the steel plate has a chemical composition containing increased contents of alloying chemical elements, there is a problem in that there is an increase in material cost, and, in addition to that, there is a decrease in toughness in a weld heat affected zone. In addition, the low-temperature toughness of a base metal and a weld heat affected zone are evaluated only at a temperature of -10°C.

In the case of techniques according to Patent Literatures 6 and 7, although there is an improvement in strain ageing resistance, the low-temperature toughness of a base metal and a weld heat affected zone are evaluated only at a temperature of -10°C.

[0014]

Moreover, in the case of Patent Literatures 1 through 7, it is necessary to form a ferrite phase but the ferrite phase causes a decrease in tensile strength. Therefore, it is necessary to add alloying chemical elements in order to increase strength to an X60 grade or more according to the

API standard, and there is concern that there may be an increase in alloy cost and there may be a decrease in low-temperature toughness.

[0015]

Therefore, an object of the present invention is, by solving such problems in conventional techniques, to provide a steel plate having a low yield ratio, high strength and high toughness excellent in terms of strain ageing resistance of an API 5L X70 grade or less which can be manufactured at high productivity, a method for manufacturing the steel plate and a high strength welded steel pipe made of the steel plate.

[Solution to Problem]

[0016]

In order to solve the problems described above, the present inventors diligently conducted investigations regarding a method for manufacturing a steel plate, in particular, regarding a manufacturing process including controlled rolling, accelerated cooling after controlled rolling has been performed and subsequent reheating. From the results of the investigations, the present inventors found that, by controlling rolling conditions so that a bainite phase in the surface portion of the steel plate becomes soft, it is possible to achieve high deformation performance and high ductility without a significant

decrease in strength, and obtained the following findings.

[0017]

(a) By stopping accelerated cooling in the middle of bainite transformation of a steel plate, that is, in a temperature range in which a non-transformed austenite phase is present in an accelerated cooling process, and by subsequently starting reheating from a temperature higher than a temperature (hereinafter, referred to as a Bf point) at which bainite transformation finishes, the metallographic structure of the steel plate, in which hard island martensite (hereinafter, referred to as MA) is uniformly formed in a bainite phase, is formed, and thereby a decrease in yield ratio is achieved.

[0018]

(b) By adding an appropriate amount of Mn to the chemical composition of steel as a chemical element for stabilizing an austenite phase, a non-transformed austenite phase become stabilized, and thus, it is possible to form hard MA without adding a large amount of chemical elements for increasing hardenability such as Cu, Ni, and Mo.

[0019]

(c) By performing rolling under conditions such that the cumulative rolling reduction ratio is 50% or more in a temperature range of 900°C or lower which is a no-recrystallization temperature range in austenite, fine MA

can be uniformly dispersed, and thereby an increase in uniform elongation can be achieved, while a low yield ratio is maintained.

[0020]

(d) Moreover, by appropriately controlling both the rolling conditions in the temperature range which is a no-recrystallization temperature range in austenite described in item (c) above and the reheating conditions described in item (a) above, the shape of MA can be controlled. That is, the grain size of MA can be controlled to be as small as 3.0 μm or less in terms of an average value of equivalent circle diameters. Then, as a result, since MA is only slightly decomposed even if steel is subjected to a thermal history which causes deterioration of a yield ratio due to ageing in the case of a conventional steel, the desired microstructure and properties can be maintained even after ageing has occurred.

[0021]

(e) Moreover, by performing first stage cooling at a cooling rate of 150°C/sec. or less down to a temperature of 600°C or lower, a bainite structure in the upper and lower surface portions can be softened. As a result, since there is a decrease in hardness in the upper and lower surface portions of a steel plate, high ductility can be achieved.

[0022]

The present invention has been completed on the basis of the findings described above and further investigations, and the subject matter of the present invention is as follows.

[1] A high strength steel plate having a low yield ratio, the steel plate having a chemical composition containing, by mass%, C: 0.03% or more and 0.08% or less, Si: 0.01% or more and 1.0% or less, Mn: 1.2% or more and 3.0% or less, P: 0.015% or less, S: 0.005% or less, Al: 0.08% or less, Nb: 0.005% or more and 0.07% or less, Ti: 0.005% or more and 0.025% or less, N: 0.010% or less, O: 0.005% or less and the balance being Fe and inevitable impurities, a metallographic structure being a dual-phase structure consisting of a bainite phase and island martensite, in which the area fraction of the island martensite is 3% or more and 15% or less, in which the equivalent circle diameter of the island martensite is 3.0 μm or less, and in which the remainder of the metallographic structure is a bainite phase, a hardness variation in the thickness direction of ΔHV30 or less in terms of Vickers hardness, a hardness variation in the width direction of ΔHV30 or less in terms of Vickers hardness, a maximum hardness in the surface portions of the steel plate of HV230 or less in terms of Vickers hardness and a yield ratio of 85% or less and an elongation of 22% or more in a full-

thickness tensile test using a test piece having a shape in accordance with GOST standards.

[0023]

[2] The high strength steel plate having a low yield ratio according to item [1] above, the steel plate having the chemical composition further containing, by mass%, one or more selected from among Cu: 0.5% or less, Ni: 1% or less, Cr: 0.5% or less, Mo: 0.5% or less, V: 0.1% or less, Ca: 0.0005% or more and 0.003% or less and B: 0.005% or less.

[0024]

[3] A method for manufacturing a high strength steel plate having a low yield ratio excellent in terms of strain ageing resistance, the method including heating steel having the chemical composition according to item [1] or [2] above at a temperature of 1000°C or higher and 1300°C or lower, performing hot rolling under conditions such that the cumulative rolling reduction ratio is 50% or more in a temperature range of 900°C or lower, finishing hot rolling at a temperature equal to or higher than the Ar_3 transformation point, starting cooling at a temperature equal to or higher than the Ar_3 transformation point, performing cooling at a cooling rate of 150°C/sec. or less in terms of the surface temperature of the steel plate until the surface temperature becomes 600°C or lower, performing cooling at a cooling rate of 15°C/sec. or more in terms of

the average temperature of the steel plate until the average temperature becomes 450°C or higher and 650°C or lower, and, immediately after the cooling has been performed, performing reheating at a heating rate of 1.0°C/sec. or more in terms of the surface temperature of the steel plate until the surface temperature becomes 550°C or higher and 750°C or lower.

[0025]

[4] A high strength welded steel pipe, the steel pipe being manufactured by forming the steel plate according to item [1] or [2] above into a cylindrical shape, by welding butted portions of the shaped cylinder under conditions such that single-layer welding is performed on each of the outer and inner surfaces, and by thereafter performing a pipe-expanding treatment under a condition of a pipe expanding ratio of 0.4% or more and 2.0% or less, and having a yield ratio of 90% or less and an elongation of 20% or more in a full-thickness tensile test using a test piece having a shape in accordance with GOST standards, and, further having a yield ratio of 90% or less and an elongation of 20% or more even after a strain ageing treatment has been performed under conditions such that the temperature is 250°C or lower and the duration is 30 minutes or less.

[Advantageous Effects of Invention]

[0026]

According to the present invention, a steel plate having a low yield ratio, high strength and high toughness excellent in terms of strain ageing resistance can be manufactured without decreasing toughness in a weld heat affected zone or adding a large amount of alloying chemical elements. Therefore, a steel plate which is used mainly for a linepipe can be stably manufactured in a large amount, and there is a significant increase in productivity and economic efficiency. Moreover, since a high strength welded steel pipe excellent in terms of buckling resistance and ductility can be manufactured using this steel plate, there is a significant industrial effect.

[Brief Description of Drawings]

[0027]

[Fig. 1] Fig. 1 is a schematic diagram illustrating a thermal history to which a steel plate is subjected after accelerated cooling according to the present invention has been started, in which a solid line represents the average temperature of the steel plate and a dot-dash line represents the surface temperature of the steel plate.

[Description of Embodiments]

[0028]

The reason for limitations on the features of the present invention will be described hereafter.

[0029]

1. Regarding chemical composition

Firstly, the reason for the limitations on the chemical composition of the steel according to the present invention will be described. Here, % used when describing a chemical composition always represents mass%.

[0030]

C: 0.03% or more and 0.08% or less

C is a chemical element which contributes by forming carbide to precipitation strengthening and is important for forming MA. In the case where the C content is less than 0.03%, the content is not sufficient to form MA and sufficient strength cannot be achieved. In the case where the C content is more than 0.08%, there is a decrease in the toughness of a base metal and toughness in a weld heat affected zone (HAZ). Therefore, the C content is set to be 0.03% or more and 0.08% or less, preferably 0.04% or more and 0.06% or less.

[0031]

Si: 0.01% or more and 1.0% or less

Si is added in order to perform deoxidation. In the case where the Si content is less than 0.01%, there is an insufficient effect of deoxidation, and, in the case where the Si content is more than 1.0%, there is a decrease in toughness and weldability. Therefore, the Si content is set to be 0.01% or more and 1.0% or less, preferably 0.01% or

more and 0.3% or less.

[0032]

Mn: 1.2% or more and 3.0% or less

Mn is added in order to increase strength and toughness. Mn is also added in order to increase hardenability that promotes formation of MA. In the case where the Mn content is less than 1.2%, these effects cannot be sufficiently obtained, and, in the case where the Mn content is more than 3.0%, there is a decrease in toughness and weldability. Therefore, the Mn content is set to be 1.2% or more and 3.0% or less, preferably 1.8% or more and 3.0% or less in order to stably form MA regardless of variations in a chemical composition and manufacturing conditions.

[0033]

P: 0.015% or less

In the present invention, P is an inevitable impurity, and the upper limit of the P content is specified. In the case where the P content is large, there is a significant increase in the degree of central segregation, resulting in a decrease in the toughness of a base metal. Therefore, the P content is set to be 0.015% or less, preferably 0.010% or less.

[0034]

S: 0.005% or less

In the present invention, S is an inevitable impurity,

and the upper limit of the S content is specified. In the case where the S content is large, there is a significant increase in the amount of MnS formed, resulting in a decrease in the toughness of a base metal. Therefore, the S content is set to be 0.005% or less, preferably 0.002% or less.

[0035]

Al: 0.08% or less

Al is added as a deoxidation agent. In the case where the Al content is more than 0.08%, there is a decrease in the cleanliness of steel, resulting in a decrease in toughness. Therefore, the Al content is set to be 0.08% or less, preferably 0.01% or more and 0.08% or less, more preferably 0.01% or more and 0.05% or less.

[0036]

Nb: 0.005% or more and 0.07% or less

Nb is a chemical element which increases toughness as a result of decreasing a grain size in a microstructure and contributes to an increase in strength due to an increase in hardenability through solid solute Nb. These effects are obtained in the case where the Nb content is 0.005% or more. However, in the case the Nb content is more than 0.07%, there is a decrease in toughness in a weld heat affected zone, and thus the Nb content is set to be 0.005% or more and 0.07% or less, more preferably 0.01% or more and 0.05%

or less.

[0037]

Ti: 0.005% or more and 0.025% or less

Ti is an important chemical element which increases the toughness of a base metal as a result of preventing an increase in the grain size of an austenite phase through the pinning effect of TiN when slab heating is performed. This effect is obtained in the case where the Ti content is 0.005% or more. However, in the case where the Ti content is more than 0.025%, there is a decrease in toughness in a weld heat affected zone, and thus the Ti content is set to be 0.005% or more and 0.025% or less, preferably 0.005% or more and less than 0.02% from the view point of toughness in a weld heat affected zone, more preferably 0.007% or more and 0.016% or less.

[0038]

N: 0.010% or less

N is treated as an inevitable impurity. Since there is a decrease in toughness in a weld heat affected zone in the case where the N content is more than 0.010%, the N content is set to be 0.010% or less, preferably 0.007% or less, more preferably 0.006% or less.

[0039]

O: 0.005% or less

In the present invention, O is an inevitable impurity

and the upper limit of the O content is specified. Since O causes formation of coarse inclusions which has a negative effect on toughness, the O content is set to be 0.005% or less, more preferably 0.003% or less.

[0040]

The basic chemical composition in the present invention is as described above. Furthermore, in order to improve the strength and toughness of a steel plate and to increase hardenability to promote formation of MA, one or more of Cu, Ni, Cr, Mo, V, Ca, and B described below may be added.

[0041]

Cu: 0.5% or less

Cu may be added, since Cu contributes to an increase in the hardenability of steel in the case where Cu is added. In order to obtain this effect, it is preferable that the Cu content be 0.05% or more. However, in the case where the Cu content is more than 0.5%, there is a decrease in toughness. Thus, in the case where Cu is added, it is preferable that the Cu content be 0.5% or less, more preferably 0.4% or less.

[0042]

Ni: 1% or less

Since Ni contributes to an increase in the hardenability of steel, and, in particular, since there is not a decrease in toughness even in the case where the Ni content is large, Ni may be added. In order to obtain this

effect, it is preferable that the Ni content be 0.05% or more. However, since Ni is an expensive chemical element, in the case where Ni is added, it is preferable that the Ni content be 1% or less, more preferably 0.4% or less.

[0043]

Cr: 0.5% or less

Since Cr is, like Mn, a chemical element which is effective for achieving sufficient strength even in the case where the C content is low, Cr may be added. It is preferable that the Cr content be 0.1% or more in order to obtain this effect. However, since there is a decrease in weldability in the case where the Cr content is excessively large, in the case where Cr is added, it is preferable that the Cr content be 0.5% or less, more preferably 0.4% or less.

[0044]

Mo: 0.5% or less

Since Mo is a chemical element which increases hardenability, and since Mo is a chemical element which contributes to an increase in strength as a result of formation of MA and strengthening a bainite phase, Mo may be added. In order to obtain this effect, it is preferable that the Mo content be 0.05% or more. However, in the case where the Mo content is more than 0.5%, there is a decrease in toughness in a weld heat affected zone. Therefore, in the case where Mo is added, it is preferable that the Mo

content be 0.5% or less, more preferably 0.3% or less.

[0045]

V: 0.1% or less

Since V is a chemical element which contributes to an increase in strength with an increase in hardenability, V may be added. In order to obtain this effect, it is preferable that the V content be 0.005% or more. However, in the case where the V content is more than 0.1%, there is a decrease in toughness in a weld heat affected zone. Therefore, in the case where V is added, it is preferable that the V content be 0.1% or less, more preferably 0.06% or less.

[0046]

Ca: 0.0005% or more and 0.003% or less

Since Ca is a chemical element which increases toughness as a result of controlling the shape of sulfide containing inclusions, Ca may be added. This effect is obtained in the case where the Ca content is 0.0005% or more. In the case where the Ca content is more than 0.003%, this effect becomes saturated and there is conversely a decrease in toughness as a result of decreasing cleanliness. Therefore, in the case where Ca is added, it is preferable that the Ca content be 0.0005% or more and 0.003% or less, more preferably 0.001% or more and 0.003% or less.

[0047]

B: 0.005% or less

Since B is a chemical element which contributes to an increase in strength and an increase in toughness in a weld heat affected zone, B may be added. In order to obtain these effects, it is preferable that the B content be 0.0005% or more. However, in the case where the B content exceeds 0.005%, there is a decrease in weldability. Therefore, in the case where B is added, it is preferable that the B content be 0.005% or less, more preferably 0.003% or less.

[0048]

Incidentally, by optimizing the ratio of the Ti content and the N content Ti/N, it is possible to prevent an increase in the grain size of an austenite phase in a weld heat affected zone due to presence of TiN particles. Thus, it is possible to achieve good toughness in a weld heat affected zone. Therefore, it is preferable that Ti/N be 2 or more and 8 or less, more preferably 2 or more and 5 or less.

[0049]

The remainder of the chemical composition other than those described above in a steel plate according to the present invention, consists of Fe and inevitable impurities. However, as long as the effects of the present invention are not decreased, chemical elements other than those described

above may be added. For example, from the viewpoint of increasing toughness, Mg: 0.02% or less and/or REM (rare-earth metal): 0.02% or less may be added.

[0050]

Subsequently, the metallographic structure according to the present invention will be described.

[0051]

2. Regarding metallographic structure

In the present invention, a metallographic structure is controlled so that, in addition to a bainite phase, an island martensite (MA) is uniformly included in an amount of 3% or more and 15% or less in terms of area fraction.

[0052]

By forming a duplex-phase structure serving as a main body of the structure, in which hard MA is included in a soft bainite phase, a decrease in yield ratio and an increase in uniform elongation are obtained.

[0053]

In the case where steel plates and steel pipes are used in earthquake regions or the like in which a large deformation is to be applied to these materials, there are some cases where these materials are required to have not only a low yield ratio but also high uniform elongation and high ductility. In the case of a duplex-phase structure consisting of a soft bainite phase and hard MA as described

above, a soft phase undergoes deformation. In addition, by controlling hardness variations in the thickness direction and in the width direction to be about ΔHV30 or less and a maximum hardness in the surface portions of the steel plate to be about HV230 or less, it is possible to achieve a high elongation of 22% or more in a full-thickness tensile test using a test piece having a shape in accordance with GOST standards for a steel plate.

[0054]

The content ratio of MA in a metallographic structure is set to be 3% or more and 15% or less in terms of area fraction of MA (calculated as the average value of the area ratios of MA in arbitrary cross sections in the rolling direction, the width direction and the like of a steel plate). In the case where the area fraction of MA is less than 3%, there is a case where there is an insufficient decrease in yield ratio. In the case where the area fraction of MA is more than 15%, there is a case where there is a decrease in the toughness of a base metal.

[0055]

In addition, it is preferable that the area fraction of MA be 5% or more and 12% or less from the viewpoint of decreasing yield ratio and increasing uniform elongation and base-material toughness. More preferably, the area fraction of MA is 5% or more and 10% or less. Incidentally, the area

fraction of MA means the ratio with respect to the whole microstructure of steel.

[0056]

MA can be easily identified by observing a sample prepared by etching a steel plate using, for example, a 3% nital solution (nital: nitric acid alcohol solution) and by subsequently performing electrolytic etching. By observing the microstructure of a steel plate using a scanning electron microscope (SEM), MA is recognized as a distinct white portion.

[0057]

Incidentally, the area fraction of MA can be calculated, for example, as the average value of the area ratios of MA by performing image processing on microstructure photographs of at least 4 microscopic fields taken through observation using a scanning electron microscope (SEM).

[0058]

In addition, the equivalent circle diameter of MA is set to be 3.0 μm or less from the viewpoint of achieving sufficient toughness for a base metal and increasing uniform elongation. This is because, in the case where the equivalent circle diameter of MA is more than 3.0 μm, there is a case where there is a decrease in the toughness of a base metal.

Incidentally, the equivalent circle diameter of MA can

be calculated as the average value of the diameters of the circles respectively having the same areas as MA grains obtained by performing image processing on microstructure taken through observation using a SEM.

[0059]

In the present invention, Mn and Si are added in order to form MA without adding large amounts of expensive alloying chemical elements such as Cu, Ni, and Mo. It is important to stabilize a non-transformed austenite phase with this method in order to suppress pearlite transformation or formation of a cementite phase during reheating and when air-cooling is performed after reheating has been performed.

[0060]

The mechanism according to the present invention for formation of MA and softening of a bainite phase in the surface portions will be roughly described hereafter. Manufacturing conditions will be described in detail later.

After having heated a slab, rolling is finished in a temperature range for forming an austenite phase, and then accelerated cooling is started at a temperature equal to or higher than the Ar_3 transformation point from the viewpoint of suppressing formation of a ferrite phase.

In a manufacturing process, in which, after accelerated cooling has been finished in the middle of a bainite

transformation process, that is, in a temperature range in which a non-transformed austenite phase is present, reheating is started at a temperature higher than a temperature (Bf point) at which bainite transformation is finished, and then cooling is performed, changes in a microstructure will be described hereafter.

[0061]

When accelerated cooling is finished, the microstructure consists of a bainite phase and a non-transformed austenite phase. After that, by performing reheating starting at a temperature higher than the Bf point, transformation from a non-transformed austenite phase to a bainite phase occurs. Since the solid solubility limit of C in a bainite phase is small, C is evacuated into the surrounding non-transformed austenite phase. Therefore, the C content in a non-transformed austenite phase increases as bainite transformation progresses when reheating is performed. At this time, in the case where austenite stabilizing chemical elements such as Cu and Ni are contained in certain amounts or more, a non-transformed austenite phase in which C is concentrated are retained even after reheating has been finished and transforms into MA when cooling is performed after reheating has been performed, and finally, a microstructure in which MA is formed in a bainite structure is formed.

[0062]

In the present invention, it is important to start reheating in a temperature range in which a non-transformed austenite phase is present after accelerated cooling has been performed. In the case where the reheating-start temperature is equal to or lower than the Bf point, bainite transformation is completed, and a non-transformed austenite phase is not present. Therefore, it is necessary that the reheating-start temperature be higher than the Bf point.

[0063]

In addition, there is no particular limitation on cooling after reheating has been performed because it has no influence on MA transformation. However, it is basically preferable that air-cooling be used. In the present invention, by using steel containing a certain amount of Mn, by stopping accelerated cooling in the middle of a bainite transforming process and by subsequently starting reheating immediately after the accelerated cooling has been stopped, hard MA can be formed without decreasing productivity.

[0064]

Incidentally, although the steel according to the present invention has a metallographic structure in which a certain amount of MA is uniformly included in a bainite phase, the present invention may include a steel plate having other types of microstructures and precipitates as

long as the effects of the present invention is not decreased.

[0065]

Specifically, in the case where one, two or more of other phases such as a pearlite phase and a cementite phase are mixed in addition, there is a decrease in strength. However, in the case where the area fractions of the microstructures other than a bainite phase and MA is small, the effect on decrease in strength is negligible. Therefore, as long as the total area fraction of a metallographic structures other than a bainite phase and MA, that is, a pearlite phase, a cementite phase and the like is 3% or less with respect to the whole microstructure, one or more of these metallographic structures may be included.

[0066]

The metallographic structure described above can be formed using steel having the chemical composition described above and the manufacturing methods described hereafter.

[0067]

3. Regarding manufacturing conditions

It is preferable that steel having the chemical composition described above be produced with a common method using melting and refining means such as a converter or an electric furnace and be made into a steel material such as a slab using a common method such as a continuous casting

method or an ingot casting-slabbing method. Incidentally, a melting and refining method and a casting method are not limited to those described above. The steel material is then rolled into a desired shape, and cooling and heating are performed after rolling has been performed.

[0068]

Incidentally, in the present invention, a cooling start temperature and a reheating temperature are expressed in terms of the surface temperature of a steel plate, and a cooling rate and a cooling stop temperature are expressed in terms of both the surface temperature of a steel plate and the average temperature of a steel plate, unless otherwise noted. Other temperatures such as a slab heating temperature, a controlled rolling start temperature, and a controlled rolling finish temperature are expressed in terms of the average temperature of a steel plate.

[0069]

Here, the average temperature of a steel plate is calculated from the surface temperature of a slab or a steel plate in consideration of parameters such as thickness and thermal conductivity. In addition, a cooling rate is an average cooling rate which is derived by dividing a temperature difference necessary for cooling from a temperature after hot rolling has been finished down to a cooling stop temperature (450°C to 650°C) by a time spent

for the cooling.

[0070]

In addition, a heating rate is an average heating rate which is derived by dividing a temperature difference necessary for reheating from a temperature after cooling has been finished up to a reheating temperature (550°C to 750°C) by a time spent for the reheating. Manufacturing conditions will be described in detail hereafter.

[0071]

Incidentally, as the Ar_3 transformation temperature, the value which is derived using the following equation will be used:

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

where an atomic symbol represents the mass% of the chemical element represented by the symbol.

[0072]

Heating temperature: 1000°C or higher and 1300°C or lower

In the case where the heating temperature is lower than 1000°C, a sufficient solution of carbides cannot occur, and the desired strength cannot be achieved. In the case where the heating temperature is higher than 1300°C, there is a decrease in the toughness of a base metal. Therefore, the heating temperature is set to be 1000°C or higher and 1300°C or lower.

[0073]

Rolling finish temperature: equal to or higher than the Ar_3 transformation temperature

In the case where the rolling finish temperature is lower than the Ar_3 transformation temperature, there is a decrease in ferrite transformation rate after rolling has been finished, and there is insufficient concentration of C into a non-transformed austenite phase when reheating is performed, which results in MA not being formed. Therefore, the rolling finish temperature is set to be equal to or higher than the Ar_3 transformation temperature.

[0074]

Cumulative rolling reduction ratio in a temperature range of 900°C or lower: 50% or more

This condition is one of the important manufacturing conditions in the present invention. A temperature range of 900°C or lower corresponds to the lower part of no-recrystallization temperature range in austenite. By controlling the cumulative rolling reduction ratio to be 50% or more in this temperature range, a decrease in austenite grain size can be achieved. With this method, there is an increase in the number of forming sites of MA to be formed at prior-austenite grain boundaries afterward, which contributes to the suppression of an increase in MA grain size.

[0075]

In the case where the cumulative rolling reduction ratio in a temperature range of 900°C or lower is less than 50%, since the equivalent circle diameter of formed MA becomes more than 3.0 μm, there is a case where there is a decrease in uniform elongation and/or there is a decrease in the toughness of a base metal. Therefore, the cumulative rolling reduction ratio in a temperature range of 900°C or lower is set to be 50% or more.

[0076]

Fig. 1 is a schematic diagram illustrating a cooling curve expressed in terms of the average temperature of a steel plate and a cooling-heating curve expressed in terms of the surface temperature of a steel plate in an accelerated cooling process according to the present invention.

A cooling start temperature is set to be equal to or higher than the Ar_3 transformation temperature in terms of the surface temperature of a steel plate.

After rolling has been finished, accelerated cooling is started at a temperature equal to or higher than the Ar_3 transformation temperature. In the case where the cooling start temperature is lower than the Ar_3 transformation temperature, a polygonal ferrite phase is formed, which results in a decrease in strength. Therefore, the cooling

start temperature is set to be equal to or higher than the Ar_3 transformation temperature.

[0077]

Regarding the cooling of the steel plate, the cooling rate is set to be 150°C/sec. or less and the cooling stop temperature is set to be 600°C or lower in terms of the surface temperature of the steel plate.

By performing cooling under conditions such that the cooling rate is 150°C/sec. or less and the cooling stop temperature is 600°C or lower in terms of the surface temperature of the steel plate, a bainite structure in the surface portions of the steel plate becomes soft. With this method, the maximum hardness of the surface portions of the steel plate can be controlled to be HV230 or less in terms of Vickers hardness, and hardness variations in the thickness direction and in the width direction can be suppressed to be ΔHV30 or less. By controlling hardness variations in the thickness and width directions to be ΔHV30 or less and by controlling the maximum hardness of the surface portions to be HV230 or less as described above, the elongation variation and a decrease in elongation of the steel plate can be suppressed. Thus, an elongation of 22% or more can be stably achieved. In addition, the reason why the yield ratio of a steel plate is set to be 85% or less and the elongation of the steel plate is set to be 22% or

more will be described hereafter. This is because it is necessary that the target value of the yield ratio be 85% or less and the target value of the elongation be 22% or more at the stage of the steel plate in order to achieve a yield ratio of 90% or less and an elongation of 20% or more at the stage of a steel pipe in consideration of changes in properties of the materials due to the working strain induced by forming a steel plate to a steel pipe.

[0078]

The cooling stop temperature is set to be 600°C or lower in terms of the surface temperature of the steel plate in order to control the surface temperature of the steel plate to be lower than or equal to a temperature at which bainite transformation starts, and it is preferable that the cooling stop temperature be 200°C or higher and 500°C or lower in terms of the surface temperature of the steel plate. Incidentally, since the upper limit of a cooling rate has an influence on the hardness of the surface of a steel plate, and since a polygonal ferrite phase is not formed so as to soften the surface portion of a steel plate in the present invention, the upper limit of the cooling rate is set to be 150°C/sec.

[0079]

A cooling rate in terms of the average temperature of the steel plate is set to be 15°C/sec. or more.

In the case where the cooling rate is less than 15°C/sec., a pearlite phase is formed when cooling is performed, and sufficient strength or sufficiently low yield ratio cannot be achieved. Therefore, the cooling rate in terms of the average temperature of the steel plate is set to be 15°C/sec. or more.

[0080]

In the present invention, the steel plate is supercooled down to a temperature range for bainite transformation by performing accelerated cooling, and it is possible to complete bainite transformation when reheating is subsequently performed without holding the steel plate at the reheating temperature.

[0081]

A cooling stop temperature in terms of the average temperature of the steel plate is set to be 450°C or higher and 650°C or lower.

This process is an important manufacturing condition in the present invention. In the present invention, a non-transformed austenite phase, which is present after reheating has been performed and in which C is concentrated, transforms into MA when air-cooling is performed after reheating has been performed.

That is, it is necessary that cooling be stopped in the middle of bainite transformation, that is, in a temperature

range in which a non-transformed austenite phase is present. In the case where the cooling stop temperature is lower than 450°C, bainite transformation is completed, and MA is not formed when air-cooling is performed and a decrease in yield ratio cannot be achieved. In the case where the cooling stop temperature is higher than 650°C, C is consumed by a pearlite phase which is precipitated when cooling is performed, and MA is not formed. Therefore, the accelerated cooling stop temperature is set to be 450°C or higher and 650°C or lower. The accelerated cooling stop temperature preferably be 500°C or higher and 600°C or lower from the viewpoint of achieving the area fraction of MA which is ideal for achieving better strength and toughness. Regarding this accelerated cooling, an arbitrary cooling apparatus can be used.

[0082]

Immediately after accelerated cooling has been stopped, reheating is performed up to a temperature of 550°C or higher and 750°C or lower at a heating rate of 1.0°C/sec. or more in terms of the surface temperature of the steel plate.

Here, "reheating is performed immediately after accelerated cooling has been stopped" means that reheating is performed at a heating rate of 1.0°C/sec. or more within 120 seconds after accelerated cooling has been stopped.

[0083]

This process is also an important manufacturing condition in the present invention. While a non-transformed austenite phase transforms into a bainite phase when reheating is performed after accelerated cooling has been performed as described above, C is evacuated into a non-transformed austenite phase which is remaining. Then, the non-transformed austenite phase in which C is concentrated is transformed into MA when air-cooling is performed after reheating has been performed.

In order to form MA, it is necessary that reheating be performed from a temperature higher than the Bf point to a temperature of 550°C or higher and 750°C or lower after accelerated cooling has been performed.

[0084]

In the case where the heating rate is less than 1.0°C/sec., it takes a long time for a steel plate to reach the target reheating temperature, which results in a decrease in productivity, and, there is a case where there is an increase in MA grain size. As a result, it is impossible to achieve a sufficiently low yield ratio, sufficient toughness or sufficient uniform elongation. Although the mechanisms are not necessarily clear, the reason is thought to be as follows. That is, by increasing the heating rate for reheating to be 1.0°C/sec. or more, an increase in size of a region in which C is concentrated is

suppressed, and an increase in grain size of MA which is formed in a cooling process after reheating has been performed is suppressed.

[0085]

In the case where the reheating temperature is lower than 550°C, transformation does not sufficiently progress, and a sufficient amount of C is not evacuated into a non-transformed austenite phase and a decrease in yield ratio cannot be achieved as a result of a sufficient amount of MA not being formed. In the case where the reheating temperature is higher than 750°C, sufficient strength cannot be achieved due to the softening of a bainite phase. Therefore, the reheating temperature is set to be 550°C or higher and 750°C or lower.

[0086]

In the present invention, after accelerated cooling has been performed, it is important to start reheating in a temperature range in which a non-transformed austenite phase is present. In this reheating process, in the case where a reheating start temperature is equal to or lower than the Bf point, bainite transformation is completed, and a non-transformed austenite phase disappears. Therefore, it is necessary that the reheating start temperature be higher than the Bf point.

[0087]

In order to ensure that C is concentrated in a non-transformed austenite phase, it is preferable that a steel plate be heated up to a temperature 50°C or more higher than the reheating start temperature. It is not particularly necessary that a holding time be set during which the steel plate is held at the reheating temperature.

[0088]

By using the manufacturing method according to the present invention, even in the case where cooling is performed immediately after reheating has been performed, a sufficient amount of MA is achieved, and thus a decrease in yield ratio and an increase in uniform elongation are achieved. However, in order to ensure that sufficient volume fraction of MA is achieved by promoting the diffusion of C more into a non-transformed austenite phase more, the steel plate may be held at the reheating temperature for 30 minutes or less.

[0089]

In the case where the temperature holding time is more than 30 minutes, there is a case where there is a decrease in strength due to the occurrence of recovery of a bainite phase. In addition, it is preferable to perform cooling fundamentally using an air-cooling method after reheating has been performed.

[0090]

As an apparatus for performing reheating after accelerated cooling has been performed, a heating apparatus may be equipped downstream of a cooling apparatus for performing accelerated cooling. Among heating apparatuses, it is preferable to use a gas-fired furnace or an induction heating apparatus with which a steel plate can be heated at a high heating rate.

[0091]

As described above, in the present invention, firstly, rolling is performed under conditions such that the cumulative rolling reduction ratio is 50% or more in a temperature range of 900°C or lower which is a no-recrystallization temperature range in austenite. With this method, there is an increase in the number of MA formation sites as a result of a decrease in an austenite grain size, and fine MA can be uniformly dispersed, which results in a low yield ratio of 85% or less in the state of a steel plate and of 90% or less in the state of a steel pipe being achieved.

[0092]

Moreover, in the present invention, by performing reheating at a high heating rate after accelerated cooling has been performed, an increase in an MA grain size is suppressed so that the equivalent circle diameter of MA is decreased to 3.0 μm or less. In addition, by performing

cooling at a cooling rate of 150°C/sec. or less in terms of the surface temperature of a steel plate down to a temperature of 600°C or lower in terms of the surface temperature of a steel plate, a bainite phase in the surface portion of the steel plate can be softened, which results in an elongation of 22% or more in the state of a steel plate and of 20% or more in the state of a steel pipe being achieved in a full-thickness tensile test using a test piece having a shape in accordance with GOST standards.

[0093]

With this method, even if a thermal history which causes deterioration of properties due to strain ageing in the case of a conventional steel is applied, MA is less likely to decompose in the case of the steel according to the present invention, and therefore, it is possible to maintain the specified metallographic structure consisting of a duplex-phase structure of a bainite phase and MA.

[0094]

As a result, in the present invention, even if a thermal history is applied at a temperature of 250°C for 30 minutes, which is classified into a thermal history at a high temperature and for a long duration among common coating processes for steel pipes, an increase in yield stress (YS) that is caused by strain aging and accompanied by an increase in yield ratio and a decrease in uniform

elongation can be suppressed. That is, the specified properties in the state of a steel plate and in the state of a steel pipe can be assured in the case of the steel according to the present invention even if a thermal history which causes deterioration of properties due to strain ageing in the case of a conventional steel plate is applied.

[0095]

Incidentally, in the case where a steel pipe is manufactured using the steel plate according to the present invention, the steel plate according to the present invention is formed into a cylindrical shape and the butted portions thereof are welded under conditions such that single-layer welding is performed on each of the outer and inner surfaces. Then, by performing a pipe-expanding treatment under a condition of a pipe expanding ratio of 0.4% or more and 2.0% or less, a steel pipe having good roundness can be obtained.

[EXAMPLES 1]

[0096]

The steels (steel types A through J) having the chemical compositions given in Table 1 were cast into slabs using a continuous casting method, and thick steel plates (Nos. 1 through 17) having a thickness of 20 mm, 28 mm or 33 mm were manufactured.

[0097]

[Table 1]

Steel Type	Chemical Composition (mass%)																	Ar_3 Transformation Point (°C)	Ti/N	Note
	C	Si	Mn	P	S	Al	Nb	Ti	Cu	Ni	Cr	Mo	V	Ca	B	N	O			
A	0.032	0.20	2.5	0.008	0.001	0.03	0.034	0.014	-	-	-	-	-	-	-	0.004	0.002	700	3.5	Example
B	0.051	0.56	1.8	0.008	0.002	0.04	0.023	0.011	0.24	0.20	-	-	-	-	-	0.005	0.001	734	2.2	Example
C	0.072	0.06	1.8	0.011	0.001	0.03	0.044	0.013	-	-	-	0.22	-	-	-	0.004	0.001	726	3.3	Example
D	0.064	0.15	1.7	0.008	0.001	0.03	0.021	0.009	0.20	0.20	-	0.18	-	0.0018	-	0.005	0.002	725	1.8	Example
E	0.054	0.15	2.2	0.008	0.001	0.04	0.025	0.008	-	-	0.10	-	-	-	-	0.005	0.002	716	1.6	Example
F	0.058	0.16	1.7	0.009	0.001	0.03	0.009	0.016	0.16	0.18	0.03	0.20	0.030	0.0016	-	0.006	0.002	725	2.7	Example
G	0.063	0.13	1.9	0.008	0.001	0.03	0.014	0.013	-	-	-	0.20	-	-	0.0010	0.004	0.002	722	3.3	Example
H	<u>0.023</u>	0.38	2.4	0.008	0.002	0.03	0.032	0.010	-	-	-	-	-	-	-	0.005	0.001	711	2.0	Comparative Example
I	0.062	0.65	<u>1.1</u>	0.009	0.001	0.03	0.024	0.011	-	-	-	0.10	-	-	0.0008	0.004	0.002	795	2.8	Comparative Example
J	<u>0.071</u>	0.34	2.2	0.008	0.001	0.03	0.035	0.014	-	-	-	-	-	-	-	0.004	0.002	734	3.5	Comparative Example

Annotation: <u>Underlined value</u> is out of the range according to the present invention.

Annotation: Ar_3 Transformation Point (°C) =910-310C-80Mn-20Cu-15Cr-55Ni-80Mo (An atomic symbol represents the content (mass%) of the chemical element represented by the symbol.)

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[0098]

Immediately after hot-rolling had been performed on the heated slabs, cooling was performed using a water-cooling type accelerated cooling apparatus, and then reheating was performed using an induction heating furnace or a gas-fired furnace. The induction heating furnace or a gas-fired furnace was equipped on the same line as an accelerated cooling apparatus.

[0099]

The manufacturing conditions of the steel plates (Nos. 1 through 17) are given in Table 2. Incidentally, temperature such as a heating temperature, a rolling finish temperature, a cooling stop (finish) temperature, and a reheating temperature were represented in terms of the average temperature of the steel plate. The average temperature was calculated from the surface temperature of the slab or the steel plate using parameters such as thickness and thermal conductivity.

In addition, a cooling rate is an average cooling rate which was derived by dividing a temperature difference necessary for cooling from a temperature after hot rolling has been finished down to a cooling stop (finish) temperature (450°C to 650°C) by a time spent for the cooling. In addition, reheating rate (heating rate) is an average heating rate which was derived by dividing a temperature

difference necessary for a reheating from a temperature after cooling had been finished up to a reheating temperature (550°C to 750°C) by a time spent for the reheating.

[0100]

[Table 2]

Table 2

No.	Steel Type	Thickness	Heating Temperature	Cumulative Rolling Reduction at 900°C or under	Rolling Finish Temperature	Cooling in Terms of Surface Temperature*1			Cooling in Terms of Average Temperature*2		Reheating Apparatus	Reheating Rate	Reheating Temperature	Note
						Cooling Start Temperature of Steel Plate Surface	Cooling Rate of Steel Plate Surface	Cooling Stop Temperature of Steel Plate Surface before Recuperation	Average Cooling Rate of Steel Plate	Average Cooling Stop Temperature of Steel Plate				
		(mm)	(°C)	(%)	(°C)	(°C)	(°C/sec.)	(°C)	(°C/sec.)	(°C)		(°C/sec.)	(°C)	
1	A	20	1130	65	780	720	130	390	30	590	Induction Heating Furnace	2	650	Example
2	B	20	1120	60	800	740	135	430	35	630	Induction Heating Furnace	3	650	Example
3	C	33	1080	70	780	730	120	410	20	610	Induction Heating Furnace	3	680	Example
4	D	20	1180	70	790	740	140	420	40	620	Induction Heating Furnace	5	650	Example
5	E	28	1050	60	790	730	135	340	35	540	Gas-fired Furnace	2	680	Example
6	F	33	1150	55	800	740	130	400	30	600	Induction Heating Furnace	3	660	Example
7	G	28	1150	75	780	730	135	370	35	570	Induction Heating Furnace	5	650	Example
8	E	20	970	75	790	730	135	410	35	610	Induction Heating Furnace	7	680	Comparative Example
9	E	20	1150	40	790	730	140	380	40	580	Induction Heating Furnace	5	650	Comparative Example
10	E	20	1180	75	790	730	130	400	3	600	Induction Heating Furnace	6	680	Comparative Example
11	F	28	1100	65	800	740	135	260	35	430	Induction Heating Furnace	5	650	Comparative Example
12	F	28	1200	60	800	740	135	410	35	610	Induction Heating Furnace	0.2	680	Comparative Example
13	F	28	1080	70	800	740	140	350	40	550	Induction Heating Furnace	7	530	Comparative Example
14	F	28	1080	70	760	720	230	200	40	550	Induction Heating Furnace	7	650	Comparative Example
15	H	20	1150	75	790	730	140	420	40	620	Induction Heating Furnace	6	650	Comparative Example
16	I	20	1090	70	840	800	140	310	40	510	Induction Heating Furnace	7	680	Comparative Example
17	J	28	1180	75	800	750	135	380	35	580	Induction Heating Furnace	2	650	Comparative Example

Annotation: Underlined value is out of the range according to the present invention.

*1: Cooling is controlled on the basis of a surface temperature of a steel plate.

*2: Cooling is controlled on the basis of an average temperature of a steel plate.

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[0101]

The mechanical properties of the steel plates manufactured as described above were determined. The results are given in Table 3. Tensile strength was evaluated on the basis of the average value thereof derived by collecting 2 test pieces for a full-thickness tensile test in a direction (C direction) at a right angle to the rolling direction and by performing a tensile test. The strength required in the present invention is a tensile strength of 590 MPa or more (API 5L X60 or more). A yield ratio and an elongation were evaluated on the basis of the average value thereof derived by collecting 2 test pieces for a full-thickness tensile test in a direction at a right angle to the rolling direction and by performing a tensile test. The deformation performances required in the present invention are a yield ratio of 85% or less and an elongation of 22% or more.

[0102]

The toughness of a base metal was evaluated, by collecting 3 full-size V notch Charpy test pieces in a direction at a right angle to the rolling direction, by performing a Charpy test, by determining absorbed energy at a temperature of -40°C and by calculating the average value of the 3 values. A case where the absorbed energy at a temperature of -40°C was 200 J or more was evaluated as good.

[0103]

Regarding toughness in a weld heat affected zone (HAZ), 3 test pieces that had been subjected to a thermal history corresponding to a heat input of 40 kJ/cm using a reproducing apparatus of weld thermal cycles were collected, and a Charpy impact test was performed for those test pieces. Then, absorbed energy at a temperature of -40°C was determined, and the average value for the 3 test pieces was calculated. A case where the Charpy absorbed energy at a temperature of -40°C was 100 J or more was evaluated as good.

[0104]

[Table 3]

No.	Steel Type	Thickness	MA Fraction in Steel Plate	MA Equivalent Circle Diameter in Steel Plate	C Direction in Plate State (Before Aging Treatment of 250°C×30min.) Tensile Strength	Yield Ratio	Uniform Elongation	Total Elongation	Hardness Variation in Thickness Direction	Hardness Variation in Width Direction	Surface Hardness	Base Metal Toughness	HAZ Toughness	Note
												vE-40°C	vE-40°C	
		(mm)	(%)	(μm)	(MPa)	(%)	(%)	(%)	ΔHV	ΔHV	HV	(J)	(J)	
1	A	20	11	1.6	621	75	12	24	20	25	222	307	141	Example
2	B	20	8	1.2	610	74	12	25	15	18	203	312	124	Example
3	C	33	13	2.6	677	71	11	23	24	23	214	294	118	Example
4	D	20	7	1.7	622	75	13	25	22	19	211	274	164	Example
5	E	28	6	1.6	655	73	11	26	14	16	220	318	155	Example
6	F	33	10	1.3	662	78	13	24	18	13	217	333	131	Example
7	G	26	4	1.5	636	70	12	25	11	14	216	361	182	Example
8	E	20	1	2.5	556	89	9	26	22	23	210	335	178	Comparative Example
9	E	20	7	3.5	608	77	12	27	25	24	222	129	124	Comparative Example
10	E	20	2	2.4	620	87	11	24	23	22	211	273	138	Comparative Example
11	F	28	0	1.5	655	94	9	21	26	34	232	285	161	Comparative Example
12	F	28	1	1.6	660	83	8	20	23	16	224	288	144	Comparative Example
13	F	28	0	1.3	633	89	9	20	25	25	215	312	116	Comparative Example
14	F	28	4	1.8	645	84	12	21	35	32	235	277	136	Comparative Example
15	H	20	1	1.4	655	88	8	25	16	18	207	293	122	Comparative Example
16	I	20	0	1.8	581	86	9	21	22	27	227	281	133	Comparative Example
17	J	28	14	4.3	643	66	12	21	34	33	233	302	28	Comparative Example

Annotation: Underlined value is out of the range according to the present invention.

Annotation: Tensile test was performed according to GOST (long gauge length).

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[0105]

In Table 3, in all cases of the examples of the present invention Nos. 1 through 7, the chemical compositions and the manufacturing conditions were in the range of the present invention, and each of these examples had a high strength of 590 MPa or more in terms of tensile strength, a low yield ratio of 85% or less, a high elongation of 22% or more and good toughness for a base metal and a weld heat affected zone.

In addition, the microstructure in the central portion of the steel plate included mainly a bainite phase in which MA is dispersed, in which the area fraction of MA was 3% or more and 15% or less, and in which the equivalent circle diameter of MA was 3.0 μm or less. Here, the area fraction of MA was determined by performing image processing on a microstructure observed using a scanning electron microscope (SEM).

[0106]

On the other hand, in the case of the comparative examples Nos. 8 through 14, the chemical compositions were in the range of the present invention, but the manufacturing methods were out of the range of the present invention. As a result, the microstructures were out of the range of the present invention, and the yield ratio or the elongation was insufficient or the sufficient strength or toughness was not

obtained in both conditions of before and after a strain ageing treatment at a temperature of 250°C for 30 minutes. In the case of Nos. 15 through 17, since the chemical compositions were out of the range of the present invention, the yield ratio and elongation of No. 15 were out of the range of the present invention, and the tensile strength, elongation, and yield ratio of No. 16 were all out of the range of the present invention. The toughness in a weld heat affected zone (HAZ) of No. 17 was out of the range of the present invention.

[0107]

Subsequently, UOE steel pipes were manufactured using the steel plates (Nos. 1 through 17) that had been manufactured under conditions given in Table 2.

After the obtained steel plates were formed by performing U-press and O-press, using a submerged arc welding method, inner surface seam welding was performed, and then outer surface seam welding was performed. Subsequently, by performing pipe-expanding treatment under a condition of a pipe-expanding ratio of 0.6% to 1.2%, steel pipes having an outer diameter of 400 to 1626 mm were manufactured. Using a tensile test piece in accordance with GOST standards cut out of the base metal of the steel pipe, tensile properties were evaluated. In addition, using a tensile test piece in accordance with GOST standards cut out

of a sample material by the same method described above, which had been cut out of the base metal of the pipe in order to form the test pieces and subjected to an ageing treatment at a temperature of 250°C for 30 minutes, tensile properties after having undergone ageing treatment were evaluated. In addition, using a V-notch Charpy impact test piece according to JIS Z 2202 (1980) cut out of the central portion in the thickness direction of the base metal of the steel pipe, a Charpy impact test was performed at a test temperature of -40°C. Moreover, using a test piece for a DWTT (Drop Weight Tear Test) according to API-5L cut out of the steel pipe, a DWTT was performed at a test temperature of -20°C in order to determine an SA value (Shear Area: ductile fracture area ratio). In addition, using a V-notch Charpy impact test piece according to JIS Z 2202 (1980) cut out of the portion of the outer surface FL (Fusion Line) of the welded joint of the steel pipe, a Charpy impact test was performed at a test temperature of -40°C. Here, the notch was formed at a position where the HAZ and the weld metal were included at a ratio of 1:1.

[0108]

The test results are given in Table 4.

[0109]

[Table 4]

No.	Steel Type	Thickness	MA Fraction in Steel Plate	MA Equivalent Circle Diameter in Steel Plate	C Direction in UOE Pipe State (Before Aging Treatment of 250°C×30min.)								C Direction in Ageing Treated Pipe State (After Aging Treatment of 250°C×30min.)								L Direction in Ageing Treated Pipe State (After Aging Treatment of 250°C×30min.)					Note
					Yield Strength	Tensile Strength	Yield Ratio	Uniform Elongation	Total Elongation	Base Metal Toughness		HAZ Toughness	Yield Strength	Tensile Strength	Yield Ratio	Uniform Elongation	Total Elongation	Base Metal Toughness		HAZ Toughness	Yield Strength	Tensile Strength	Yield Ratio	Uniform Elongation	Total Elongation	
										vE-40°C	DWTT SA-20°C	vE-40°C						vE-40°C	DWTT SA-20°C	vE-40°C						
		(mm)	(%)	(μm)	(MPa)	(MPa)	(%)	(%)	(%)	(J)	(%)	(J)	(MPa)	(MPa)	(%)	(%)	(%)	(J)	(%)	(J)	(MPa)	(MPa)	(%)	(%)	(%)	
1	A	18	11	1.6	532	626	85	10	22	307	100	141	544	632	86	9	22	321	100	132	542	630	86	10	22	Example
2	B	21	8	1.2	517	616	84	10	23	312	100	124	531	625	85	10	22	304	100	133	530	623	85	11	21	
3	C	29	13	2.6	547	675	81	9	21	294	100	118	575	684	84	10	21	288	100	122	571	680	84	10	22	
4	D	21	7	1.7	534	628	85	11	23	274	100	164	532	633	84	11	23	268	100	141	529	630	84	10	23	
5	E	28	6	1.6	547	659	83	9	24	318	100	155	557	671	83	9	23	307	100	146	549	661	83	9	24	
6	F	30	10	1.3	559	665	84	12	22	333	100	131	584	679	86	10	21	311	100	120	580	674	86	11	22	
7	G	26	4	1.5	514	642	80	10	23	361	100	182	534	651	82	10	22	341	100	152	533	650	82	11	22	
8	E	20	1	2.5	514	559	92	7	24	335	100	178	524	570	92	7	23	311	100	141	521	566	92	7	22	Comparative Example
9	E	20	7	3.5	509	613	83	10	25	129	75	124	543	617	88	10	24	134	80	102	540	614	88	9	24	
10	E	20	2	2.4	571	628	91	9	22	273	100	138	538	626	86	8	21	266	100	108	536	623	86	8	22	
11	F	28	0	1.5	617	663	93	7	19	285	100	161	620	674	92	5	19	277	100	114	617	671	92	6	20	
12	F	28	1	1.6	559	666	84	6	18	288	100	144	569	677	84	6	17	269	100	138	567	675	84	7	18	
13	F	28	0	1.3	581	639	91	7	18	312	100	116	588	646	91	6	18	274	100	104	586	644	91	7	19	
14	F	28	5	1.8	554	652	85	11	19	277	100	136	572	657	87	8	18	266	100	122	570	655	87	8	18	
15	H	21	1	1.4	599	661	91	6	23	293	100	122	621	675	92	7	22	288	100	133	626	673	93	7	23	
16	I	20	0	1.8	534	580	92	7	19	281	95	133	535	588	91	7	18	278	100	103	533	586	91	7	19	
17	J	28	14	4.3	518	648	80	10	19	302	100	28	544	664	82	9	19	245	100	19	542	661	82	9	19	

Annotation: Underlined value is out of the range according to the present invention.

Annotation: Tensile test was performed according to GOST (long gauge length).

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[0110]

The target ranges of the present invention regarding the base metal of a pipe are: a tensile strength of 590 MPa or more, an elongation of 20% or more, and a ratio of a 0.5%-proof stress to a tensile strength of 90% or less, where all of those were determined before and after an ageing treatment at a temperature of 250°C for 30 minutes. At the same time, the target ranges of the present invention regarding the base metal are a Charpy absorbed energy at a test temperature of -40°C of 200 J or more and a DWTTSA-20°C (ductile fracture area ratio in a DWTT test at a test temperature of -20°C) of 85% or more, and the target range of the present invention regarding the seam weld joint of steel pipe is a Charpy absorbed energy of 100 J or more at an outer surface FL notch at a test temperature of -40°C.

[0111]

In Table 4, in all cases of the examples of the present invention Nos. 1 through 7, the chemical compositions and the manufacturing methods were in the ranges of the present invention. Therefore, before and after a strain ageing treatment at a temperature of 250°C for 30 minutes, each of these examples had a high tensile strength of 590 MPa or more, a low yield ratio of 90% or less and a high elongation of 20% or more and, further, the toughness of a base metal and in a weld heat affected zone thereof were good.

[0112]

On the other hand, in the case of the comparative examples Nos. 8 through 14, the chemical compositions were in the range of the present invention, but the manufacturing methods were out of the range of the present invention. As a result, the microstructures were out of the range of the present invention, and the yield ratio or elongation was insufficient or the sufficient strength or toughness was not obtained in both conditions of before and after a strain ageing treatment at a temperature of 250°C for 30 minutes. In the case of Nos. 15 through 17, since the chemical compositions were out of the range of the present invention, the yield ratio and uniform elongation of No. 15 were out of the range of the present invention, and the tensile strength, yield ratio, uniform elongation, and elongation of No. 16 were all out of the range of the present invention. The elongation and toughness in a weld heat affected zone of No. 17 were out of the range of the present invention.

CLAIMS

[Claim 1]

A high strength steel plate having a low yield ratio, the steel plate having a chemical composition containing, by mass%, C: 0.03% or more and 0.08% or less, Si: 0.01% or more and 1.0% or less, Mn: 1.2% or more and 3.0% or less, P: 0.015% or less, S: 0.005% or less, Al: 0.08% or less, Nb: 0.005% or more and 0.07% or less, Ti: 0.005% or more and 0.025% or less, N: 0.010% or less, O: 0.005% or less and the balance being Fe and inevitable impurities, a metallographic structure being a dual-phase structure consisting of a bainite phase and island martensite, wherein the area fraction of the island martensite is 3% or more and 15% or less, wherein the equivalent circle diameter of the island martensite is 3.0 μm or less, and wherein the remainder of the metallographic structure is a bainite phase, a hardness variation in the thickness direction of ΔHV30 or less in terms of Vickers hardness, a hardness variation in the width direction of ΔHV30 or less in terms of Vickers hardness, a maximum hardness in the surface portions of the steel plate of HV230 or less in terms of Vickers hardness and a yield ratio of 85% or less and an elongation of 22% or more in a full-thickness tensile test using a test piece having a shape in accordance with GOST standards.

[Claim 2]

The high strength steel plate having a low yield ratio according to Claim 1, the steel plate having the chemical composition further containing, by mass%, one or more selected from among Cu: 0.5% or less, Ni: 1% or less, Cr: 0.5% or less, Mo: 0.5% or less, V: 0.1% or less, Ca: 0.0005% or more and 0.003% or less and B: 0.005% or less.

[Claim 3]

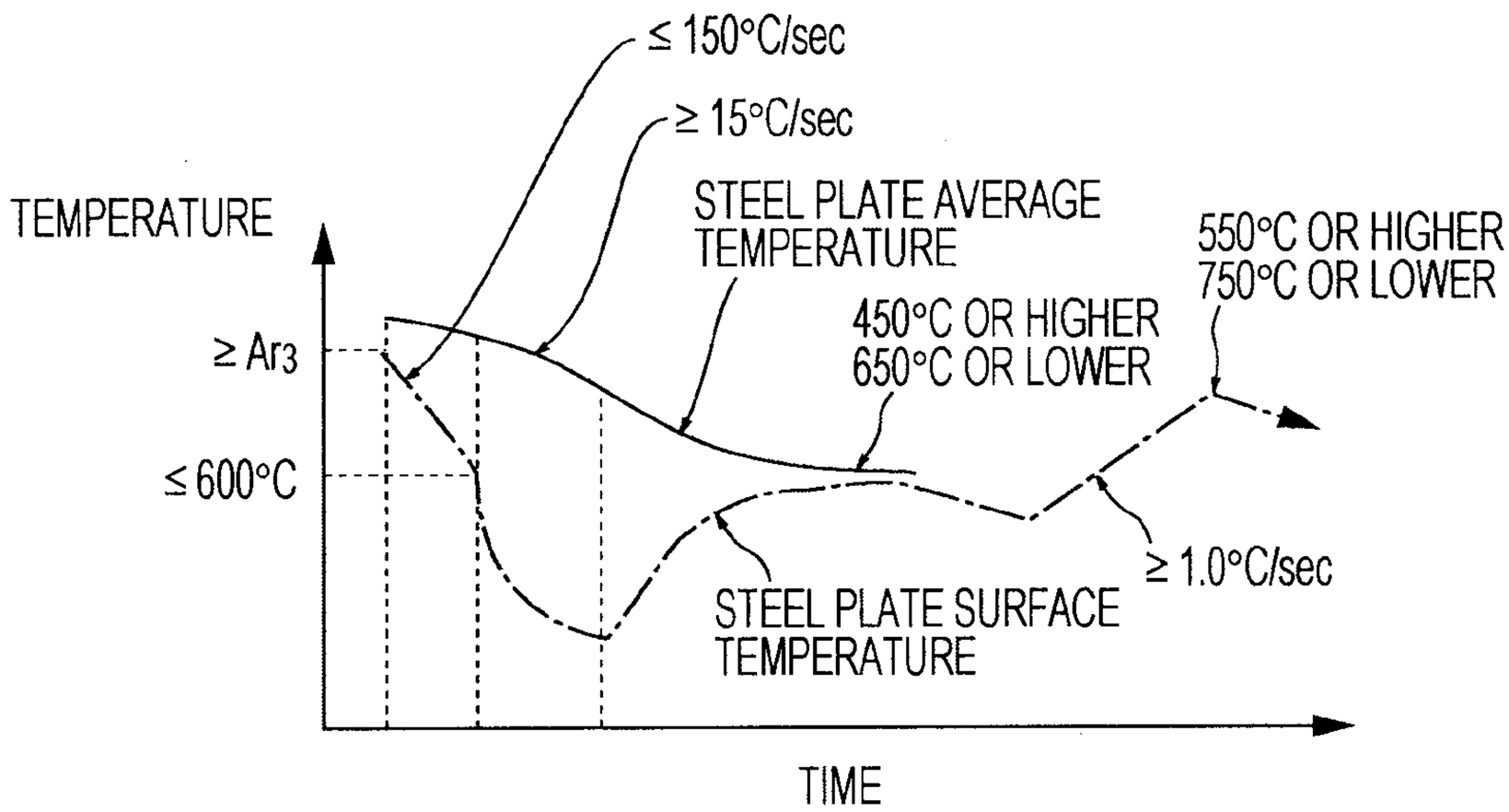
A method for manufacturing a high strength steel plate having a low yield ratio, the method comprising heating steel having the chemical composition according to Claim 1 or 2 at a temperature of 1000°C or higher and 1300°C or lower, performing hot rolling under conditions such that the cumulative rolling reduction ratio is 50% or more in a temperature range of 900°C or lower, finishing hot rolling at a temperature equal to or higher than the Ar_3 transformation point, starting cooling at a temperature equal to or higher than the Ar_3 transformation point, performing cooling at a cooling rate of 150°C/sec. or less in terms of the surface temperature of the steel plate until the surface temperature becomes 600°C or lower, performing cooling at a cooling rate of 15°C/sec. or more in terms of the average temperature of the steel plate until the average temperature becomes 450°C or higher and 650°C or lower, and, immediately after the cooling has been performed, performing reheating at a heating rate of 1.0°C/sec. or more in terms

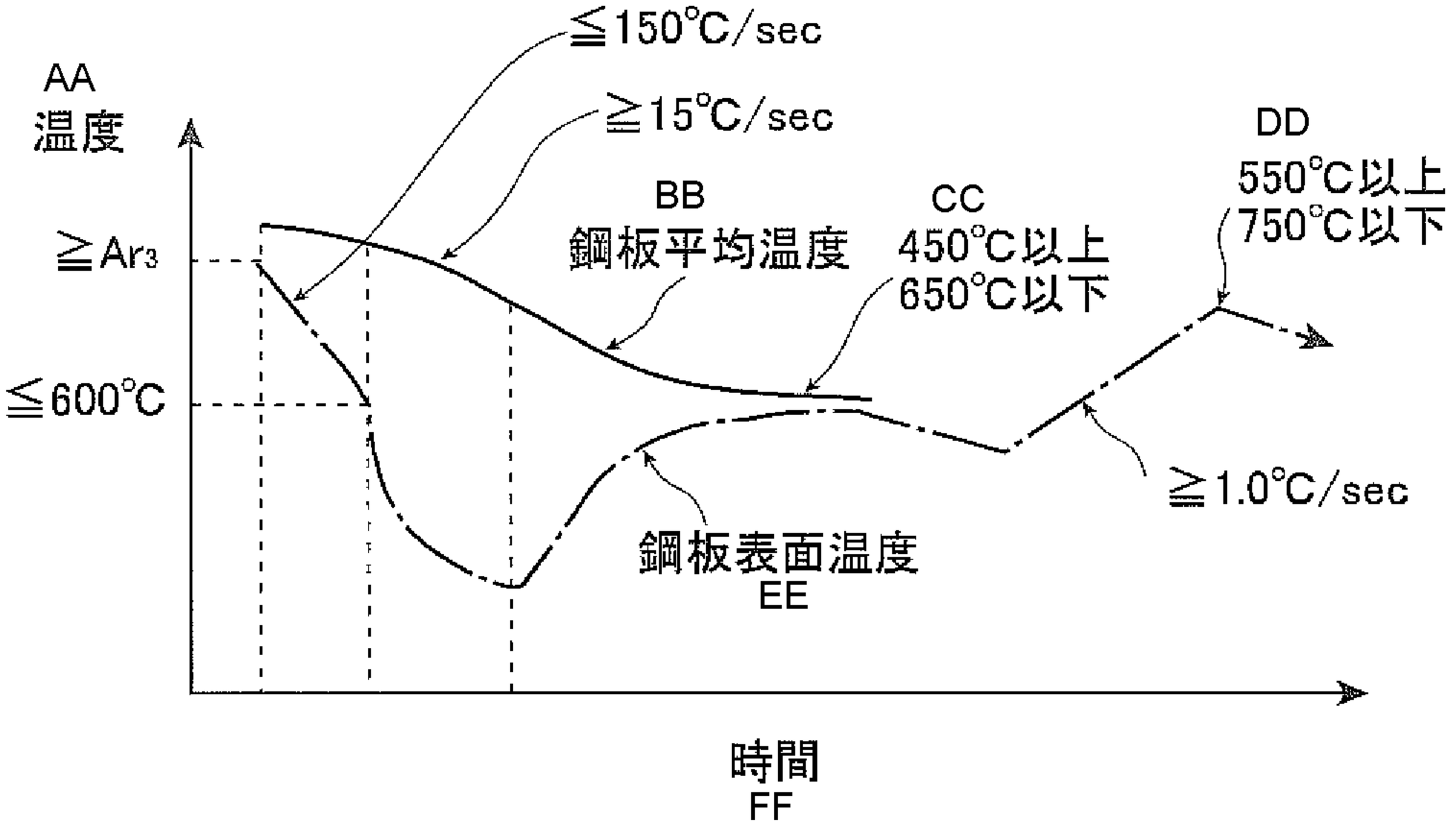
of the surface temperature of the steel plate until the surface temperature becomes 550°C or higher and 750°C or lower.

[Claim 4]

A high strength welded steel pipe, the steel pipe being manufactured by forming the steel plate according to Claim 1 or 2 into a cylindrical shape, by welding butted portions of the shaped cylinder under conditions such that single-layer welding is performed on each of the outer and inner surfaces, and by thereafter performing a pipe-expanding treatment under a condition of a pipe expanding ratio of 0.4% or more and 2.0% or less, and having a yield ratio of 90% or less and an elongation of 20% or more in a full-thickness tensile test using a test piece having a shape in accordance with GOST standards, and, further having a yield ratio of 90% or less and an elongation of 20% or more even after a strain ageing treatment has been performed under conditions such that the temperature is 250°C or lower and the duration is 30 minutes or less.

FIG. 1


≤ 150°C/sec
≥ 15°C/sec
TEMPERATURE
STEEL PLATE AVERAGE TEMPERATURE
550°C OR HIGHER
750°C OR LOWER
450°C OR HIGHER
650°C OR LOWER
≥ Ar3
≤ 600°C
≥ 1.0°C/sec
STEEL PLATE SURFACE TEMPERATURE
TIME



AA Temperature
BB Average temperature of steel plate
CC 450° C to 650° C
DD 550° C to 750° C
EE Surface temperature of steel plate
FF Time