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(72) Inventors:
• **KOJIMA, Akihiko**
Tokyo 100-8071 (JP)
• **TANAKA, Yoichi**
Tokyo 100-8071 (JP)
• **SHIRAHATA, Hiroyuki**
Tokyo 100-8071 (JP)
• **NAKASHIMA, Kiyotaka**
Tokyo 100-8071 (JP)
• **NAGAI, Yoshihide**
Tokyo 100-8701 (JP)

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(74) Representative: **Vossius & Partner**
Siebertstrasse 4
81675 München (DE)

(71) Applicant: **Nippon Steel Corporation**
Chiyoda-ku
Tokyo
100-8071 (JP)

(54) **PROCESS FOR PRODUCING THICK HIGH-STRENGTH STEEL PLATE EXCELLENT IN BRITTLE FRACTURE ARRESTABILITY AND TOUGHNESS OF ZONE AFFECTED BY HEAT IN LARGE-HEAT-INPUT WELDING AND THICK HIGH-STRENGTH STEEL PLATE EXCELLENT IN BRITTLE FRACTURE ARRESTABILITY AND TOUGHNESS OF ZONE AFFECTED BY HEAT IN LARGE-HEAT-INPUT WELDING**

(57) This method for producing a thicker high-strength steel plate that includes: cooling a continuously cast slab containing, in terms of mass %, C: 0.05 to 0.12%, Si: not more than 0.3%, Mn: 1 to 2%, P: not more than 0.015%, S: not more than 0.005%, B: 0.0003 to 0.003%, V: 0.01 to 0.15%, Al: 0.001 to 0.1%, Ti: 0.005 to 0.02%, N: 0.002 to 0.01%, and O: not more than 0.004%, with the remainder being iron and unavoidable impurities, to a temperature of Ar₃-200°C or lower, and subsequently reheating the slab to 950 to 1,100°C; subjecting the continuously cast slab to rough rolling at 900°C or higher with a cumulative reduction ratio of at least 30%; subsequently performing finish rolling at 700°C or higher

with a cumulative reduction ratio of at least 50% under conditions that both of the finish rolling start temperature and the finish rolling completion temperature are not higher than the temperature represented by the formula: $\{-0.5 \times (\text{slab heating temperature (°C)}) + 1.325\}$ (°C), thereby forming a rolled plate; and then cooling the rolled plate to 500°C or lower by accelerated cooling to obtain a steel plate. In the continuously cast slab, a calculated value of an amount of B {effective B amount: Bef(%)} which is solid-solubilized into an austenite base material prior to transformation is not more than 0%, and a carbon equivalent Ceq satisfies a range from 0.32 to 0.42%.

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Description

TECHNICAL FIELD

[0001] The present invention relates to a method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone (hereafter also abbreviated as HAZ) in high heat-input welding, as well as a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding. The thicker high-strength steel plate according to the present invention is used mainly for the construction of ships such as large container ships, but may also be used for other welded structures such as buildings, bridges, tanks, and marine structures. The present application claims priority on Japanese Patent Application No. 2007-315840, filed on December 6, 2007, the content of which is incorporated herein by reference.

BACKGROUND ART

[0002] Examples of the current needs for welded structures typified by boats and ships include increased size for the structures, a high level of safety in terms of cracking, improved welding efficiency during construction, and favorable economic viability of the steel plate used as the structural material. In response to these trends, the demands placed on the steel plate used in the welded structures, including (1) a high degree of strength at large plate thickness values, (2) favorable brittle fracture arrestability, (3) favorable HAZ toughness in high heat-input welding, and (4) low production costs continue to become more stringent. As a result, large container ships are now produced using hull construction steel plate having a yield strength in the order of 390 MPa (a tensile strength in the order of 510 MPa) or a yield strength in the order of 460 MPa (a tensile strength in the order of 570 MPa).

Specifically, as disclosed in Non-Patent Document 1 and the like, the steel plate used in large ships such as large container ships are required to simultaneously satisfy: (1) a yield strength in the order of 390 to 460 MPa (equivalent to a tensile strength in the order of 510 to 570 MPa) for a thick steel plate having a plate thickness of 50 to 80 mm (hereafter referred to as a "thick material"), (2) a temperature $T_{kca=6000}$ (hereafter referred to as the "arrestability indicator $T_{kca=6000}$ ") at which the brittle fracture arrestability Kca reaches 6,000 N/mm^{1.5} that satisfies $T_{kca=6000} \leq -10^{\circ}\text{C}$, (3) an HAZ toughness (Charpy impact absorbed energy) vE (-20°C) of a weld formed by welding at a heat input of at least 20 kJ/mm that satisfies $vE (-20^{\circ}\text{C}) \geq 47 \text{ J}$, and (4) a reduction in the amount of expensive alloy elements (such as a Ni amount of $\leq 1\%$).

[0003] Patent Document 1 represents one example of a technique relating to a thicker high-strength steel plate for use within ships, and this Patent Document 1 discloses a technique for producing a steel plate that has a plate thickness of 50 to 80 mm, and is able to partially satisfy the above requirements (1), (3) and (4). However, the thicker high-strength steel plate disclosed in Patent Document 1 is unable to satisfy the above requirement (2), as is evident from the examples disclosed in the document.

[0004] Further, Non-Patent Document 2 discloses a thick steel plate having a plate thickness of 65 mm, which although having a satisfactorily high Charpy impact absorbed energy of vE (-40°C) = 170 J for a small test piece, exhibits an unsatisfactory brittle fracture arrestability of $T_{kca=6000} = 18^{\circ}\text{C}$ that is confirmed in a large-scale crack test (see FIG. 7 of the Patent Document 2). These findings indicate that for a thick steel plate, by using the Charpy impact absorbed energy vE (-40°C) for a small test piece, it is not possible to guarantee that the brittle fracture arrestability obtained in a large-scale crack test satisfies $T_{kca=6000} \leq -10^{\circ}\text{C}$. In other words, it is not possible for conventional techniques to determine the brittle fracture arrestability required for a thicker high-strength steel plate for use within large ships by correlation with the Charpy impact properties measured using a small test piece, and an accurate evaluation of the brittle fracture arrestability cannot be made without conducting a large-scale crack test of a full thickness test piece, typified by the ESSO test (compliant with WES 3003).

[0005] Conventionally, it has been considered that the brittle fracture arrestability is dependent on the plate thickness, and the crack arrestability deteriorates as the plate thickness increases. However, for thick materials having a thickness of 50 mm or greater that represent the target for the present invention, experimental data for this plate thickness effect is limited, and the degree to which the brittle fracture arrestability deteriorates as a result of the increased thickness is unclear.

[0006] However, in a thick steel plate produced by a TMCP (Thermo Mechanical Control Process), the addition of boron (B) has conventionally been used to increase the strength. One example of the effect of this B addition is that solid solution B, which is segregated at the austenite (γ) grain boundaries during the accelerated cooling conducted after rolling, enhances the hardenability upon transformation. In Patent Document 1, increased strength is targeted by the combined addition of Nb and B. As disclosed in the examples of Patent Document 1, this technique is characterized in that the rolling finishing temperature is in a range from 930 to 1,000°C which is high, and the strength is increased by the combined addition of Nb and C to achieve superior hardenability, with the essential condition that accelerated cooling must be conducted from recrystallized austenite (recrystallized γ). In contrast, the Patent Document 1 also discloses

that when a low-temperature rolling is conducted with a rolling finishing temperature of lower than 930°C that is in the non-recrystallization region, although the toughness is satisfactory, the resulting strength properties are unsatisfactory, and it is difficult to realize the increasing of the strength due to a Nb-B combined effect.

[0007] Furthermore, Patent Document 1 discloses a technique for using B in a high heat-input HAZ, and describes the effectiveness of combining a grain boundary ferrite inhibiting effect (a hardenability improving effect) due to solid solution B within the γ and an intra-granular ferrite promoting effect (a hardenability reducing effect) due to BN within the γ , at a Ceq value of 0.30 to 0.38%. In other words, B has two mutually opposing roles in relation to the hardenability. To summarize the technique for using B disclosed in Patent Document 1, the hardenability improvement effect provided by the solid solution B within the γ is utilized in directly quenched base materials and the high heat-input HAZ, while at the same time, the hardenability reducing effect provided by the precipitated B (BN in this case) within the γ is utilized in the high heat-input HAZ.

[0008] Furthermore, the inventors of the present invention have previously completed inventions in which, in order to enhance the toughness of a high heat-input HAZ, the VN that precipitates within the γ during the HAZ cooling step is subjected to a co-precipitation with a pinning particle (an oxide or sulfide), and these VN composite particles act as ferrite transformation nuclei, thereby reducing the size of the HAZ microstructure. These inventions are disclosed in Patent Documents 2 and 3. Further, as disclosed in Non-Patent Document 3, it is well known that the addition of V provides an effect that improves the strength of the base material.

As described above, the addition of B or V is known to yield an improvement in the base material strength and an improvement in the toughness of a high heat-input HAZ.

[0009] Ni is generally known as a rare element capable of improving the toughness of the base materials or HAZ, and the effective use of Ni is typically considered in terms of requirements (2) and (3) above. However, Ni is an extremely expensive element, and the price of Ni has increased dramatically in recent years. Further, steels containing added Ni tend to be prone to developing surface blemishes; therefore, steps must be taken to avoid such blemishes. Accordingly, Ni addition produces conflicting interests between satisfying the above requirement (4) and satisfying the requirements (2) and (3). Furthermore, from the viewpoint of the above requirement (1), as increasing the amount of added alloy elements, the carbon equivalent (Ceq) increases; thereby, the HAZ becomes harder and more brittle during high heat-input welding, and therefore another conflict of interest develops between the requirement (1) and the requirement (3). Moreover, from the viewpoint of the requirement (2), if refinement of the pre-transformation γ structures in a TMCP is pursued, then the hardenability deteriorates and the strength decreases; thereby, another conflict of interest exists between the requirement (1) and the requirement (2).

For these reasons, development of a steel plate that is able to simultaneously satisfy the above requirements (1) to (4), which tend to represent mutually opposing interests, has been keenly sought.

Patent Document 1: Japanese Patent No. 3,599,556

Patent Document 2: Japanese Unexamined Patent Application, First Publication No. 2005-298900

Patent Document 3: Japanese Unexamined Patent Application, First Publication No. 2007-262508

Non-Patent Document 1: "Guidelines on the Application of YP47 Steel for Hull Structures of Large Container Carriers", published by Nippon Kaiji Kyokai (ClassNK) (October 2008)

Non-Patent Document 2: Journal of The Japan Society of Naval Architects and Ocean Engineers, 2006A-G4-10

Non-Patent Document 3: CAMP-ISIJ, 6 (1993), p. 684

DISCLOSURE OF INVENTION

PROBLEMS TO BE SOLVED BY THE INVENTION

[0010] The present invention takes the above circumstances into consideration, and aims to provide a method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding, that is capable of realizing (1) high strength for thick steel plate including a yield strength in the order of 390 to 460 MPa and a tensile strength in the order of 510 to 570 MPa for a plate thickness of 50 to 80 mm, (2) favorable brittle fracture arrestability indicated by an arrestability indicator $T_{kca=6000} \leq -10^\circ\text{C}$, (3) favorable HAZ toughness in high heat-input welding indicated by $vE (-20^\circ\text{C}) \geq 47 \text{ J}$ even when the heat input during welding is $\geq 20 \text{ kJ/mm}$, and (4) reduced production costs due to a reduction in the amount of expensive alloy elements (such as a Ni amount of $\leq 1\%$), as well as to provide a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding.

MEANS TO SOLVE THE PROBLEMS

[0011] In order to achieve the above object, the present invention adopts the aspects described below.

A method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention includes: cooling a continuously cast slab containing, in mass % values, C: 0.05 to 0.12%, Si: not more than 0.3%, Mn: 1 to 2%, P: not more than 0.015%, S: not more than 0.005%, B: 0.0003 to 0.003%, V: 0.01 to 0.15%, Al: 0.001 to 0.1%, Ti: 0.005 to 0.02%, N: 0.002 to 0.01%, and O: not more than 0.004%, with the remainder being iron and unavoidable impurities, to a temperature of Ar_3 -200°C or lower, and subsequently reheating the slab to 950 to 1,100°C; subjecting the continuously cast slab to rough rolling at 900°C or higher with a cumulative reduction ratio of at least 30%; subsequently performing finish rolling at 700°C or higher with a cumulative reduction ratio of at least 50% under conditions that both of the finish rolling start temperature and the finish rolling completion temperature are not higher than a temperature represented by a formula: $\{-0.5 \times (\text{slab heating temperature (°C)}) + 1,325\}$ (°C), thereby forming a rolled plate; and cooling the rolled plate to 500°C or lower by accelerated cooling to obtain a steel plate. In the above continuously cast slab, a calculated value of an amount of B {effective B amount: B_{ef} (%)} which is solid-solubilized into an austenite base material prior to transformation is not more than 0%, and a carbon equivalent C_{eq} satisfies a range from 0.32 to 0.42%.

If the amount of residual oxygen O_{Ti} (%) that remains after deoxidation by strong deoxidizing elements and is able to undergo deoxidation by Ti that is a weak deoxidizing element is an amount represented by formula (1) shown below, then the effective B amount B_{ef} (%) is represented by formula (2) shown below. Further, the carbon equivalent C_{eq} (%) is represented by formula (3) shown below, and the Ar_3 is represented by formula (4) shown below.

$$O_{Ti} (\%) = O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al \quad \cdots (1)$$

{in formula (1), component elements that represent unavoidable impurities are also included within the calculation.}

$$B_{ef} (\%) = B - 0.77\{N - 0.29(Ti - 2O_{Ti})\} \quad \cdots (2)$$

{in formula (2), when $O_{Ti} \leq 0$, it is deemed that $O_{Ti} = 0$, when $O_{Ti} > 0$, it is deemed that $Ti - 2O_{Ti} \geq 0.005$ (%), and when $N - 0.29(Ti - 2O_{Ti}) \leq 0$ (although when $O_{Ti} \leq 0$, $O_{Ti} = 0$), it is deemed that $N - 0.29(Ti - 2O_{Ti}) =$

$$C_{eq} (\%) = C + Mn/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15 \quad \cdots (3)$$

$$Ar_3 (\text{°C}) = (910 - 310C - 80Mn - 20Cu - 55Ni - 80Mo) \quad \cdots (4)$$

In the method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention, the method may further include conducting a tempering heat treatment at a temperature within a range from 350 to 700°C for 5 to 60 minutes after the accelerated cooling mentioned above.

In the continuously cast slab, the S content may be within a range from 0.0005 to 0.005%, and the O content may be within a range from 0.001 to 0.004%, and the continuously cast slab may further contain, in mass % values, either or both of Ca: 0.0003 to 0.004% and Mg: 0.0003 to 0.004%.

The continuously cast slab may further contain, in mass % values, one or more selected from the group consisting of Ni: 0.01 to 1%, Cu: 0.01 to 1%, Cr: 0.01 to 1%, Mo: 0.01 to 0.5%, and Nb: 0.003 to 0.03%.

The continuously cast slab may further contain, in mass % values, either or both of REM: 0.0003 to 0.02% and Zr: 0.0003 to 0.02%.

[0012] A thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention contains, in mass % values, C: 0.05 to 0.12%, Si: not more than 0.3%, Mn: 1 to 2%, P: not more than 0.015%, S: not more than 0.005%, B: 0.0003 to 0.003%, V: 0.01 to 0.15%, Al: 0.001 to 0.1%, Ti: 0.005 to 0.02%, N: 0.002 to 0.01%, and O: not more than 0.004%, with the remainder being iron and unavoidable impurities, wherein if an amount of residual oxygen that remains after deoxidation by strong deoxidizing elements and is able to undergo deoxidation by Ti that is a weak deoxidizing element is an amount represented by formula (5) shown below, then a calculated value of an amount of B {effective B amount: B_{ef}(%)} which is solid-solubilized into an austenite base material prior to transformation is not more than 0%, a carbon equivalent C_{eq}

represented by formula (7) shown below satisfies a range from 0.32 to 0.42%, a plate thickness is within a range from 50 to 80 mm, a yield strength is in the order of 390 to 460 MPa, a temperature $T_{kca=6000}$ at which the brittle fracture arrestability Kca reaches 6,000 N/mm^{1.5} is -10°C or lower, and a Charpy impact absorbed energy vE (-20°C), which is an indicator of high heat-input HAZ toughness with 20 kJ/mm or greater of heat input, is at least 47 J.

$$O_{Ti} (\%) = O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al \quad \cdots (5)$$

{in formula (5), component elements that represent unavoidable impurities are also included within the calculation.}

$$Bef (\%) = B - 0.77\{N - 0.29(Ti - 2O_{Ti})\} \quad \cdots (6)$$

{in formula (6), when $O_{Ti} \leq 0$, it is deemed that $O_{Ti} = 0$, when $O_{Ti} > 0$, it is deemed that $Ti - 2O_{Ti} \geq 0.005 (\%)$, and when $N - 0.29(Ti - 2O_{Ti}) \leq 0$ (although when $O_{Ti} \leq 0$, $O_{Ti} = 0$), it is deemed that $N - 0.29(Ti - 2O_{Ti}) = 0$.}

$$Ceq (\%) = C + Mn/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15 \quad \cdots (7)$$

In the thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention, the S content may be within a range from 0.0005 to 0.005%, and the O content may be within a range from 0.001 to 0.004%, and the steel plate may further contain, in mass % values, either or both of Ca: 0.0003 to 0.004% and Mg: 0.0003 to 0.004%.

The steel plate may further contain, in mass % values, one or more selected from the group consisting of Ni: 0.01 to 1%, Cu: 0.01 to 1%, Cr: 0.01 to 1%, Mo: 0.01 to 0.5%, and Nb: 0.003 to 0.03%.

The steel plate may further contain, in mass % values, either or both of REM: 0.0003 to 0.02% and Zr: 0.0003 to 0.02%.

EFFECTS OF THE INVENTION

[0013] The method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding, and the thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding of the present invention are capable of realizing (1) high strength for a thick steel plate including a yield strength in the order of 390 to 460 MPa (namely, a tensile strength in the order of 510 to 570 MPa) for a plate thickness of 50 to 80 mm, (2) favorable brittle fracture arrestability indicated by an arrestability indicator $T_{kca=6000} \leq -10^\circ\text{C}$, (3) favorable HAZ toughness in high heat-input welding indicated by $vE (-20^\circ\text{C}) \geq 47 \text{ J}$ even when the heat input during welding is $\geq 20 \text{ kJ/mm}$, and (4) reduced production costs due to a reduction in the amount of expensive alloy elements (such as a Ni amount of $\leq 1\%$). By using this thicker high-strength steel plate of the present invention in all manner of welded structures including large ships, an increase in the size of the welded structure, a high level of safety in terms of cracking, improved welding efficiency during construction, and favorable economic viability of the steel plate used as the structural material can all be achieved at the same time, and therefore the industrial effect of the invention is immense.

BEST MODE FOR CARRYING OUT THE INVENTION

[0014] As follows is a description of embodiments of the method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention, and the thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding of the present invention.

In these embodiments, a detailed description is provided in order to facilitate better understanding of the intent of the present invention, although unless specifically stated, the embodiments in no way limit the scope of the present invention.

<Steel plate production conditions (production process)>

[0015] The demands placed on the steel plate used in welded structures such as ships, including (1) a high degree of strength at large plate thickness values, (2) favorable brittle fracture arrestability, (3) favorable HAZ toughness in high

heat-input welding, and (4) low production costs continue to become more stringent.

In response to these demands, a method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention includes: cooling a continuously cast slab containing, in mass % values, C: 0.05 to 0.12%, Si: not more than 0.3%, Mn: 1 to 2%, P: not more than 0.015%, S: not more than 0.005%, B: 0.0003 to 0.003%, V: 0.01 to 0.15%, Al: 0.001 to 0.1%, Ti: 0.005 to 0.02%, N: 0.002 to 0.01%, and O: not more than 0.004%, with the remainder being iron and unavoidable impurities, to a temperature of Ar_3 -200°C or lower after a continuous casting, subsequently reheating the slab to 950 to 1,100°C; next subjecting the continuously cast slab to rough rolling at 900°C or higher with a cumulative reduction ratio of at least 30%; subsequently performing finish rolling at 700°C or higher with a cumulative reduction ratio of at least 50% under conditions that both of the finish rolling start temperature and the finishing rolling completion temperature are not higher than a temperature represented by a formula: $\{-0.5 \times (\text{slab heating temperature (°C)}) + 1,325\}$ (°C), thereby forming a rolled plate; and then cooling the rolled plate to 500°C or lower by accelerated cooling to obtain a steel plate. In the above continuously cast slab, a calculated value of an amount of solid solution B {effective B amount: $B_{ef}(\%)$ } which is solid-solubilized into an austenite base material prior to transformation is not more than 0%, and a carbon equivalent C_{eq} satisfies a range from 0.32 to 0.42%.

If an amount of residual oxygen O_{Ti} (%) that remains after deoxidation by strong deoxidizing elements and is able to undergo deoxidation by Ti that is a weak deoxidizing element is an amount represented by formula (1) shown below, then the effective B amount B_{ef} (%) is represented by formula (2) shown below. Further, the carbon equivalent C_{eq} (%) is represented by formula (3) shown below, and the Ar_3 is represented by formula (4) shown below.

Furthermore, the "slab heating temperature" refers to the temperature used when reheating the continuously cast slab (namely, the reheating temperature).

$$O_{Ti} (\%) = O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al \quad \cdots (1)$$

{in formula (1), component elements that represent unavoidable impurities are also included within the calculation.}

$$B_{ef} (\%) = B - 0.77\{N - 0.29(Ti - 2O_{Ti})\} \quad \cdots (2)$$

{in formula (2), when $O_{Ti} \leq 0$, it is deemed that $O_{Ti} = 0$, when $O_{Ti} > 0$, it is deemed that $Ti - 2O_{Ti} \geq 0.005$ (%), and when $N - 0.29(Ti - 2O_{Ti}) \leq 0$ (although when $O_{Ti} \leq 0$, $O_{Ti} = 0$), it is deemed that $N - 0.29(Ti - 2O_{Ti}) = 0$.}

$$C_{eq} (\%) = C + Mn/6 + (Cr+Mo+V)/5 + (Ni+Cu)/15 \quad \cdots (3)$$

$$Ar_3 (\text{°C}) = (910 - 310C - 80Mn - 20Cu - 55Ni - 80Mo) \quad \cdots (4)$$

In this description, element symbols used within the formulas each represent the amount (mass % value) for that particular element within the continuously cast slab or the thicker high-strength steel plate.

Further, in the present invention, there are no particular limitations on the method used for producing the continuously cast slab. For example, after melting in a blast furnace, converter furnace, electric furnace, or the like, a component adjustment process can be conducted using any of the various secondary refining techniques to achieve the targeted amount of each element, and the slab may then be produced via a typical continuous casting method.

[0016] In the method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention, with regard to the elements listed in the above chemical composition, the lower limit for the S content may be set to 0.0005%, and the lower limit for the O content may be set to 0.001%. Moreover, if required, one or more selected from the group consisting of Ca: 0.0003 to 0.004%, Mg: 0.0003 to 0.004%, Ni: 0.01 to 1%, Cu: 0.01 to 1%, Cr: 0.01 to 1%, Mo: 0.01 to 0.5%, Nb: 0.003 to 0.03%, REM: 0.0003 to 0.02%, and Zr: 0.0003 to 0.02% may also be added selectively.

The abbreviation REM refers to "rare earth mentals", and represents one or more elements selected from Sc, Y, and the lanthanoid elements of La, Ce, Pr, Nd, Pm, Sm, Eu, Gd, Tb, Dy, Ho, Er, Tm, Yb and Lu.

[0017] The main point of the present invention is a technique in which the combined addition of B and V is conducted

in order to simultaneously achieve favorable strength, brittle fracture arrestability, high heat-input HAZ toughness, and low production costs for a thick steel plate produced by a TMCP, and by strictly controlling the amount of N that bonds to these nitride-forming elements B and V, the state in which the B and V exist within the austenite (γ) can be optimized; thereby, enabling the transformation structures of the base material or a high heat-input HAZ to be controlled.

Specifically, in terms of the existence state of B within the γ , the present invention is based on the technical concept of ensuring that, within both of the base material and the high heat-input HAZ, no solid solution B exists, and all of the B is precipitated as BN. In terms of the existence state of V within the γ , the present invention is based on the technical concept of employing solid solution V in the base material, but employing precipitated V (such as VN) within the high heat-input HAZ.

[0018] A more detailed description is presented below.

First, in order to satisfy the demand for favorable brittle fracture arrestability, which represents the major technical challenge for the present invention, various TMCP conditions were investigated to determine the conditions that enable the crystal grain size within the thick steel plate to be reduced as fine as possible.

Here, the smallest unit when brittle cracking occurs along a crystallographically unique crystal plane (the cleavage plane: corresponding with the {100} plane in the case of iron having a body-centered cubic structure) is termed the "fracture facet unit", and in the present invention, the metal structural unit of a size corresponding with this fracture facet unit is referred to as the "crystal grain size".

It became evident that if the low-temperature reheating and low-temperature rolling during the TMCP were conducted at the lowest possible temperatures to enable the refinement of the pre-transformation γ structures to be pushed to the limit, then the crystal grain size was able to be satisfactorily reduced even for a thick steel plate having a plate thickness of 50 to 80 mm; thereby, enabling the brittle fracture arrestability to satisfy the target range. When Ar_3 ($^{\circ}C$) is calculated using the formula: $(910 - 310C - 80Mn - 20Cu - 55Ni - 80Mo)$, the conditions that yield favorable brittle fracture arrestability involve cooling the continuously cast slab to a temperature of $\{Ar_3(^{\circ}C) - 200(^{\circ}C)\}$ or lower, subsequently conducting low-temperature heating (reheating) at a temperature of not more than $1,100^{\circ}C$, subjecting the continuously cast slab to rough rolling at $900^{\circ}C$ or higher with a cumulative reduction ratio of at least 30%, subsequently performing finish rolling at $700^{\circ}C$ or higher with a cumulative reduction ratio of at least 50% under conditions that both of the finish rolling start temperature ($^{\circ}C$) and the finishing rolling completion temperature ($^{\circ}C$) are not higher than the temperature represented by the formula: $\{-0.5 \times (\text{slab heating temperature } (^{\circ}C)) + 1,325\} (^{\circ}C)$, and then conducting accelerated cooling to cool the rolled plate to $500^{\circ}C$ or lower.

[0019] The first TMCP condition for taking full advantage of the low-temperature heating and low-temperature rolling requires that, after continuous casting, the slab (the continuously cast slab) is cooled to a temperature of $Ar_3 - 200^{\circ}C$ or lower to effect a γ (austenite) $\rightarrow \alpha$ (ferrite) transformation, and then effecting a $\alpha \rightarrow \gamma$ transformation by low-temperature heating (reheating) of the slab to a temperature of not more than $1,100^{\circ}C$. The reason for specifying this production condition is to ensure thorough grain size reduction (uniform grain refinement) of the γ during heating.

If the slab is subjected to reheating from a higher temperature that exceeds $\{Ar_3(^{\circ}C) - 200(^{\circ}C)\}$, then the reheating occurs before complete $\gamma \rightarrow \alpha$ transformation has occurred within the interior of the slab; thereby, coarse γ structures remain within the slab during casting. The formula (4) above is a relationship that applies only for an extremely slow cooling rate when the slab is cooled after continuous casting, and does not apply in cases such as thick plate rolling where the cooling rate is comparatively large.

If the slab is reheated at a comparatively high temperature exceeding $1,100^{\circ}C$, then Ostwald growth of TiN tends to begin; thereby, the pinning effect diminishes, and it becomes difficult to generate uniformly refined γ grains in a stable manner. If a thorough grain size reduction (uniform grain refinement) of the γ cannot be achieved during heating, then under practical slab thickness restrictions (typically 200 to 400 mm), it is difficult for a steel plate having a plate thickness of 50 to 80 mm to achieve a satisfactory reduction in the size of the pre-transformation γ structures, regardless of any innovations that may be introduced in terms of rolling conditions.

[0020] The second TMCP condition for taking full advantage of the low-temperature heating and low-temperature rolling requires that rough rolling is conducted at $900^{\circ}C$ or higher with a cumulative reduction ratio of at least 30%. The reason for specifying this production condition is to ensure that by conducting rolling within the recrystallization region, γ structures can be obtained that have an even finer grain structure than that obtained upon heating.

If the rough rolling is conducted at a temperature less than $900^{\circ}C$ or with a cumulative reduction ratio of less than 30%, then the recrystallization is inadequate, strain-induced grain growth tends to occur, and there is a possibility that the resulting grains may actually be coarser than the initial γ generated during heating.

[0021] The third TMCP condition for taking full advantage of the low-temperature heating and low-temperature rolling requires that finish rolling is performed at $700^{\circ}C$ or higher with a cumulative reduction ratio of at least 50% under conditions that both of the finish rolling start temperature ($^{\circ}C$) and the finishing rolling completion temperature ($^{\circ}C$) are not higher than the temperature represented by the formula: $\{-0.5 \times (\text{slab heating temperature } (^{\circ}C)) + 1,325\} (^{\circ}C)$. The reason for specifying this production condition is to ensure that the recrystallized grains that have undergone satisfactory grain size reduction (uniform grain refinement) during the rough rolling are rolled within the non-recrystallization region; thereby,

stretching the γ grains, increasing the grain boundary surface area, and activating the grain boundaries, as well as introducing deformation bands within the γ and maximizing the nucleation site density and the nucleation frequency in the pre-transformation γ .

If the cumulative reduction ratio of the finish rolling is less than 50%, or the condition requiring temperatures not higher than the temperature represented by the formula: $\{-0.5 \times (\text{slab heating temperature } (^{\circ}\text{C})) + 1,325\} (^{\circ}\text{C})$ is not satisfied, then the grain size reduction for the pre-transformation γ tends to be inadequate.

From a metallurgical perspective, the condition requiring temperatures not higher than that represented by the formula: $\{-0.5 \times (\text{slab heating temperature } (^{\circ}\text{C})) + 1,325\} (^{\circ}\text{C})$ means that the higher the heating temperature and the coarser the initial γ grains, the greater the necessity to conduct the finish rolling at a lower temperature to strengthen the non-recrystallization region rolling. For example, if the slab heating temperature is 1,100°C, then finish rolling must be conducted at 775°C or lower, whereas if the slab heating temperature is 1,000°C, then finish rolling must be conducted at 825°C or lower. In this manner, if an extremely strict TMCP condition that restricts the finish rolling temperature in conjunction with the slab heating temperature is not specified, then it is impossible to ensure favorable brittle fracture arrestability for thick steel plate in a stable manner.

If the finish rolling is conducted at a temperature lower than 700°C, then the surface of the steel plate starts to undergo transformation, either during rolling or during the standby period prior to accelerated cooling; thereby, causing a softening and coarsening of the surface structure, and as a result, the strength and the brittle fracture arrestability deteriorate.

[0022] The fourth TMCP condition for taking full advantage of the low-temperature heating and low-temperature rolling requires that accelerated cooling is applied to cool the rolled plate to 500°C or lower. The reason for specifying this production condition is because even in the case where the pre-transformation γ grains are made as fine as possible by applying the heating and rolling conditions outlined above, if the subsequent cooling is an air cooling process, then the degree of supercooling during the $\gamma \rightarrow \alpha$ transformation is small; thereby, the crystal grain size cannot be adequately reduced.

If the accelerated cooling is stopped at a temperature higher than 500°C, then within the interior of the steel plate of which the temperature is higher than that of the surface layer of the steel plate, the accelerated cooling stops and shifts to air cooling partway through the transformation, and as a result, the crystal grain size within the interior of the plate cannot be adequately reduced.

[0023] The above are the TMCP conditions required for ensuring a satisfactory reduction in the crystal grain size in order to achieve the required level of brittle fracture arrestability under the premise of a low Ni content, thus enabling the requirements (2) and (4) listed above to be satisfied.

However, with the above TMCP conditions, the combination of maximizing the crystal grain size reduction of the pre-transformation γ and the inherent slow cooling rate for the thick steel plate causes a problem in that the hardenability upon transformation tends to decrease dramatically. As a result, within bainite/ferrite mixed structures, the bainite fraction tends to decrease while the ferrite fraction increases, and it becomes difficult to ensure a predetermined level of tensile strength. At the same time, the hardenability generated by the solid solution B within the γ tends to become unstable under the above type of TMCP conditions; thereby, not only is the strength inadequate, but the variation in strength also tends to increase. In this manner, the above TMCP conditions raise a new problem in that the requirement (1) described above cannot be satisfied.

A first reason for the variation in strength is that the amount of solid solution B within the γ , which can be estimated by the effective B amount (Bef) described below, increases or decreases with fluctuations in the steel composition during mass production (including fluctuations in the amount of O, the amount of strong deoxidizing elements, the amount of Ti, the amount of N, and the amount of B). A second reason is that for low-temperature rolled γ in the non-recrystallization region, the amount of strain-induced precipitation of iron borocarbides (such as $\text{Fe}_{23}(\text{C},\text{B})_6$) varies depending on the rolling conditions and the length of the standby period from the completion of rolling to the start of the accelerated cooling, and the reverse side of this variation in precipitation is an increase or decrease in the amount of solid solution B within the γ . As mentioned above, it is far from easy to ensure a stable level of strength for the base material on the basis of the B hardenability under the above TMCP conditions, and therefore, a strengthening technique other than B hardenability must be utilized.

[0024] Accordingly, in the present invention, in order to satisfy the requirement (1) mentioned above, the two techniques described below are employed to ensure a satisfactory and stable base material strength.

The first technique involves precipitating all of the B as BN during the TMCP so that no solid solution B exists within the γ ; thereby, eliminating any instability in the hardenability caused by fluctuations in the amount of solid solution B within the γ . This technique represents the complete opposite thinking to conventional techniques that utilize B, and is based on the technical concept of not using the property of B hardenability to ensure base material strength. This enables variations in the strength during mass production to be suppressed. Specifically, the effective B amount (Bef) described below is controlled to a value of not more than 0%. In the present invention, the significance of adding B applies only to the high heat-input HAZ, and a description of this significance is presented below.

The second technique involves utilizing precipitate strengthening due to V carbides to increase the strength of the base

material.

Under the TMCP conditions described above, it was ascertained that by adding 0.01% of V, the tensile strength of a material having a plate thickness of 70 mm could be increased by approximately 10 MPa, and it became evident that the addition of V was an extremely effective technique for strengthening the steel plate in a quantitative manner. This is because bainite/ferrite mixed structures of which the grain size has been satisfactorily reduced via application of low-temperature heating and low-temperature rolling are ideal base materials for V carbides (such as VC and V_4C_3) to precipitate finely and in high density during the accelerated cooling and tempering treatments. In the present invention, another significant reason for adding V relates to the high heat-input HAZ, and a description of this reason is presented below.

[0025] As described above, in order to ensure satisfactory base material strength in the TMCP by adding V but not employing B hardenability, the carbon equivalent Ceq value, which is employed as an indicator of the hardenability of the steel components excluding B, must be at least 0.32%, the effective B amount Bef must be restricted to not more than 0%, at least 0.01% of V must be added, the heating temperature must be controlled at a temperature of 950°C or higher, and the accelerated cooling must be continued to a temperature of 500°C or lower.

If Ceq is less than 0.32%, then even if V is added, it is difficult to ensure a stable base material strength. Moreover, softening of HAZ proceeds and there is a possibility that the tensile strength of welded joints may be inadequate.

In those cases where the effective B amount calculated using the above formula (2) is a numerical value exceeding 0%, solid solution B exists within the γ , and there is a possibility of B hardenability manifesting; thereby, causing a variation in the strength.

If the heating temperature is less than 950°C, then because the solubilization of the V carbonitrides tends to be inadequate, and the amount of solid solution V within the γ becomes unsatisfactory. As a result, the amount of V carbides that precipitate during the accelerated cooling and tempering treatments may be inadequate; thereby, stable base material strength cannot be ensured.

If air cooling is employed rather than accelerated cooling, then the cooling rate is too slow, the ferrite grains tend to coarsen, and the bainite fraction decreases; thereby, a satisfactory transformation strengthening cannot be achieved.

If the accelerated cooling is stopped at a temperature higher than 500°C, then the accelerated cooling within the higher temperature interior of the plate stops partway through the transformation, and as a result, satisfactory transformation strengthening cannot be obtained within the plate interior.

During the accelerated cooling, it is preferable to ensure a water volume density of at least 0.3 m³/m²/min in terms of obtaining fine bainite/ferrite structures that exhibit a combination of strength and toughness.

The above description presents techniques that enable satisfactory strength to be obtained under TMCP conditions that emphasize brittle fracture arrestability, and under the premise of a low Ni content. These techniques enable the above requirements (1), (2) and (4) to be satisfied simultaneously.

[0026] Furthermore, after the accelerated cooling, a tempering heat treatment may be conducted at a temperature of 350 to 700°C for a period of 5 to 60 minutes. Although this increases the production costs, it enables the strength, the elongation, and the Charpy impact properties to be controlled precisely within predetermined ranges.

In those cases where the tempering heat treatment is performed at a temperature of less than 350°C or the tempering heat treatment time is less than 5 minutes, the effects of the tempering treatment do not manifest satisfactorily. Furthermore, if the temperature of the tempering heat treatment exceeds 700°C or the tempering heat treatment time exceeds 60 minutes, then the tempering phenomenon exceeds the optimal range and has an excessive effect; thereby, a marked reduction in the strength and a marked deterioration in the Charpy impact properties occur, and as a result, the optimum mechanical properties cannot be obtained.

[0027] Next is a description of a technique for satisfying the above requirement (3) for favorable HAZ toughness in high heat-input welding.

The main factors governing HAZ toughness in high heat-input welding in the present invention can be broadly classified into the following three areas. The first factor is hardness, the second factor is the MA (martensite-austenite mixed phase), and the third factor is the effective crystal grain size.

In the present invention, for reasons of both of the hardness and the MA, the carbon equivalent Ceq is restricted to not more than 0.42%. If the carbon equivalent exceeds 0.42%, then the HAZ becomes excessively hard and the MA increases; thereby, a significant increase in the brittleness of the HAZ occurs.

Moreover, by restricting the effective B amount (Bef) to not more than 0%, B hardenability can be prevented from occurring within the HAZ, and an increase in the hardness and an increase in the amount of MA can be suppressed.

[0028] The inventors of the present invention discovered the advantage V addition offered in terms of the hardness. Further, they also found that in cases such as the present invention where the HAZ is mainly bainite, the HAZ is resistant to hardening even when V is added.

In other words, if the base material is strengthened by addition of an element other than V such as C or Mn, then the HAZ mainly containing bainite hardens dramatically, and the brittleness of the HAZ significantly increases. In contrast, if the base material is strengthened by adding V as per the present invention, then the hardening of the HAZ mainly

containing bainite is suppressed. Based on this new finding, if the amounts of C and Mn are reduced so as to cancel out the increase in base material strength provided by V, resulting in a lower C_{eq} value, then the hardness in the HAZ is reduced by an amount equivalent to the reduction in the C_{eq} ; thereby, the HAZ toughness is improved. This type of technique in which the HAZ toughness is improved by utilizing the difference in the V hardening behavior between the base material and the HAZ has not existed conventionally.

[0029] In the present invention, from the viewpoint of the MA, the amount of Si must be reduced as low as possible. Further, under the TMCP conditions of the present invention, although the contribution of Nb to the base material is small, it promotes MA growth. In the comparatively high C_{eq} range of the present invention, although Mo is expensive, it promotes MA growth. Accordingly, Nb and Mo must be reduced as low as possible in the present invention.

[0030] In the present invention, from the viewpoint of effective crystal grain size, two techniques are employed to reduce the size of the HAZ structures.

The first technique involves simultaneously using the B precipitates and V precipitates within the γ as transformation nuclei. By suitably increasing the N amount so that the effective B amount {Bef (%)} represented by the above formula (2) is not more than 0%, BN, VN and V(C,N) are precipitated at the γ grain boundaries and within the γ grains during the cooling after high heat-input welding, and any one or more of these precipitated particles function effectively as transformation nuclei for not only ferrite, but also for bainite; thereby, ensuring a favorable reduction in size of the HAZ structures.

Moreover, the second technique for reducing the size of the HAZ structures involves appropriate addition of Ca and/or Mg to ensure dispersion of a multitude of very fine oxides or sulfides; thereby, suppressing γ grain growth by a pinning effect and ensuring a very fine bainite packet size. Co-precipitation of B precipitates and V precipitates occurs within a portion of the fine oxides and/or sulfides, and a transformation nucleus function is imparted to the pinning particles; thereby, the effect can also be obtained which enables the bainite that transforms from the γ grain boundaries to be made even finer.

The HAZ structure size reduction techniques described above are able to effectively lower the HAZ hardenability, and therefore contribute to reducing the amount of MA and the hardness. The first technique ensures a favorable Charpy absorption energy at -20°C, and if the second technique is used in combination with the first technique to enable maximum reduction in the size of the HAZ structures, a favorable Charpy absorption energy can be obtained at -40°C.

By adopting the measures described above for reducing the hardness, reducing the MA, and reducing the size of the HAZ structures, a high heat-input HAZ according to the present invention is able to achieve a high vE (-20°C) value. Accordingly, the above requirement (3) can be satisfied in addition to the requirements (1), (2) and (4).

<Chemical element composition (thicker high-strength steel plate)>

[0031] As described above, in order to satisfy the above-mentioned requirements, namely (1) a high degree of strength at large plate thickness values, (2) favorable brittle fracture arrestability, (3) favorable HAZ toughness in high heat-input welding, and (4) low production costs, a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention includes, in mass % values, C: 0.05 to 0.12%, Si: not more than 0.3%, Mn: 1 to 2%, P: not more than 0.015%, S: not more than 0.005%, B: 0.0003 to 0.003%, V: 0.01 to 0.15%, Al: 0.001 to 0.1%, Ti: 0.005 to 0.02%, N: 0.002 to 0.01 %, and O: not more than 0.004%, with the remainder being iron and unavoidable impurities, wherein if the amount of residual oxygen that remains after deoxidation by strong deoxidizing elements and is able to undergo deoxidation by Ti which is the weak deoxidizing element is an amount represented by formula (5) shown below, then the calculated value of the amount of B {effective B amount: Bef(%)} which is solid-solubilized into the austenite base material prior to transformation is not more than 0%, the carbon equivalent C_{eq} represented by formula (7) shown below satisfies a range from 0.32 to 0.42%, the plate thickness is within a range from 50 to 80 mm, the yield strength is in the order of 390 to 460 MPa, the tensile strength is in the order of 510 to 570 MPa, the temperature $T_{Kca=6000}$ at which the brittle fracture arrestability Kca reaches 6,000 N/mm^{1.5} is -10°C or lower, and the Charpy impact absorbed energy vE (-20°C), which is an indicator of the high heat-input HAZ toughness with 20 kJ/mm or greater of heat input, is at least 47 J.

$$O_{Ti} (\%) = O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al \quad \cdots (5)$$

{in formula (5), component elements that represent unavoidable impurities are also included within the calculation.}

$$Bef (\%) = B - 0.77\{N - 0.29(Ti - 2O_{Ti})\} \quad \cdots (6)$$

{in formula (6), when $O_{Ti} \leq 0$, it is deemed that $O_{Ti} = 0$, when $O_{Ti} > 0$, it is deemed that $Ti - 20_{Ti} \cdot 0.005$ (%), and when $N - 0.29(Ti - 20_{Ti}) \leq 0$ (although when $O_{Ti} \leq 0$, $O_{Ti} = 0$), it is deemed that $N - 0.29(Ti - 20_{Ti}) = 0$ }

$$C_{eq} (\%) = C + Mn/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15 \quad \cdots (7)$$

With regard to the above formulas (5) to (7), formula (5) is the same as the aforementioned formula (1), formula (6) is the same as the aforementioned formula (2), and formula (7) is the same as the aforementioned formula (3).

[0032] Furthermore, in the thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention, with regard to the elements listed in the above chemical composition, the lower limit for the S content may be set to 0.0005%, and the lower limit for the O content may be set to 0.001%. Moreover, if required, the steel plate may also selectively include one or more selected from the group consisting of Ca: 0.0003 to 0.004%, Mg: 0.0003 to 0.004%, Ni: 0.01 to 1%, Cu: 0.01 to 1%, Cr: 0.01 to 1%, Mo: 0.01 to 0.5%, Nb: 0.003 to 0.03%, REM: 0.0003 to 0.02%, and Zr: 0.0003 to 0.02%.

A description of the reasons for restricting the amounts of the chemical elements within the steel (thicker high-strength steel plate) of the present invention is presented below.

[C: carbon] 0.05 to 0.12%

[0033] C is an important element for increasing the steel strength. In a thick steel plate prepared by a TMCP that takes full advantage of low-temperature heating and low-temperature rolling, at least 0.05% of C must be added to ensure that a predetermined level of strength is obtained in a stable manner. Further, for the reasons outlined below, the amounts added of Nb, Ni and Mo within the present invention must be suppressed to the minimum amounts required, and therefore it is problematic to strengthen the steel by increasing the amounts of these elements. Accordingly, C becomes an extremely important strengthening element. Moreover, C also has the effect of promoting precipitation of V(C,N) transformation nuclei within a high heat-input HAZ. However, in order to ensure favorable HAZ toughness is achieved in a stable manner, the amount of C must be restricted to not more than 0.12%, and in order to enhance the HAZ toughness, an amount of not more than 0.10% is preferred.

[Si: silicon] not more than 0.3%

[0034] Si has a deoxidizing action, but is unnecessary in those cases where Al which is a powerful deoxidizing element is added in a sufficient amount. Si also has the effect of strengthening the base material, but that effect is comparatively weak compared with those of other elements. Moreover, in a high heat-input HAZ of the present invention which requires a comparatively high carbon equivalent C_{eq} , there is a considerable danger that Si may promote MA growth, and therefore the Si content must be suppressed to not more than 0.3%. From the viewpoint of the HAZ toughness, the amount of added Si is preferably suppressed as low as possible, to an amount of not more than 0.20%. In terms of ensuring favorable strength and satisfactory deoxidation, Si is preferably added in an amount of at least 0.01%.

[Mn: manganese] 1 to 2%

[0035] In order to ensure an economical improvement in the strength of the steel, the amount of added Mn must be at least 1%, and is preferably 1.40% or greater. However, if Mn is added in an amount exceeding 2%, then the harmful effects of center segregation within the slab become quite marked, and hardening of the high heat-input HAZ and promotion of MA generation are also promoted; thereby, embrittlement proceeds, and therefore 2% is set as the upper limit. In order to prevent this embrittlement, the amount of Mn is preferably restricted to not more than 1.60%.

[P: phosphorus] not more than 0.015%

[0036] P is an impurity element, and must be reduced to not more than 0.015% in order to ensure that favorable brittle fracture arrestability and favorable HAZ toughness in high heat-input welding can be achieved in a stable manner. In order to enhance the HAZ toughness, the amount of P is preferably 0.010% or less.

[S: sulfur] 0.0005 to 0.005%

[0037] S must be suppressed to not more than 0.005%. If the amount of S exceeds 0.005%, then it tends to cause a portion of the sulfides to coarsen and act as crack origins which are harmful, and the toughness of both of the base

material and the high heat-input HAZ tend to deteriorate. In order to minimize these harmful effects, the S content is preferably not more than 0.003%. On the other hand, in order to utilize the HAZ pinning effect, the amount of S must be at least 0.0005%. The reason for this requirement is to ensure that by appropriate addition of Ca and Mg, a multitude of fine sulfides can be dispersed in the vicinity of the HAZ fusion line; thereby, strengthening the pinning effect to enable better size reduction of the γ grains for the purpose of increasing the HAZ toughness. If the amount of S is less than 0.0005%, then the number of sulfides tends to be inadequate; thereby, a satisfactory pinning effect cannot be obtained.

[B: boron] 0.0003 to 0.003%

[0038] B is a feature element within the present invention. As already described in detail, in the present invention, in both of the base material and the high heat-input HAZ, the calculated value of the effective B amount (Bef) represented by the above formula (2) is controlled to a value of not more than 0% so as to precipitate all of the B as BN in a state where no solid solution B exists within the γ ; thereby, eliminating any B hardenability. The BN particles precipitated within the γ function as transformation nuclei, and improve the toughness by reducing the size of the HAZ structures, reducing the hardness, and reducing the amount of MA. For this reason, B must be added in an amount of not less than 0.0003%. However, if an amount of B exceeding 0.003% is added, then coarse B precipitates are produced; thereby, a deterioration in the HAZ toughness is caused, and therefore 0.003% is set as the upper limit. In order to ensure favorable HAZ toughness in a stable manner, the B content is preferably not more than 0.0020%.

[V: vanadium] 0.01 to 0.15%

[0039] V is a feature element in the present invention. As already described in detail, V effectively strengthens the base material under the TMCP conditions of the present invention. On the other hand, V also suppresses the increasing of MA and the hardening within the high heat-input HAZ of the present invention, and the VN and V(C,N) which are precipitated within the γ act as transformation, nuclei; thereby, reducing the size of the HAZ structures and enhancing the toughness. In order to ensure these effects manifest satisfactorily, at least 0.01% of V must be added. However, if the amount of V exceeds 0.15%, then the refinement effect of HAZ microstructure becomes saturated, and at the same time, the HAZ hardness increases dramatically; thereby, a deterioration in the HAZ toughness is caused. Accordingly, the upper limit for the V content is 0.15%, and this limit is preferably 0.10% or lower.

[Al: aluminum] 0.001 to 0.1%

[0040] Al is a deoxidizing element, and is necessary for reducing O and enhancing the cleanliness of the steel. Elements other than Al such as Si, Ti, Ca, Mg, REM and Zr also exhibit deoxidizing activity, but even in the case where these other elements are added, if the Al content is not 0.001 % or greater, then it is difficult to stably suppress the amount of O (oxygen) to 0.004% or less. However, if the amount of Al exceeds 0.1%, then there is an increased tendency for coarse alumina-based oxides to form clusters; thereby, blockages of the steelmaking nozzles are caused or the coarse alumina-based oxides act as harmful crack origins, and therefore 0.1% is set as the upper limit. In order to minimize the possibility of these harmful effects, the Al content is preferably not more than 0.060%.

[Ti: titanium] 0.005 to 0.02%, [N: nitrogen] 0.002 to 0.01 %, and [effective B amount: Bef (%)] not more than 0% (for the calculated value from formula (2))

[0041] Ti bonds with N to form TiN, and contributes to the pinning effect during slab reheating and in the high heat-input HAZ, thus contributing to the reduction in the γ grain size. As a result, Ti reduces the size of the structures of the base material and the HAZ; thereby, enhancing the toughness. The remaining N after formation of the TiN bonds with B to form BN; thereby, precipitating all of the B as BN so that no solid solution B exists within the γ . As a result, the manifestation of B hardenability is prevented.

In order to enable the above effects to be achieved simultaneously, the amount of Ti must be 0.005 to 0.02%, the amount of N must be 0.002 to 0.01 %, and the calculated value of the effective B amount (Bef) represented by the above formula (2) must be not more than 0%.

If the amounts of Ti and N do not reach 0.005% and 0.002% respectively, then the pinning effect due to TiN does not manifest satisfactorily, and the toughness of the base material and the HAZ tend to deteriorate. If the amounts of Ti and N exceed 0.02% and 0.01% respectively, then the TiC precipitates and the amount of solid solution N increases; thereby, causing the toughness of the base material and the HAZ to deteriorate. In order to better enhance the HAZ toughness, the amounts of Ti and N are preferably not more than 0.015% and 0.007%, respectively. Moreover, even in the case where the amounts of Ti and N are within the appropriate ranges, if the effective B amount exceeds 0%, then the amount of solid solution B within the γ increases and B hardenability appears; thereby, causing variation in the base material

strength and hardening (embrittlement) of the HAZ.

[0042] A description of the thinking related to the effective B amount is presented below.

The Ti added as a chemical component may sometimes be consumed by the deoxidation that occurs within the melted steel (this is more likely in cases where the amount of A1 is low), and the residual Ti left after this deoxidation forms TiN within the solidified γ . During this process, if N exists in excess relative to the amount of Ti, then the N that remains after formation of TiN bonds with a portion of the B to form BN. The residual B that is left after formation of the BN exists as solid solution B that yields hardenability. In the present invention, the amount of this solid solution B within the γ that contributes to the hardenability is referred to as the effective B amount Bef(%).

[0043] A method of calculating the effective B amount Bef based on the added amount of each element, the thermodynamic reaction sequence, and the stoichiometric composition of the product is described below.

Firstly, the assumption is made that in order of deoxidizing power, Ca, Mg, REM (rare earth metal elements), Zr and Al undergo bonding with O. The amount of deoxidized O is calculated on the assumption that the deoxidation products are CaO, MgO, REM₂O₃, ZrO₂ and Al₂O₃, respectively.

In those cases where the deoxidation is not completed by these elements having a stronger deoxidizing power than Ti, if the amount of residual oxygen O_{Ti} (%) that remains after deoxidation by the strong deoxidizing elements and is able to undergo deoxidation by Ti which is the weaker deoxidizing element is an amount represented by formula (1) shown below, then the formula.: {O_{Ti} (%) > 0} is satisfied.

$$O_{Ti} (\%) = O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al \cdots (1)$$

wherein in the formula (1), component elements that represent unavoidable impurities are also included within the calculation.

[0044] In this case, the residual O (=O_{Ti}) undergoes deoxidation by Ti. Assuming the production of Ti₂O₃, the amount of residual Ti obtained by subtracting the amount of Ti consumed as a result of deoxidation is represented by Ti - 2O_{Ti} ≥ 0.005 (%), and this value must be at least 0.005%. Here, the reason that the amount of residual Ti obtained by subtracting the Ti consumed as a result of deoxidation must be at least 0.005% is to ensure that, as described above, the amount of TiN required for the present invention is obtained.

If the amount of residual Ti obtained by subtracting the Ti consumed as a result of deoxidation is less than 0.005%, then the pinning effect due to TiN does not manifest satisfactorily, and the toughness of the thick base material and the high heat-input HAZ tend to deteriorate.

[0045] Furthermore, the 0.005% or more of Ti that remains after the deoxidation forms TiN, and if any N remains after this TiN formation, then the following formula yields a positive value, whereas if no N remains, the formula yields either 0 or a negative value.

N - 0.29(Ti - 2O_{Ti}) > 0 : in cases where N remains

N - 0.29(Ti - 2O_{Ti}) ≤ 0 : in cases where no N remains

[0046] Furthermore, in those cases where the above formula {N - 0.29(Ti - 2O_{Ti})} yields a positive value, a portion of the B is consumed as BN; thereby, the effective B amount Bef can be calculated using formula (2) shown below.

$$Bef (\%) = B - 0.77\{N - 0.29(Ti - 2O_{Ti})\} \cdots (2)$$

wherein in the formula (2), when O_{Ti} ≤ 0, it is deemed that O_{Ti} = 0, when O_{Ti} > 0, it is deemed that {Ti - 2O_{Ti} ≥ 0.005 (%)}, and when {N - 0.29(Ti - 2O_{Ti}) ≤ 0 (although when O_{Ti} ≤ 0, O_{Ti} = 0)}, it is deemed that {N - 0.29(Ti - 2O_{Ti}) = 0}.

[0047] Furthermore, in those cases where the formula {N - 0.29(Ti - 2O_{Ti})} yields 0 or a negative value and no N remains, the effective B amount Bef is represented by simply by Bef (%) = B.

[0048] In terms of the coefficients for Ca, Mg, REM, Zr and Al within the above formula for the amount of residual oxygen O_{Ti}, if the assumption is made that the products (oxides) of the deoxidation reactions (oxidation reactions) within the melted steel are CaO, MgO, REM₂O₃, ZrO₂ and Al₂O₃ respectively, then the amount of O that exists as each of these oxides can be calculated as a mass % value. For example, in the case of CaO, because the atomic weight of Ca is 40 and that of O is 16, relative to the mass % of Ca, 16/40 = 0.4 of O is bonded. In the case of Al₂O₃, the atomic weights are 27 for Al and 16 for O, and therefore relative to the mass % of Al, (16 × 3)/(27 × 2) = 0.89 of O is bonded. Using similar calculations, the coefficients for each of the elements in the above O_{Ti} formula (0.66 for Mg, 0.17 for REM, and 0.35 for Zr) can be determined.

[0049] Furthermore, if the derived formula concept for the effective B amount is represented in a backward manner from the low-temperature side to the high-temperature side, then the following sequence is obtained.

Effective B amount B_{ef} (%) = amount of component B - B as BN

→ B as BN = 0.77(N - N as TiN)

→ N as TiN = 0.29(Ti - Ti as Ti₂O₃)

→ Ti - Ti as Ti₂O₃ = 2(O - O as CaO - O as MgO - O as REM₂O₃ - O as ZrO₂ - O as Al₂O₃)

→ O as CaO = 0.4Ca

→ O as MgO = 0.66Mg

→ O as REM₂O₃ = 0.17REM

→ O as ZrO₂ = 0.35Zr

→ O as Al₂O₃ = 0.89Al

[0050] Next, if the derived formula concept for the effective B amount is represented in the reaction sequence from the high-temperature side to the low-temperature side, then the following sequence is obtained. In other words, in the steps from refining → solidification during the steelmaking process, the reaction sequence is as follows.

[1] Deoxidation reaction within liquid phase (melted steel) (in the vicinity of 1,600°C)

[0051] The deoxidation reactions occur in the order of the strength of the chemical affinity of each element for O, namely CaO → MgO → REM₂O₃ → ZrO₂ → Al₂O₃; thereby, reducing the dissolved O within the melted steel. In those cases where these reactions complete the deoxidation process, the formula $O_{Ti} \leq 0$ satisfies. In those cases where the deoxidation is not complete and residual dissolved oxygen still exists, the formulas $O_{Ti} > 0$ and $Ti - 2O_{Ti} \geq 0.005$ (%) satisfy, so that Ti which is a weaker deoxidizing element than Al contributes to deoxidation via formation of Ti₂O₃, and the amount of residual Ti obtained by subtracting the amount of Ti consumed as Ti₂O₃ from the amount of the component Ti is at least 0.005%.

[2] Denitrification reaction within solid phase (solidified γ) (near 1,300°C to near 800°C)

[0052] The denitrification reactions occur in the order of the strength of the chemical affinity of each element for N, namely TiN → BN → AlN; thereby, reducing the amount of solid solution N within the solid phase γ . First, the residual Ti left after consumption by the deoxidation undergoes a denitrification reaction. In those cases where this reaction completes the denitrification process, the formula $N - 0.29(Ti - 2O_{Ti}) \leq 0$ satisfies and no solid solution N exists within the γ ; thereby, B does not form BN but rather exists entirely as solid solution B. In contrast, in those cases where the denitrification is not completed by Ti and therefore, residual solid solution N still exists, the formula $N - 0.29(Ti - 2O_{Ti}) > 0$ satisfies and a portion of the B generates BN, while the remainder exists as solid solution B.

[0053] On the other hand, in those cases where the deoxidation is completed by one or more elements having a stronger deoxidizing power than Ti, the formula below is satisfied.

$$O_{Ti} \leq 0$$

In these cases, the Ti is not consumed by deoxidation. The Ti forms TiN, and if any N remains, the following formula is satisfied.

$$N - 0.29Ti > 0$$

The effective B amount B_{ef} in such cases is calculated using the following formula.

$$B_{ef} (\%) = B - 0.77(N - 0.29Ti)$$

If the Ti forms TiN and no residual N remains, then the formula below is satisfied.

$$N - 0.29Ti \leq 0$$

The effective B amount B_{ef} in such cases is calculated using the following formula.

$$B_{ef} (\%) = B$$

[0054] In each of the above formulas, the expression 0.29Ti within the formula $(N - 0.29Ti)$ represents N as TiN. Here, the atomic weights are 48 for Ti and 14 for N, and therefore relative to the mass % of Ti (strictly speaking, relative to the mass of residual Ti after subtraction of the mass of Ti consumed by deoxidation), $14/48 = 0.29$ of N is bonded. Further, if $N - 0.29Ti \leq 0$, then all of the N is fixed as TiN, and no solid solution N exists within the γ base material. On the other hand, if $N - 0.29Ti > 0$, then solid solution N exists in the γ base material in addition to TiN, and therefore this solid solution N bonds with B to generate BN; thereby, reducing the effective B amount.

[O: oxygen] 0.001 to not more than 0.004%

[0055] O must be suppressed to not more than 0.004%. If the amount of O exceeds 0.004%, then it tends to cause a portion of the oxides to coarsen and act as crack origins which are harmful, and the toughness of both of the base material and the high heat-input HAZ tend to deteriorate. On the other hand, in order to utilize the HAZ pinning effect, the amount of O must be at least 0.001%. The reason for this requirement is to ensure that by appropriate addition of Ca and Mg, a multitude of fine oxides can be dispersed in the vicinity of the HAZ fusion line; thereby, strengthening the pinning effect to enable better size reduction of the γ grains for the purpose of increasing the HAZ toughness. If the amount of O is less than 0.001%, then the number of oxides tends to be inadequate; thereby, a satisfactory pinning effect cannot be obtained.

[Ca: calcium] 0.0003 to 0.004%, and [Mg: magnesium] 0.0003 to 0.004%

[0056] By adding at least 0.0003% of either one or both of Ca and Mg, with due consideration of the order of addition to the molten steel, oxides and/or sulfides containing Ca and/or Mg and having a particle size of 10 to 500 nm can be generated in an amount of at least 1,000 particles/mm². If the amount(s) of Ca and/or Mg are less than 0.0003%, then there is a possibility that the number of oxides or sulfides that function as pinning particles for the high heat-input HAZ may be insufficient. Furthermore, if each of the added amounts exceeds 0.004%, then the oxides and/or sulfides tend to coarsen, and not only may the number of pinning particles be insufficient, but there is a strong possibility that the coarse particles may act as crack origins which are harmful; thereby, there is a possibility that favorable HAZ toughness cannot be obtained.

[Ni: nickel] 0.01 to 1%

[0057] Ni is an element that is effective in suppressing deterioration in the toughness while ensuring the strength of the steel. For this reason, at least 0.01% of Ni must be added. However, the alloy cost of Ni is extremely high, and may cause the introduction of surface blemishes. Accordingly, the Ni content must be suppressed to not more than 1%. Further, in order to avoid surface blemishes, the Ni content is preferably reduced as low as possible, and it is preferable to restrict the Ni content to not more than 0.7% or to not more than 0.5%.

[Cu: copper] 0.01 to 1%, [Cr: chromium] 0.01 to 1%, and [Mo: molybdenum] 0.01 to 0.5%

[0058] Cu, Cr and Mo are effective in ensuring favorable strength, and each exhibits a satisfactory effect at an added amount of 0.01 % or more. On the other hand, in terms of avoiding a deterioration in the HAZ toughness in high heat-input welding, the upper limits for these elements are 1%, 1% and 0.5% respectively, and the amounts are preferably restricted to not more than 0.4%, 0.3% and 0.1% respectively. Cr and Mo are particularly expensive elements similar to Ni, and there is also a significant risk that they may promote MA growth in the HAZ, and therefore Cr and Mo are preferably not added.

[Nb: niobium] 0.003 to 0.03%

[0059] Nb is effective in promoting non-recrystallization region rolling during finish rolling. For this reason, at least

0.003% of Nb is preferably added. However, Nb is harmful in terms of the HAZ toughness in high heat-input welding. Accordingly, in the present invention, a very small amount of Nb of not more than 0.03% may be added to promote non-recrystallization region rolling. From the viewpoint of the HAZ toughness, the amount of Nb is preferably suppressed to not more than 0.02%, or more preferably to not more than 0.01%. Furthermore, in those cases where a large cumulative reduction ratio can be achieved during finish rolling, a satisfactory size reduction in the base material structures is realized; thereby, favorable brittle fracture arrestability can be achieved even without adding Nb. Therefore, in terms of the HAZ toughness, not adding Nb is particularly desirable.

[REM: rare earth metal elements (lanthanoid-based elements)] 0.0003 to 0.02%, and [Zr; zirconium] 0.0003 to 0.002%

[0060] REM (rare earth metal elements) and Zr contribute to deoxidation and desulfurization, suppress the generation of coarse stretched MnS in the center segregation zone and convert sulfides to harmless spherical forms; thereby, improving the toughness of the base material and the high heat-input HAZ. In order to realize these effects, the lower limits for both of REM and Zr are 0.0003%. However, if the added amounts of these elements are increased, then the effects are soon saturated, and therefore from the viewpoint of economic viability, the upper limits for REM and Zr are set to 0.02% in both cases. The REM added in the present invention refers to lanthanoids such as La, Ce, and the like.

[0061] As described above, according to the method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding, and the thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding of the present invention, by adding the steel components so that each element is included in an amount that satisfies the range described above and the element composition satisfies the relational formulas listed above, and by setting each of the production conditions in the manner described above, a thicker high-strength steel plate can be obtained that realizes (1) high strength for the thick steel plate including a yield strength in the order of 390 to 460 MPa (namely, a tensile strength in the order of 510 to 570 MPa) for a plate thickness of 50 to 80 mm, (2) favorable brittle fracture arrestability indicated by an arrestability indicator $T_{kca=6000} \leq -10^{\circ}\text{C}$, (3) favorable HAZ toughness in high heat-input welding indicated by $vE (-20^{\circ}\text{C}) \geq 47 \text{ J}$ even when the heat input during welding is $\geq 20 \text{ kJ/mm}$, and (4) reduced production costs due to a reduction in the amount of expensive alloy elements (such as a Ni amount of $\leq 0.5\%$).

By using this type of thicker high-strength steel plate of the present invention in all manner of welded structures, including large ships, an increase in the size of the welded structure, a high level of safety in terms of cracking, improved welding efficiency during construction, and favorable economic viability of the steel plate used as a structural material can all be achieved at the same time, and therefore the industrial effect of the invention is immense.

EXAMPLES

[0062] A description of specifics of the present invention is presented below, using examples of the method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding of the present invention, and examples of the thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding of the present invention. However, the present invention is in no way limited by the examples presented below, and various modifications may be made within the scope of the invention described above and below, and such modifications are all deemed to be included within the technical scope of the present invention.

[Sample preparation]

[0063] By performing deoxidation and desulfurization of the melted steel and controlling the chemical element composition during the steelmaking process, and then conducting continuous casting, slabs (continuously cast slabs) were prepared with the chemical component compositions shown below in Tables 1 to 4. Subsequently, using the production conditions shown below in Tables 5 to 10, each slab was reheated and subjected to thick plate rolling to realize a finished plate thickness of 50 to 80 mm, and was then subjected to accelerated cooling, and if necessary, further subjected to off-line tempering, thus completing preparation of a thick steel plate sample.

[0064] In these examples, lists of the chemical element compositions for thick steel plates according to the steel of the present invention are shown in Tables 1 and 2, and lists of the chemical element compositions for comparative steels are shown in Tables 3 and 4. Further, lists of the production conditions for the steel plates of the steels according to the present invention are shown in Tables 5 and 6, whereas lists of the production conditions for the steel plates of the comparative steels are shown in Tables 7 and 8. Furthermore, lists of the conditions for comparative steels in which the steel plate is produced using the chemical element composition of "Steel No. 1" of the present invention shown in Tables 1 and 2, but with altered values for various production conditions, are shown in Tables 9 and 10.

[0065] In Tables 2 and 4, Ceq, formula A, formula B, formula C, formulas D, and Ar₃ are each defined as follows.

$$C_{eq} = C + Mn/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15$$

$$\text{Formula A} = O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al$$

$$\text{Formula B} = N - 0.29Ti$$

$$\text{Formula C} = Ti - 2(O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al)$$

$$\text{Formula D} = N - 0.29[Ti - 2(O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al)]$$

$$Ar_3 (^{\circ}C) = 910 - 310C - 80Mn - 20Cu - 55Ni - 80Mo \text{ (slab)}$$

Furthermore, the effective B amount is defined as follows.

(i) When the value of formula A < 0

(a) when the value of formula B > 0, effective B amount = B - 0.77(N - 0.29Ti)

(b) when the value of formula B ≤ 0, effective B amount = B

(ii) When the value of formula A ≥ 0

The value of formula C ≥ 0.005

(a) when the value of formula D > 0, effective B amount = B - 0.77{N - 0.29[Ti - 2(O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al)]}

(b) when the value of formulas D ≤ 0, effective B amount = B

[0066]

Table 1

Steel chemical components																	Units: mass %				
Steel No	C	Si	Mn	P	S	B	V	Al	Ti	N	O	Ca	Mg	Ni	Cu	Cr	Mo	Nb	REM	Zr	
1	0.05	0.15	1.95	0.006	0.003	0.0003	0.02	0.010	0.005	0.003	0.003	--	--	--	--	--	--	--	--	--	
2	0.08	0.10	1.65	0.010	0.001	0.0010	0.06	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--	
3	0.07	0.08	1.45	0.008	0.002	0.0011	0.04	0.015	0.012	0.005	0.001	--	--	--	--	--	--	0.018	--	--	
4	0.09	0.08	1.50	0.008	0.002	0.0020	0.05	0.001	0.015	0.006	0.004	0.0008	0.0005	--	--	--	--	--	--	--	
5	0.12	0.15	1.00	0.004	0.001	0.0023	0.01	0.002	0.009	0.005	0.004	0.0008	0.0004	0.1	--	0.6	--	--	--	--	
6	0.05	0.01	1.85	0.008	0.003	0.0004	0.01	0.015	0.005	0.002	0.001	--	--	0.8	--	--	--	0.015	--	--	
7	0.09	0.12	1.10	0.012	0.002	0.0012	0.03	0.008	0.014	0.007	0.002	0.0035	--	0.2	0.5	--	0.40	--	--	--	
8	0.08	0.15	1.20	0.005	0.002	0.0007	0.03	0.015	0.010	0.004	0.002	--	0.0010	0.3	0.8	--	0.06	0.008	--	--	
9	0.10	0.10	1.80	0.008	0.003	0.0005	0.11	0.034	0.017	0.009	0.002	--	--	--	--	--	--	--	--	--	
10	0.07	0.15	1.55	0.006	0.003	0.0010	0.09	0.001	0.010	0.005	0.004	0.0005	--	--	--	--	0.08	--	0.001	0.002	
11	0.09	0.15	1.10	0.006	0.003	0.0003	0.08	0.025	0.009	0.004	0.001	--	--	0.8	0.2	0.3	--	--	--	--	
12	0.08	0.03	1.50	0.006	0.004	0.0016	0.03	0.005	0.015	0.007	0.003	--	0.0035	--	--	--	--	--	--	--	
13	0.08	0.15	1.55	0.006	0.003	0.0016	0.06	0.003	0.012	0.006	0.004	0.0003	0.0018	0.5	--	--	0.05	0.012	--	--	
14	0.05	0.28	1.95	0.006	0.003	0.0003	0.02	0.010	0.005	0.003	0.003	--	--	--	--	--	--	--	--	--	
15	0.05	0.15	1.95	0.006	0.003	0.0003	0.02	0.080	0.005	0.003	0.003	--	--	--	--	--	--	--	--	--	
16	0.05	0.15	1.95	0.006	0.003	0.0003	0.02	0.010	0.005	0.003	0.003	--	--	0.05	0.05	--	--	--	--	--	

Inventive steel

[0067]

Table 2

Steel No.	Ceq	Formula A	Formula B	Formula C	Formula D	Effective B amount	Ar ₃ (°C)
1	0.38	<0	>0	-	-	-0.0009	739
2	0.37	<0	>0	-	-	-0.0003	753
3	0.32	<0	>0	-	-	-0.0001	772
4	0.35	>0	-	0.0101	>0	-0.0004	762
5	0.42	>0	-	0.0057	>0	-0.0003	787
6	0.41	<0	>0	-	-	0.0000	703
7	0.41	<0	>0	-	-	-0.0011	741
8	0.37	<0	>0	-	-	-0.0001	752
9	0.42	<0	>0	-	-	-0.0026	735
10	0.36	>0	-	0.0059	>0	-0.0015	758
11	0.42	<0	>0	-	-	-0.0008	746
12	0.34	<0	>0	-	-	-0.0004	765
13	0.39	>0	-	0.0120	>0	-0.0004	730
14	0.38	<0	>0	-	-	-0.0009	739
15	0.38	<0	>0	-	-	-0.0009	739
16	0.39	<0	>0	-	-	-0.0009	735

Inventive steel

[0068]

Table 3

Steel chemical components																				Units: mass %			
Steel No.	C	Si	Mn	P	S	B	V	Al	Ti	N	O	Ca	Mg	Ni	Cu	Cr	Mo	Nb	REM	Zr			
17	0.04	0.14	1.55	0.006	0.003	0.0010	0.05	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
18	0.13	0.15	1.55	0.007	0.001	0.0010	0.05	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
19	0.08	0.33	1.55	0.006	0.003	0.0010	0.05	0.010	0.008	0.004	0.002	--	--	0.5	0.5	--	--	--	--	--			
20	0.12	0.15	0.90	0.009	0.003	0.0010	0.05	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
21	0.07	0.15	2.10	0.006	0.003	0.0010	0.05	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
22	0.08	0.15	1.55	0.005	0.002	0.0002	0.05	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
23	0.11	0.16	1.70	0.006	0.003	0.0010	0.009	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
24	0.10	0.16	1.70	0.006	0.003	0.0010	0.16	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
25	0.08	0.18	1.55	0.004	0.001	0.0010	0.05	0.005	0.010	0.002	0.003	0.0010	--	--	--	--	--	--	--	--			
26	0.08	0.11	1.55	0.011	0.002	0.0010	0.05	0.001	0.015	0.003	0.004	0.0008	0.0005	--	--	--	--	--	--	--			
27	0.08	0.11	1.55	0.011	0.002	0.0010	0.05	0.001	0.010	0.002	0.004	0.0005	--	--	--	--	--	--	0.001	0.002			
28	0.08	0.15	1.55	0.018	0.003	0.0010	0.05	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
29	0.08	0.15	1.55	0.006	0.006	0.0010	0.05	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
30	0.08	0.15	1.55	0.006	0.003	0.0040	0.05	0.010	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
31	0.08	0.15	1.55	0.006	0.003	0.0010	0.05	0.0005	0.008	0.004	0.005	--	--	--	--	--	--	--	--	--			
32	0.08	0.15	1.55	0.006	0.003	0.0010	0.05	0.120	0.008	0.004	0.002	--	--	--	--	--	--	--	--	--			
33	0.08	0.15	1.55	0.006	0.003	0.0010	0.05	0.010	0.004	0.004	0.002	--	--	--	--	--	--	--	--	--			
34	0.08	0.15	1.55	0.006	0.003	0.0010	0.05	0.010	0.025	0.004	0.002	--	--	--	--	--	--	--	--	--			
35	0.08	0.15	1.55	0.006	0.003	0.0010	0.05	0.010	0.008	0.001	0.002	--	--	--	--	--	--	--	--	--			
36	0.08	0.15	1.55	0.006	0.003	0.0010	0.05	0.010	0.008	0.012	0.002	--	--	--	--	--	--	--	--	--			

Comparative steel

[0069]

Table 4

Steel No.	Ceq	Formula A	Formula B	Formula C	Formula D	Effective B amount	Ar ₃ (°C)
17	<u>0.31</u>	<0	>0	-	-	-0.0003	774
18	0.40	<0	>0	-	-	-0.0003	746
19	0.35	<0	>0	-	-	-0.0003	761
20	0.34	<0	>0	-	-	-0.0003	763
21	<u>0.43</u>	<0	>0	-	-	-0.0003	720
22	0.35	<0	>0	-	-	-0.0011	761
23	0.40	<0	>0	-	-	-0.0003	740
24	0.42	<0	>0	-	-	-0.0003	743
25	0.35	<0	<0	-	-	<u>0.0010</u>	761
26	0.35	>0	-	0.0101	>0	<u>0.0009</u>	761
27	0.35	>0	-	0.0059	>0	<u>0.0008</u>	761
28	0.35	<0	>0	-	-	-0.0003	761
29	0.35	<0	>0	-	-	-0.0003	761
30	0.35	<0	>0	-	-	<u>0.0027</u>	761
31	0.35	>0	<0	-	-	<u>0.0010</u>	761
32	0.35	<0	>0	-	-	-0.0003	761
33	0.35	<0	>0	-	-	-0.0012	761
34	0.35	<0	<0	-	-	<u>0.0010</u>	761
35	0.35	<0	<0	-	-	<u>0.0010</u>	761
36	0.35	<0	>0	-	-	-0.0003	761

Comparative steel

[0070]

Table 5

Steel plate production conditions

Steel No.	Slab reheating			Rough rolling		Finish rolling		Finish rolling	
	Ar ₃ (°C)	Start temperature (°C)	Heating temperature (°C)	Finish temperature (°C)	Cumulative reduction ratio (%)	upper limit restriction* (°C)	Start temperature (°C)	Completion temperature (°C)	Cumulative reduction ratio (%)
1	739	400	1050	950	35	800	750	740	60
2	753	100	1080	960	40	785	750	750	55
3	772	200	1030	940	35	810	720	730	57
4	762	550	1000	910	30	825	820	800	65
5	783	50	1100	980	35	775	730	730	55
6	703	500	1060	950	40	795	770	760	50
7	741	100	1090	940	30	780	710	730	62
8	752	100	1050	910	45	800	780	770	53
9	735	450	980	900	30	835	820	830	60
10	758	500	1050	990	35	800	750	740	55
11	746	540	1000	950	40	825	790	800	65
12	765	20	960	900	30	845	720	710	60
13	730	20	1030	920	40	810	750	730	65
14	739	400	1050	950	35	800	750	740	60
15	739	400	1050	950	35	800	750	740	60
16	735	400	1050	950	35	800	750	740	60

Inventive steel

* Finish rolling upper limit restriction (°C) = $-0.5 \times (\text{slab heating temperature (°C)}) + 1,325$

[0071]

Table 6

Steel production conditions				Accelerated cooling		Tempering treatment	
Steel No.	Plate thickness (mm)	Water volume density (m ³ /m ² /min)	Stop temperature (°C)	Temperature (°C)	Time (min)		
1	60	1.0	100	550	20		
2	70	1.5	200	530	20		
3	65	2.0	300	500	30		
4	50	0.8	500	--	--		
5	80	2.0	20	550	15		
6	50	1.3	100	600	15		
7	75	1.8	200	450	20		
8	80	2.5	350	--	--		
9	70	1.0	100	550	20		
10	70	0.5	300	--	--		
11	60	2.0	150	500	20		
12	70	1.5	450	--	--		
13	80	2.0	150	500	20		
14	60	1.0	100	550	20		
15	60	1.0	100	550	20		
16	60	1.0	100	550	20		

Inventive steel

[0072]

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Table 7

Steel plate production conditions

Steel No.	Slab reheating			Rough rolling		Finish rolling		Finish rolling	
	Ar ₃ (°C)	Start temperature (°C)	Heating temperature (°C)	Finish temperature (°C)	Cumulative reduction ratio (%)	upper limit restriction* (°C)	Start temperature (°C)	Completion temperature (°C)	Cumulative reduction ratio (%)
17	774	400	1050	950	35	800	750	740	60
18	746	400	1050	950	35	800	750	740	60
19	761	400	1050	950	35	800	750	740	60
20	763	400	1050	950	35	800	750	740	60
21	720	400	1050	950	35	800	750	740	60
22	761	400	1050	950	35	800	750	740	60
23	740	400	1050	950	35	800	750	740	60
24	743	400	1050	950	35	800	750	740	60
25	761	400	1050	950	35	800	750	740	60
26	761	400	1050	950	35	800	750	740	60
27	761	400	1050	950	35	800	750	740	60
28	761	400	1050	950	35	800	750	740	60
29	761	400	1050	950	35	800	750	740	60
30	761	400	1050	950	35	800	750	740	60
31	761	400	1050	950	35	800	750	740	60
32	761	400	1050	950	35	800	750	740	60
33	761	400	1050	950	35	800	750	740	60
34	761	400	1050	950	35	800	750	740	60
35	761	400	1050	950	35	800	750	740	60
36	761	400	1050	950	35	800	750	740	60

Comparative steel

* Finish rolling upper limit restriction (°C) = $-0.5 \times (\text{slab heating temperature (°C)} + 1,325)$

[0073]

Table 8

Steel production conditions				Accelerated cooling		Tempering treatment	
Steel No.	Plate thickness (mm)	Water volume density (m ³ /m ² /min)	Stop temperature (°C)	Temperature (°C)	Time (min)		
17	60	1.0	100	550	20		
18	60	1.0	100	550	20		
19	60	1.0	100	550	20		
20	60	1.0	100	550	20		
21	60	1.0	100	550	20		
22	60	1.0	100	550	20		
23	60	1.0	100	550	20		
24	60	1.0	100	550	20		
25	60	1.0	100	550	20		
26	60	1.0	100	550	20		
27	60	1.0	100	550	20		
28	60	1.0	100	550	20		
29	60	1.0	100	550	20		
30	60	1.0	100	550	20		
31	60	1.0	100	550	20		
32	60	1.0	100	550	20		
33	60	1.0	100	550	20		
34	60	1.0	100	550	20		
35	60	1.0	100	550	20		
36	60	1.0	100	550	20		
Comparative steel							

[0074]

Table 9
Steel production conditions using the composition of Steel No. 1 (conventional production conditions)

Steel No.	Ar ₃ (°C)	Slab reheating		Rough rolling		Finish rolling		Finish rolling	
		Start temperature (°C)	Heating temperature (°C)	Finish temperature (°C)	Cumulative reduction ratio (%)	upper limit restriction* (°C)	Start temperature (°C)	Completion temperature (°C)	Cumulative reduction ratio (%)
1	739	400	1050	950	35	800	750	740	60
1A	739	<u>580</u>	1050	950	35	800	750	740	60
1B	739	400	<u>1130</u>	950	35	760	750	740	60
1C	739	400	<u>940</u>	<u>890</u>	30	855	750	740	60
1D	739	400	1050	<u>890</u>	35	800	750	740	60
1E	739	400	1050	950	<u>28</u>	800	750	740	60
1F	739	400	1050	950	35	800	<u>840</u>	<u>850</u>	60
1G	739	400	1050	950	35	800	<u>830</u>	780	60
1H	739	400	1050	950	35	800	750	740	<u>45</u>
1I	739	400	1050	950	35	800	750	740	60

* Finish rolling upper limit restriction (°C) = $-0.5 \times (\text{slab heating temperature (°C)} + 1,325)$

[0075]

Table 10
Steel production conditions using the composition of Steel No. 1 (conventional production conditions)

Steel No.	Plate thickness (mm)	Accelerated cooling		Tempering treatment	
		Water volume density (m ³ /m ² /min)	Stop temperature (°C)	Temperature (°C)	Time (min)
1	60	1.0	100	550	20
Inventive steel					
1A	60	1.0	100	550	20
1B	60	1.0	100	550	20
1C	60	1.0	100	550	20
1D	60	1.0	100	550	20
1E	60	1.0	100	550	20
1F	60	1.0	100	550	20
1G	60	1.0	100	550	20
1H	60	1.0	100	550	20
1I	60	1.0	530	550	20
Comparative steel					

[Evaluation tests]

[0076] The thick steel plate samples prepared using the above method were each subjected to the following evaluation tests.

The tensile properties and Charpy impact properties of the base material were evaluated by taking a test piece from the mid-thickness of steel plate sample - in a rolled longitudinal (L) direction, and then testing this test piece.

The brittle fracture arrestability of the base material was evaluated by conducting a crack test of a full thickness test piece using the temperature gradient ESSO test (compliant with WES 3003) to determine the value of the arrestability indicator $T_{kca=6000}$.

The joint HAZ toughness was investigated by performing one-pass butt welding using electro-gas welding (EGW), and then inserting a notch in the HAZ 1 mm, from the weld line in a 1/2 plate thickness portion. In this case, three samples were subjected to Charpy impact tests at -20°C, and the average absorption energy value was determined. For reference purposes, the Charpy impact properties were also evaluated at -40°C.

[0077] In terms of the mechanical properties of the thick steel plates and the welded joints, a list of the mechanical properties for the steels of the present invention produced under the production conditions shown in Tables 5 and 6 are shown in Table 11, whereas a list of the mechanical properties for the comparative steels produced under the production conditions shown in Tables 7 and 8 are shown in Table 12.

A list of the mechanical properties for thick steel plates and welded joints for comparative steels in which the steel plate was produced using the chemical element composition of "Steel No. 1" of the present invention, but with altered values for various production conditions shown in Tables 9 and 10, are shown in Table 13.

[0078]

Table 11

Mechanical properties of steel plate

Steel No.		1/2 plate thickness		Brittle fracture arrestability		High heat-input welded joint			
		- L direction		- C direction		Welding method	Heat input (kJ/mm)	vE(-20°C) (J)	vE(-40°C) (J)
Plate thickness (mm)	Yield strength (MPa)	Tensile strength (MPa)	vTrs (°C)	T _{kca} = 6000 (°C)					
1	60	470	580	-100	-20	EGW	35	150	50
2	70	460	570	-90	-15	EGW	50	120	50
3	65	420	520	-120	-20	EGW	40	110	60
4	50	500	610	-120	-35	EGW	30	200	100
5	80	460	570	-90	-25	EGW	70	210	110
6	50	530	650	-110	-40	EGW	30	140	50
7	75	490	600	-100	-20	EGW	60	220	100
8	80	440	550	-100	-18	EGW	70	210	120
9	70	500	620	-80	-20	EGW	50	110	50
10	70	400	530	-90	-25	EGW	50	230	140
11	60	510	610	-100	-40	EGW	35	150	50
12	70	410	530	-110	-25	EGW	50	240	100
13	80	420	540	-90	-30	EGW	70	200	110
14	60	480	590	-90	-20	EGW	35	130	50
15	60	470	580	-100	-20	EGW	35	150	50
16	60	470	590	-100	-20	EGW	35	150	50

Inventive steel

[0079]

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Table 12

Mechanical properties of steel plate									
Steel No.	Plate thickness (mm)	1/2 plate thickness - L direction		Brittle fracture arrestability - C direction		High heat-input welded joint			
		Yield strength (MPa)	Tensile strength (MPa)	vTrs (°C)	T _{kca} = 6000 (°C)	Welding method	Heat input (kJ/mm)	vE(-20°C) (J)	vE(-40°C) (J)
17	60	<u>360</u>	<u>460</u>	-100	-30	EGW	35	200	80
18	60	480	580	-70	-20	EGW	35	<u>30</u>	5
19	60	450	550	-100	-30	EGW	35	<u>30</u>	10
20	60	<u>350</u>	<u>460</u>	-100	-25	EGW	35	100	40
21	60	550	670	-70	-15	EGW	35	<u>30</u>	5
22	60	460	550	-100	-30	EGW	35	<u>40</u>	5
23	60	440	550	-100	-25	EGW	35	<u>30</u>	5
24	60	600	700	-70	-10	EGW	35	<u>30</u>	5
25	60	500	610	-80	-10	EGW	35	<u>30</u>	5
26	60	440	550	-100	-10	EGW	35	<u>30</u>	5
27	60	450	580	-70	-20	EGW	35	<u>30</u>	5
28	60	450	550	-30	-5	EGW	35	<u>10</u>	5
29	60	450	550	-50	-10	EGW	35	<u>30</u>	10
30	60	600	700	-40	-10	EGW	35	<u>30</u>	10
31	60	500	600	-30	-10	EGW	35	<u>30</u>	10
32	60	450	550	-30	-10	EGW	35	<u>30</u>	10
33	60	530	620	-30	-5	EGW	35	<u>30</u>	10
34	60	520	630	-30	-10	EGW	35	<u>30</u>	10
35	60	500	640	-30	-5	EGW	35	<u>30</u>	10
36	60	450	550	-30	-10	EGW	35	<u>30</u>	10

Comparative steel

[0080]

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Table 13
Mechanical properties of steel plate produced under altered production conditions and using the composition of steel No. 1

Steel No.	Plate thickness (mm)	1/2 plate thickness - L direction		Brittle fracture arrestability - C direction		Welding method	High heat-input welded joint		
		Yield strength (MPa)	Tensile strength (MPa)	vTrs (°C)	T _{Kca} = 6000 (°C)		Heat input (kJ/mm)	vE(-20°C) (J)	vE(-40°C) (J)
1	60	470	580	-100	-20	EGW	35	150	50
1A	60	460	560	-50	<u>15</u>	EGW	35	150	50
1B	60	470	580	-100	<u>-5</u>	EGW	35	150	50
1C	60	450	560	-100	<u>-5</u>	EGW	35	150	50
1D	60	470	580	-80	<u>-5</u>	EGW	35	150	50
1E	60	470	580	-70	<u>-5</u>	EGW	35	150	50
1F	60	470	580	-70	<u>0</u>	EGW	35	150	50
1G	60	470	580	-80	<u>-5</u>	EGW	35	150	50
1H	60	470	580	-60	<u>10</u>	EGW	35	150	50
1I	60	<u>380</u>	<u>500</u>	-60	<u>0</u>	EGW	35	150	50

Inventive steel

Comparative steel

[Evaluation results]

[0081] The steels No. 1 to 16 shown in Tables 1 and 2 represent steels of the present invention, and it is evident from the results that by optimizing the chemical composition of the steel and maximizing the effects of the low-temperature heating and low-temperature rolling in the TMCP, a steel plate is obtained which, as shown in Table 11, despite being a thick steel plate, exhibits a yield strength in the order of 390 to 460 MPa, a tensile strength in the order of 510 to 570 MPa and a favorable brittle fracture arrestability $T_{kca=6000}$ of less than -10°C , and which despite undergoing high heat-input welding, yields a favorable HAZ toughness at -20°C while suppressing the amount of added Ni to not more than 1%.

[0082] In contrast, the comparative steels No. 17 to 36 shown in Tables 3 and 4 do not have an optimized chemical composition for the steel, whereas the comparative steels 1A to 1I shown in Tables 9 and 10 are produced with steel production conditions that are not optimal, and therefore as is evident from Tables 12 and 13, one of the yield strength, the tensile strength, the $T_{kca=6000}$ value or the HAZ toughness in high heat-input welding deteriorates, and this plurality of required properties is unable to be satisfied in the same manner as the thicker high-strength steel plates of the present invention.

[0083] Steel No. 17 has low values for C and Ceq, and steel No. 20 has a low Mn content, and therefore the hardenability is unsatisfactory. As a result, a deterioration in the yield strength and tensile strength occurs.

Steel No. 18 has a high C content, steel No. 19 has a high Si content, steel No. 21 has a high Mn content, and steel No. 22 has a low B content, and therefore each steel has poor toughness of the high heat-input HAZ.

Steel No. 23 has a low V content, and therefore has a lower strength than steel No. 1 that has the same plate thickness but a lower Ceq value. Further, although the Ceq value of steel No. 23 is higher than that of steel No. 1, the steel No. 23 is unable to satisfy the yield strength in the order of 460 MPa and the tensile strength in the order of 570 MPa which are satisfied by steel No. 1. Moreover, the high heat-input HAZ toughness is also inferior.

Steel No. 24 has a high V content, and therefore has a considerably higher strength than steel No. 11 that has the same plate thickness and Ceq value, but the high heat-input HAZ toughness is inferior.

Steels No. 25, 26, 27, 30, 31, 34 and 35 have the same Ceq and plate thickness values, and the TMCP conditions shown in Tables 7 and 8 are also the same. However, because the effective B amount is within a range from 8 to 10 ppm, the yield strength is from 440 to 600 MPa, and the tensile strength is from 550 to 700 MPa, these results indicate that there is large variation in the strength. Moreover, the high heat-input HAZ toughness is inferior.

[0084] Steel No. 28 has a high P content and steel No. 29 has a high S content, and therefore the base material toughness and the high heat-input HAZ toughness are inferior in both cases.

Steel No. 31 has a low Al content and therefore the O content becomes higher, and steel No. 32 has a very high Al content and therefore contains alumina clusters. In both cases, the amount of coarse harmful oxides increases; thereby, causing a deterioration in the toughness of both of the base material and the high heat-input HAZ.

Steel No. 33 has a low Ti content and steel No. 35 has a low N content, and therefore the production of TiN is insufficient in both steels; thereby, the crystal grains within the base material and the HAZ are unable to be adequately reduced in size. As a result, inferior results for the base material toughness, the arrestability, and the high heat-input HAZ toughness are obtained.

Steel No. 34 has a high Ti content and steel No. 36 has a high N content, and therefore in both steels, TiC embrittlement and solid solution B embrittlement result in inferior base material toughness and inferior high heat-input HAZ toughness.

[0085] Steel No. 1A is produced with a high start temperature for the slab reheating, and steel No. 1B is produced using a high heating temperature. Therefore, in both cases, the γ grains coarsen during heating; thereby, causing a deterioration in the brittle fracture arrestability $T_{kca=6000}$.

In steel No. 1C, the heating temperature is too low, and therefore the solubilization of V carbonitrides is inadequate. As a result, the amount of V carbides which operate for the precipitate strengthening is insufficient; thereby, the base material strength is reduced. As a result, the yield strength and the tensile strength are each 20 MPa lower than that of steel No. 1, and the strength advantage of adding 0.02% of V is unable to be realized. Moreover, because the finishing temperature of the rough rolling is too low, the size of the recrystallized grains is unable to be reduced sufficiently (uniform grain refinement is not conducted sufficiently); thereby, the $T_{kca=6000}$ value is inferior.

In steel No. 1D, the finishing temperature of the rough rolling is too low, whereas in steel No. 1E, the cumulative reduction ratio for the rough rolling is too low, and therefore in both steels, the size of the recrystallized grains is unable to be reduced sufficiently (uniform grain refinement is not conducted sufficiently); thereby, a $T_{kca=6000}$ value is inferior.

In steel No. 1F and steel No. 1G, both of the start temperature and the completion temperature of the finish rolling are too high, and the above formula $\{-0.5 \times (\text{slab heating temperature } (^{\circ}\text{C})) + 1,325\}$ is not satisfied. As a result, the size reduction of the crystal grains of the base material is inadequate; thereby, the $T_{kca=6000}$ value is inferior.

In steel No. 1H, the cumulative reduction ratio for the finish rolling is too low, and therefore the size reduction of the crystal grains of the base material is inadequate; thereby, the $T_{kca=6000}$ value is inferior.

In steel No. 1I, the stop temperature for the accelerated cooling is too high, and therefore the transformation strengthening and crystal grain size reduction within the interior of the plate are unsatisfactory, resulting in inferior tensile strength and

an inferior $T_{kca} = 6000$ value.

[0086] Based on the results of the examples described above, it is clear that the thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of the heat affected zone in high heat-input welding according to the present invention is capable of realizing (1) high strength for thick steel plate including a yield strength in the order of 390 to 460 MPa (namely, a tensile strength in the order of 510 to 570 MPa) for a plate thickness of 50 to 80 mm, (2) favorable brittle fracture arrestability indicated by an arrestability indicator $T_{kca=6000} \leq -10^{\circ}\text{C}$, (3) favorable HAZ toughness in high heat-input welding indicated by $vE (-20^{\circ}\text{C}) \geq 47$ J even when the heat input during welding is ≥ 20 kJ/mm, and (4) reduced production costs due to a reduction in the amount of expensive alloy elements (such as a Ni amount of $\leq 1\%$).

INDUSTRIAL APPLICABILITY

[0087] By using the thicker high-strength steel plate of the present invention in all manner of welded structures, including large ships, an increase in the size of the welded structure, a high level of safety in terms of cracking, improved welding efficiency during construction, and favorable economic viability of the steel plate used as a structural material can all be achieved simultaneously. Accordingly, the thicker high-strength steel plate of the present invention can be applied to the construction of ships including large container ships, as well as other welded structures such as buildings, bridges, tanks and marine structures.

Claims

1. A method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding, the method comprising:

cooling a continuously cast slab which comprises, in terms of mass %, C: 0.05 to 0.12%, Si: not more than 0.3%, Mn: 1 to 2%, P: not more than 0.015%, S: not more than 0.005%, B: 0.0003 to 0.003%, V: 0.01 to 0.15%, Al: 0.001 to 0.1%, Ti: 0.005 to 0.02%, N: 0.002 to 0.01%, and O: not more than 0.004%, with a remainder being iron and unavoidable impurities, to a temperature of $Ar_3-200^{\circ}\text{C}$ or lower, and subsequently reheating said slab to 950 to $1,100^{\circ}\text{C}$;

subjecting said continuously cast slab to rough rolling at 900°C or higher with a cumulative reduction ratio of at least 30%;

subsequently performing finish rolling at 700°C or higher with a cumulative reduction ratio of at least 50% under conditions that both of a finish rolling start temperature and a finish rolling completion temperature are not higher than a temperature represented by a formula: $\{-0.5 \times (\text{slab heating temperature } (^{\circ}\text{C})) + 1,325\} (^{\circ}\text{C})$, thereby forming a rolled plate; and

cooling said rolled plate to 500°C or lower by accelerated cooling to obtain a steel plate,

wherein in said continuously cast slab, a calculated value of an amount of B {effective B amount: $B_{ef}(\%)$ } which is solid-solubilized into an austenite base material prior to transformation is not more than 0%, and a carbon equivalent C_{eq} satisfies a range from 0.32 to 0.42%, and

if an amount of residual oxygen $O_{Ti}(\%)$ that remains after deoxidation by strong deoxidizing elements and is able to undergo deoxidation by Ti that is a weak deoxidizing element is an amount represented by formula (1) shown below, then said effective B amount $B_{ef}(\%)$ is represented by formula (2) shown below, said carbon equivalent $C_{eq}(\%)$ is represented by formula (3) shown below, and said Ar_3 is represented by formula (4) shown below,

$$O_{Ti}(\%) = O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al \cdots (1)$$

{in formula (1), component elements that represent unavoidable impurities are also included within a calculation},

$$B_{ef}(\%) = B - 0.77\{N - 0.29(Ti - 2O_{Ti})\} \cdots (2)$$

{in formula (2), when $O_{Ti} \leq 0$, it is deemed that $O_{Ti} = 0$, when $O_{Ti} > 0$, it is deemed that $Ti - 2O_{Ti} \geq 0.005(\%)$,

and when $N - 0.29(Ti - 2O_{Ti}) \leq 0$ (although when $O_{Ti} \leq 0$, $O_{Ti} = 0$), it is deemed that $N - 0.29(Ti - 2O_{Ti}) = 0$,

$$Ceq (\%) = C + Mn/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15 \quad \cdots (3)$$

$$Ar_3 (^\circ C) = (910 - 310C - 80Mn - 20Cu - 55Ni - 80Mo) \quad \cdots (4).$$

2. A method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding according to claim 1, wherein the method further comprises conducting a tempering heat treatment at a temperature within a range from 350 to 700°C for 5 to 60 minutes after said accelerated cooling.
3. A method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding according to claim 1 or 2, wherein in said continuously cast slab, said S content is within a range from 0.0005 to 0.005%, and said O content is within a range from 0.001 to 0.004%, and said continuously cast slab further comprises, in terms of mass %, either or both of Ca: 0.0003 to 0.004% and Mg: 0.0003 to 0.004%.
4. A method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding according to any one of claims 1 to 3, wherein said continuously cast slab further comprises, in terms of mass %, one or more selected from the group consisting of Ni: 0.01 to 1%, Cu: 0.01 to 1%, Cr: 0.01 to 1%, Mo: 0.01 to 0.5%, and Nb: 0.003 to 0.03%.
5. A method for producing a thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding according to any one of claims 1 to 4, wherein said continuously cast slab further comprises, in terms of mass %, either or both of REM: 0.0003 to 0.02% and Zr: 0.0003 to 0.02%.
6. A thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding, comprising, in terms of mass %, C: 0.05 to 0.12%, Si: not more than 0.3%, Mn: 1 to 2%, P: not more than 0.015%, S: not more than 0.005%, B: 0.0003 to 0.003%, V: 0.01 to 0.15%, Al: 0.001 to 0.1%, Ti: 0.005 to 0.02%, N: 0.002 to 0.01%, and O: not more than 0.004%, with a remainder being iron and unavoidable impurities, wherein if an amount of residual oxygen that remains after deoxidation by strong deoxidizing elements and is able to undergo deoxidation by Ti that is a weak deoxidizing element is an amount represented by formula (5) shown below, then a calculated value of an amount of B {effective B amount: $B_{ef}(\%)$ } which is solid-solubilized into an austenite base material prior to transformation is not more than 0%, a carbon equivalent Ceq represented by formula (7) shown below satisfies a range from 0.32 to 0.42%, a plate thickness is within a range from 50 to 80 mm, a yield strength is in an order of 390 to 460 MPa, a temperature $T_{kca=6000}$ at which brittle fracture arrestability K_{ca} reaches 6,000 N/mm^{1.5} is -10°C or lower, a Charpy impact absorbed energy vE (-20°C), which is an indicator of high heat-input HAZ toughness with 20 kJ/mm or greater of heat input, is at least 47 J,

$$O_{Ti} (\%) = O - 0.4Ca - 0.66Mg - 0.17REM - 0.35Zr - 0.89Al \quad \cdots (5)$$

{in formula (5), component elements that represent unavoidable impurities are also included within a calculation},

$$B_{ef} (\%) = B - 0.77\{N - 0.29(Ti - 2O_{Ti})\} \quad \cdots (6)$$

{in formula (6), when $O_{Ti} \leq 0$, it is deemed that $O_{Ti} = 0$, when $Q_{Ti} > 0$, it is deemed that $Ti - 2O_{Ti} \geq 0.005 (\%)$, and

when $N - 0.29(Ti - 2O_{Ti}) \leq 0$ (although when $O_{Ti} \leq 0$, $O_{Ti} = 0$), it is deemed that $N - 0.29(Ti - 2O_{Ti}) = 0$, and

$$C_{eq} (\%) = C + Mn/6 + (Cr + Mo + V)/5 + (Ni + Cu)/15 \dots (7).$$

7. A thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding according to claim 6,
wherein said S content is within a range from 0.0005 to 0.005%, and said O content is within a range from 0.001 to 0.004%, and
said steel plate further comprises, in terms of mass %, either or both of Ca: 0.0003 to 0.004% and Mg: 0.0003 to 0.004%.
8. A thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding according to claim 6 or 7,
wherein said steel plate further comprises, in terms of mass %, one or more selected from the group consisting of Ni: 0.01 to 1%, Cu: 0.01 to 1%, Cr: 0.01 to 1%, Mo: 0.01 to 0.5%, and Nb: 0.003 to 0.03%.
9. A thicker high-strength steel plate having excellent brittle fracture arrestability and excellent toughness of a heat affected zone in high heat-input welding according to any one of claims 6 to 8,
wherein said steel plate further comprises, in terms of mass %, either or both of REM: 0.0003 to 0.02% and Zr: 0.0003 to 0.02%.

INTERNATIONAL SEARCH REPORT

International application No.

PCT/JP2008/072051

A. CLASSIFICATION OF SUBJECT MATTER C21D8/02 (2006.01) i, C22C38/00 (2006.01) i, C22C38/14 (2006.01) i, C22C38/58 (2006.01) i According to International Patent Classification (IPC) or to both national classification and IPC		
B. FIELDS SEARCHED Minimum documentation searched (classification system followed by classification symbols) C21D8/02, C22C38/00, C22C38/14, C22C38/58 Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched Jitsuyo Shinan Koho 1922-1996 Jitsuyo Shinan Toroku Koho 1996-2009 Kokai Jitsuyo Shinan Koho 1971-2009 Toroku Jitsuyo Shinan Koho 1994-2009 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	JP 2007-284712 A (Nippon Steel Corp.), 01 November, 2007 (01.11.07), (Family: none)	1-9
A	JP 2007-46096 A (Nippon Steel Corp.), 22 February, 2007 (22.02.07), (Family: none)	1-9
A	JP 11-36042 A (Sumitomo Metal Industries, Ltd.), 09 February, 1999 (09.02.99), (Family: none)	1-9
☐ Further documents are listed in the continuation of Box C. ☐ See patent family annex.		
* Special categories of cited documents: "A" document defining the general state of the art which is not considered to be of particular relevance "E" earlier application or patent but published on or after the international filing date "L" document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified) "O" document referring to an oral disclosure, use, exhibition or other means "P" document published prior to the international filing date but later than the priority date claimed "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 "X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone "Y" document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art "&" document member of the same patent family		
Date of the actual completion of the international search 09 March, 2009 (09.03.09)		Date of mailing of the international search report 17 March, 2009 (17.03.09)
Name and mailing address of the ISA/ Japanese Patent Office		Authorized officer Telephone No.

REFERENCES CITED IN THE DESCRIPTION

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- *Journal of The Japan Society of Naval Architects and Ocean Engineers*, 2006, 4-10 [0009]
- *CAMP-ISIJ*, 1993, vol. 6, 684 [0009]