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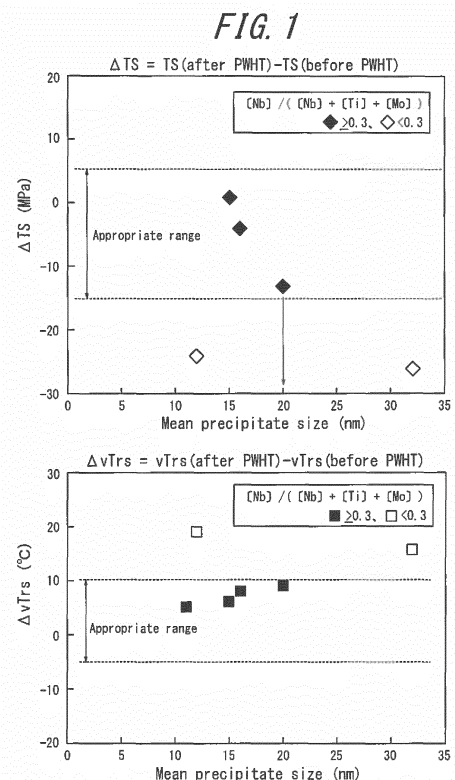
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(54) **STEEL SHEET AND METHOD FOR MANUFACTURING SAME**

(57) A steel plate includes a predetermined chemical composition, in which B is controlled to be less than 0.0003 %, and the balance is Fe and incidental impurities. The steel plate also includes precipitates containing Ti, Nb, and Mo and having a mean particle size of 20 nm or less, the relationship $[Nb]/([Ti] + [Nb] + [Mo]) \geq 0.3$ being satisfied, where [Ti] is the Ti content, [Nb] is the Nb content, and [Mo] is the Mo content, thereby providing a thick, high tensile strength steel plate that is suitable for use in steel structures such as marine structures, ships, pressure vessels, and penstocks, has a yield stress (YS) of 460 MPa or greater, and has excellent low-temperature toughness of the heat-affected zone in a multilayer weld (CTOD property) and excellent strength and toughness after Post Weld Heat Treatment (PWHT property).



Description

TECHNICAL FIELD

[0001] This disclosure relates to a high tensile strength steel plate, and a method for manufacturing the same, used in steel structures such as marine structures and ships, pressure vessels, and penstocks. In particular, this disclosure relates to a thick, high tensile strength steel plate, and a method for manufacturing the same, that not only has a yield stress (YS) of 460 MPa or greater and excellent strength and toughness, but that also has excellent low-temperature toughness in a multilayer weld (CTOD property) and that, after Post Weld Heat Treatment (PWHT), has excellent strength and toughness (PWHT property).

BACKGROUND

[0002] Steel plates used in ships, marine structures, pressure vessels, and the like are subjected to welding to form structures with desired shapes. Therefore, from the perspective of structural safety, these steel plates are not only required to have high strength and excellent toughness but also to have excellent toughness in weld joints (weld metal and Heat-Affected Zone (HAZ)) upon welding.

[0003] The absorbed energy by a Charpy impact test has mainly been used as the basis for evaluating the toughness of steel plates. In recent years, however, a Crack Tip Opening Displacement test (CTOD test) has often been used for greater reliability. This test evaluates the resistance to occurrence of brittle fracture by forming a fatigue precrack in a test piece at the location to be evaluated for toughness, subjecting the test piece to three-point bending, and measuring the amount of the crack opening (plastic deformation volume) immediately before fracture.

[0004] Since a fatigue precrack is used in a CTOD test, an extremely small region is evaluated for toughness. Therefore, if a local brittle zone exists in the steel plate, a low toughness may in some cases be indicated by the CTOD test, even if a good toughness is obtained with a Charpy impact test.

[0005] In a thick steel plate or the like, local brittle zones easily occur in the heat-affected zone, which is subjected to a complicated thermal history due to multilayer welding. In addition, the bond (the boundary between weld metal and base material) and a region in which the bond is formed into a dual phase region by reheating (a region in which coarse grains are formed in the first cycle of welding and which is heated into a ferrite and austenite dual phase region by the subsequent welding pass, hereinafter referred to as a dual phase reheating area) are also prone to becoming local brittle zones.

[0006] Since the bond is exposed to a high temperature just below the melting point at the time of welding, austenite grains are coarsened and are likely to be transformed, by the subsequent cooling, into an upper bainite structure that has a low toughness. Therefore, the toughness of the matrix itself tends to reduce. Furthermore, brittle structures such as a Widmanstätten structure or isolated martensite (also referred to as M-A) easily occur. If such a brittle structure occurs, the toughness of the steel plate tends to reduce even further.

[0007] In order to improve the toughness of the heat-affected zone, for example a technique that incorporates TiN in the steel plate by fine particle distribution to reduce coarsening of austenite grains and to create ferrite nucleation sites has been put to practical use. The bond, however, may be heated to a temperature region at which TiN dissolves. As the demand for low-temperature toughness becomes more stringent, the heating temperature rises, thus making it more difficult to obtain the above-described effect from fine particle distribution of TiN.

[0008] To resolve these problems, JP H03-053367 B2 (PTL 1) and JP S60-184663 A (PTL 2) disclose techniques in which, by dispersing fine grains in steel plates by means of combined addition of rare-earth elements (REM) and Ti, grain growth of austenite is suppressed, thereby improving the toughness of the weld zone.

[0009] PTL 1 and PTL 2 also disclose a technique for dispersing Ti oxides, a technique for combining the capability of ferrite nucleation of BN with oxide dispersion, a technique for adding Ca and a REM to control the morphology of sulfides so as to increase the toughness, and other such techniques.

[0010] JP 3697202 B2 (PTL 3) discloses a technique for dispersing Ti oxides into steel so as to improve the HAZ toughness.

[0011] Furthermore, in the dual phase reheating area, carbon becomes concentrated in a region where reverse transformation to austenite occurs due to dual phase reheating, and brittle bainite structures including isolated martensite are generated during cooling, resulting in reduced toughness of the steel. To avoid this reduction in toughness, techniques have been disclosed to reduce the contents of C and Si in the chemical composition, inhibit the generation of isolated martensite, and improve the toughness, and to ensure the base material strength by adding Cu (for example, JP 3045856 B2 (PTL 4) and JP 4432905 B2 (PTL 5)).

[0012] The technique disclosed in PTL 4 sets the cooling rate after rolling to 0.1 °C/s or less and adopts a method to precipitate Cu particles during this process. Such a technique is problematic, however, in terms of manufacturing stability.

[0013] With the technique disclosed in PTL 5, the N/Al ratio is set between 0.3 and 3.0, thereby suppressing toughness

degradation due to the adverse effect of coarsened AlN and of solute N, but it is easier to control solute N with Ti.

[0014] In thick material for which YS exceeds 460 MPa, Post Weld Heat Treatment (PWHT) may be applied. During PWHT, the base material is also heated at the same time. Therefore, the base material properties need to be maintained even upon PWHT. Conventionally, in order to control a reduction in strength due to heat, an element that forms a precipitate at that temperature is generally added.

CITATION LIST

Patent Literature

[0015]

PTL 1: JP 1-103-053367 B2

PTL 2: JP S60-184663 A

PTL 3: JP 3697202 B2

PTL 4: JP 3045856 B2

PTL 5: JP 4432905 B2

SUMMARY

(Technical Problem)

[0016] The techniques disclosed in PTL 1 and PTL 2 target relatively low strength steel material with a small amount of alloying elements. Unfortunately, these techniques cannot be applied to higher strength, thick material with a large amount of alloying elements, since the HAZ structure does not include ferrite.

[0017] With the technique disclosed in PTL 3, it is difficult to incorporate Ti oxides stably into steel by fine particle distribution.

[0018] Furthermore, ensuring strength by Cu precipitates often leads to a reduction in toughness, causing problems for ensuring the low-temperature toughness of the steel plate. In steel material that uses strengthening by Cu precipitation as recited in PTL 5, Cu particles also grow large during the PWHT, causing the problem of the strength tending to reduce.

[0019] Additionally, in recent years, steel structures such as ships, marine structures, pressure vessels, and penstocks have increased in size, leading to a desire for even higher strength steel.

[0020] The steel material used in these steel structures is often thick material, for example with a plate thickness of 35 mm or greater. Therefore, in order to ensure a strength such that the yield stress is at least 460 MPa grade, a steel chemical composition including a large amount of added alloying elements has become advantageous.

[0021] Sufficient examination has not been made, however, of how to improve toughness of the bonds and the dual phase reheating areas in high tensile strength steel material with a large amount of alloying elements. As for ensuring base metal properties after PWHT, it is difficult to maintain strength and toughness by simply adding precipitating elements as in conventional techniques.

[0022] Therefore, it would be helpful to provide thick, high tensile strength steel, and a method for manufacturing the same, that is suitable for use in steel structures such as marine structures, ships, pressure vessels, and penstocks, has a yield stress (YS) of 460 MPa or greater (satisfying this level of YS is referred to as high tensile strength in this disclosure), and has excellent low-temperature toughness of the heat-affected zone in a multilayer weld (CTOD property) and excellent strength and toughness after Post Weld Heat Treatment (PWHT property).

(Solution to Problem)

[0023] To that end, we engaged in intensive studies and made the following discoveries.

(a) Since the CTOD property is evaluated with a test piece having the entire thickness of the steel plate, the central segregation area where components are concentrated becomes the origin of fracture.

Consequently, in order to improve the CTOD property of the heat-affected zone, it is effective to control elements that easily concentrate as central segregation of the steel plate to a proper amount, thereby suppressing the hardening of the central segregation area. Furthermore, at the center of the slab, which is the last portion to solidify when the molten steel solidifies, the concentration of C, Mn, P, Ni, and Nb is higher than the concentration of other elements. Hence, it is effective to control the added amounts of these elements on the basis of the central segregation area hardness index, thereby reducing the hardness at the central segregation area.

(b) In order to improve the toughness of the heat-affected zone, it is effective to use TiN efficiently to suppress

coarsening of the austenite grains in the vicinity of the weld bond. In particular, by controlling the Ti/N ratio to an appropriate level, TiN can be uniformly and finely dispersed in the steel.

(c) It is effective to use crystallization of the Ca compound (CaS), which is added for morphological control of sulfides, to improve the toughness of the heat-affected zone.

Since CaS crystallizes at a low temperature as compared to oxides, CaS can be distributed uniformly as fine particles. Furthermore, by controlling the amount of Ca added and the amount of dissolved oxygen in the molten steel at the time of addition to be within appropriate ranges, solute S can also be guaranteed after CaS crystallization. Hence, MnS precipitates on the surface of the CaS to form a complex sulfide. Since a Mn dilute zone is formed around the MnS, ferrite transformation is further promoted.

(d) In addition to Nb, which forms precipitates, adding Ti and Mo as essential elements allows formation, at the stage of manufacturing a steel plate, of fine precipitates of a complex carbonitride of Mo, Ti, and Nb, which does not coarsen even when subjected to heating by PWHT (performed in a range of approximately 550 °C to 650 °C for 2 h to 4 h).

[0024] Conventionally, the reduction in strength after PWHT in a steel plate exceeding YS: 460 MPa grade is significant, but in a developmental steel plate, we discovered that by the stable existence of fine Mo, Ti, Nb complex precipitates (carbides, nitrides, or carbonitrides), strengthening by precipitation can be maintained, and the reduction in strength of the steel plate can be suppressed. Furthermore, due to the existence of fine Mo, Ti, Nb complex precipitates, the toughness of the steel plate can also be maintained.

[0025] This disclosure was completed based on the above discoveries, and the primary features thereof are as follows.

1. A steel plate, comprising:

a chemical composition including, by mass%,

C: 0.020 % to 0.090 %,
Si: 0.01 % to 0.35 %,
Mn: 1.40 % to 2.00 %,
P: 0.008 % or less,
S: 0.0035 % or less,
Al: 0.010 % to 0.060 %,
Ni: 0.40 % to 2.00 %,
Mo: 0.05 % to 0.50 %,
Nb: 0.005 % to 0.040 %,
Ti: 0.005 % to 0.025 %,
N: 0.0020 % to 0.0050 %,
Ca: 0.0005 % to 0.0050 %, and
O: 0.0035 % or less,

Ceq specified by formula (1) below being in a range of 0.420 % to 0.520 %, formulas (2), (3), and (4) below being satisfied, B being controlled to be less than 0.0003 %, and a balance being Fe and incidental impurities; and precipitates including Ti, Nb, and Mo and having a mean particle size of 20 nm or less, wherein $[Nb]/([Ti] + [Nb] + [Mo]) \geq 0.3$, where [Ti] is the Ti content, [Nb] is the Nb content, and [Mo] is the Mo content:

$$Ceq = [C] + [Mn]/6 + ([Cu] + [Ni])/15 + ([Cr] + [Mo] + [V])/5$$

(1)

$$1.5 \leq [Ti]/[N] \leq 4.0 \quad (2)$$

$$0 < \{[Ca] - (0.18 + 130 \times [Ca]) \times [O]\} / 1.25 / [S] < 1.5 \quad (3)$$

$$5.5[C]^{(4/3)} + 15[P] + 0.90[Mn] + 0.12[Ni] + 7.9[Nb]^{(1/2)} + 0.53[Mo] \leq 3.70 \quad (4)$$

where [M] is the content of element M by mass%.

2. The steel plate of 1., wherein the chemical composition further includes, by mass%, at least one selected from the group consisting of Cu: less than 0.7 %, Cr: 0.1 % to 1.0 %, and V: 0.005 % to 0.05 %.

3. The steel plate of 1. or 2., wherein the chemical composition further includes, by mass%, at least one selected from the group consisting of Mg: 0.0002 % to 0.0050 % and a REM: 0.0010 % to 0.0200 %.

4. A method for manufacturing a steel plate, the method comprising:

heating steel having the chemical composition of any one of 1. to 3. to 950 °C to 1150 °C;

subsequently subjecting the steel to hot rolling at a cumulative rolling reduction of 30 % or higher in a temperature range of 900 °C or higher and a cumulative rolling reduction of 30 % to 70 % in a temperature range of less than 900 °C; and

subsequently cooling the steel at least to 500 °C with a cooling rate of 1.0 °C/s or higher.

5. The method of 4., further comprising tempering at 450 °C to 650 °C after the cooling.

(Advantageous Effect)

[0026] According to this disclosure, a thick, high tensile strength steel plate, and a method for manufacturing the same, that is suitable for use in large steel structures such as marine structures, has a yield stress (YS) of 460 MPa or greater, and has an excellent CTOD property in a multilayer weld and an excellent PWHT property can be obtained. Therefore, this disclosure is extremely useful in industrial terms.

BRIEF DESCRIPTION OF THE DRAWINGS

[0027] In the accompanying drawings:

FIG. 1 illustrates the relationship between change in strength/toughness and the precipitate size composition during PWHT; and

FIG. 2 illustrates TEM replica observation and EDX analysis results for precipitates in a steel plate.

DETAILED DESCRIPTION

[0028] Our methods and products will be described in detail below.

[0029] First, reasons why the chemical composition (steel composition) of the steel plate (also referred to as thick material) has been restricted to the aforementioned ranges will be described in detail for each element. The % representations below indicating the chemical composition of the steel plate are in mass% unless stated otherwise.

C: 0.020 % to 0.090 %

[0030] C is an essential element for ensuring the strength of a high tensile strength steel plate. When the C content is less than 0.020 %, quench hardenability is degraded, and it becomes necessary to add a large amount of quench hardenability-improving elements, such as Cu, Ni, Cr, or Mo, in order to ensure strength, resulting in a rise in costs. Conversely, when the amount of C added exceeds 0.090 %, the toughness of the weld zone degrades. Therefore, the C content is set in the range of 0.020 % to 0.090 %, and preferably 0.020 % to 0.080 %.

Si: 0.01 % to 0.35 %

[0031] Si is added as a deoxidizing element and in order to obtain the steel plate strength. To obtain these effects, the Si content needs to be at least 0.01 %. On the other hand, a large Si content exceeding 0.35 % leads to deterioration in weldability and toughness of the weld joint. Therefore, the Si content needs to be set in the range of 0.01 % to 0.35 %, preferably 0.01 % to 0.23 %.

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Mn: 1.40 % to 2.00 %

5 **[0032]** in order to ensure the steel plate strength and the weld joint strength, the Mn content needs to be at least 1.40 %. Conversely, when the amount of Mn content exceeds 2.00 %, the weldability deteriorates, quench hardenability becomes excessive, and the toughness of the steel plate and the toughness of the weld joint deteriorate. Therefore, the Mn content is set in the range of 1.40 % to 2.00 %, more preferable 1.40 % to 1.95 %.

P: 0.008 % or less

10 **[0033]** P, which is an impurity element, degrades the toughness of the steel plate and the toughness of the weld zone. In particular, when the P content in the weld zone exceeds 0.008 %, the CTOD property markedly degrades. Therefore, the P content is set to 0.008 % or less, preferably 0.006 % or less. The P content is preferably as small as possible, but considering factors such as refining cost, the lower limit may be approximately 0.002 %.

15 S: 0.0035 % or less

20 **[0034]** S is an impurity element, and when the content thereof exceeds 0.0035 %, the toughness of the steel plate and the weld zone deteriorate. Therefore, the S content is set to 0.0035 % or less, preferably 0.0030 % or less. The S content is preferably as small as possible, but considering factors such as refining cost, the lower limit may be approximately 0.0004 %.

A1: 0.010 % to 0.060 %

25 **[0035]** A1 is an element to be added in order to deoxidize molten steel, and the A1 content needs to be set to 0.010 % or more. When the A1 content exceeds 0.060 %, however, the toughness of the steel plate and the toughness of the weld zone are degraded, and A1 is mixed into the weld metal by dilution due to welding, which degrades toughness. Therefore, the A1 content is limited to 0.060 % or less, preferably 0.017 % to 0.055 %. In this disclosure, the A1 content is specified in terms of acid-soluble A1 (also referred to as "Sol. A1" or the like).

30 Ni: 0.40 % to 2.00 %

35 **[0036]** Ni is an element useful for improving the strength and toughness of the steel plate and is also useful for improving the CTOD property of the weld zone. In order to obtain these effects, the added content of Ni needs to be 0.40 % or more. Ni is an expensive element, however, and excessive addition thereof also increases the likelihood of damage to the surface of the slab at the time of casting. Therefore, the upper limit of the Ni content is set to 2.00 %.

Mo: 0.05 % to 0.50 %

40 **[0037]** Mo is an element that fulfills an important function in this disclosure, and adding an appropriate amount thereof is useful for increasing the strength of the steel plate. This effect is obtained by improving the quench hardenability and the temper softening resistance at the time of tempering. Furthermore, Mo maintains the complex precipitate formed by Mo, Ti, and Nb in a fine state, thereby strengthening thick material and controlling a reduction in toughness. In order to obtain these effects, the Mo content needs to be at least 0.05 %. An excessive Mo content, however, adversely affects the toughness of the thick material. Hence, the upper limit on the Mo content is set to 0.50 %. The Mo content is more
45 preferably in the range of 0.08 % to 0.40 %, and even more preferably in the range of 0.16 % to 0.30 %.

Nb: 0.005 % to 0.040 %

50 **[0038]** Nb forms an unrecrystallized zone of austenite in the low temperature region. Therefore, by performing rolling in such a temperature region, the structure of the steel plate can be refined and the toughness of the steel plate can be increased. Furthermore, Nb has the effect of improving the quench hardenability, and by being added in combination with Mo and Ti, Nb has the effect of improving the softening resistance at the time of tempering. Nb is also a useful element for improving the strength of the steel plate. In order to obtain these effects, the Nb content needs to be at least 0.005 %. When the Nb content exceeds 0.040 %, however, the toughness deteriorates. Hence, the upper limit on the
55 Nb content is set to 0.040 %, preferably 0.035 %.

Ti: 0.005 % to 0.025 %

[0039] Ti is precipitated as TiN when molten steel solidifies, which suppresses coarsening of austenite in the weld zone, thus contributing to improvement in the toughness of the weld zone. Furthermore, by being added in combination with Mo and Nb, Ti has the effect of improving the softening resistance at the time of tempering. When the Ti content is less than 0.005 %, however, such an effect is small. On the other hand, when the Ti content exceeds 0.025 %, TiN coarsens, and it is not possible to obtain the effect of improving the toughness of the steel plate and the weld zone. Therefore, the Ti content is set in the range of 0.005 % to 0.025 %.

N: 0.0020 % to 0.0050 %

[0040] N reacts with Ti and Al to form precipitates. Crystal grains are thereby refined, and the toughness of the steel plate is improved. Furthermore, N is a necessary element for forming TiN which suppresses coarsening of the structure of the weld zone. In order to obtain such effects, the N content needs to be set to 0.0020 % or more. On the other hand, when the N content exceeds 0.0050 %, solute N markedly degrades the toughness of the steel plate and the weld zone and leads to a deterioration in strength due to a reduction in solute Nb caused by generation of complex precipitates of Ti and Nb. Therefore, the upper limit on the N content is set to 0.0050 %.

Ca: 0.0005 % to 0.0050 %

[0041] Ca is an element that improves toughness by fixing S. In order to obtain this effect, the Ca content needs to be at least 0.0005 %. On the other hand, Ca content exceeding 0.0050 % causes saturation of the effect. Therefore, Ca is added in the range of 0.0005 % to 0.0050 %.

O: 0.0035 % or less

[0042] If the O content exceeds 0.0035 %, the toughness of the steel plate deteriorates. Hence, the O content is set to 0.0035 % or less, preferably 0.0028 % or less. The O content is preferably as small as possible, but considering factors such as refining cost, the lower limit may be approximately 0.0010 %.

Ceq: 0.420 % to 0.520 %

[0043] When Ceq specified by the formula below is less than 0.420 %, a thick material strength of 460 MPa grade cannot be obtained. On the other hand, if Ceq exceeds 0.520 %, the weldability of the thick material and the toughness of the weld zone deteriorate. Hence, Ceq is set to 0.520 % or less. Ceq is preferably in the range of 0.440 % to 0.520 %. In the formula below, [M] represents the content (mass%) of element "M" in steel. Elements that are not included are calculated as zero.

$$Ceq = [C] + [Mn]/6 + ([Cu] + [Ni])/15 + ([Cr] + [Mo] + [V])/5$$

[Ti]/[N]: 1.5 to 4.0

[0044] When the value of [Ti]/[N] is less than 1.5, the amount of TiN formed decreases, and solute N not forming TiN degrades the toughness of the weld zone. On the other hand, when the value of [Ti]/[N] exceeds 4.0, TiN is coarsened to degrade the toughness of the weld zone. Therefore, the value of [Ti]/[N] is set in the range of 1.5 to 4.0, and preferably 1.8 to 3.5.

$$0 < \{[Ca] - (0.18 + 130 \times [Ca]) \times [O]\} / 1.25 / [S] < 1.5$$

[0045] The expression $\{[Ca] - (0.18 + 130 \times [Ca]) \times [O]\} / 1.25 / [S]$ is a value representing the Atomic Concentration Ratio (ACR) of Ca and S, which are effective for sulfide morphological control, and can be adjusted by controlling the amount of Ca added and the amount of dissolved oxygen in the molten steel at the time of addition to be within appropriate ranges. The sulfide morphology can be estimated by this ACR value, but in this disclosure, the ACR value is specified as an index in order to finely disperse CaS which does not dissolve even at high temperatures and which acts as nuclei for ferrite transformation.

[0046] When the ACR value is 0 or less, CaS is not crystallized. Consequently, S is precipitated in the form of MnS only, which easily dissolves in the heat-affected zone, making it impossible to obtain ferrite product nuclei. Furthermore, the MnS precipitated alone is elongated during rolling and causes degradation in the toughness of the steel plate. Therefore, in this disclosure, the ACR value needs to exceed zero.

[0047] On the other hand, when the ACR value is 1.5 or more, the proportion of oxides among Ca-based inclusions increases, and the proportion of sulfides that function as nucleation sites decreases, making it impossible to obtain the effect of improving toughness. Therefore, in this disclosure, the ACR value needs to be less than 1.5.

[0048] Accordingly, by controlling the ACR value to exceed 0 and be less than 1.5, complex sulfides mainly composed of CaS can be formed effectively and caused to function effectively as ferrite nucleation sites. The ACR value is preferably in the range of 0.15 to 1.30, and more preferably in the range of 0.20 to 1.00.

$$5.5[C]^{(4/3)} + 15[P] + 0.90[Mn] + 0.12[Ni] + 7.9[Nb]^{(1/2)} + 0.53[Mo] \leq 3.70$$

[0049] The value of the left-hand side of the formula above ($5.5[C]^{(4/3)} + 15[P] + 0.90[Mn] + 0.12[Ni] + 7.9[Nb]^{(1/2)} + 0.53[Mo]$) is the hardness index of the central segregation area formed by components that are likely to be concentrated in the central segregation area and is referred to below as the Ceq^* value.

[0050] Since the CTOD test is carried out over the entire thickness of a steel plate, test pieces include central segregation. In the case where the composition concentration in the central segregation is significant, a hardened region occurs in the heat-affected zone. Therefore, a good value cannot be obtained in the CTOD test.

[0051] Therefore, in this disclosure, by controlling the Ceq^* value to be in an appropriate range, an excessive increase in hardness in the central segregation area can be suppressed, and an excellent CTOD property can be obtained even in the weld zone of a thick steel plate.

[0052] The appropriate range of the Ceq^* value has been experimentally obtained. When the Ceq^* value exceeds 3.70, the CTOD property is degraded. Therefore, the Ceq^* value is set to be 3.70 or less, and preferably 3.50 or less. No lower limit is placed on the Ceq^* value, but considering factors such as productivity, preferably the lower limit is approximately 2.2.

[0053] In this disclosure, in addition to the above-described essential elements, at least one selected from the group consisting of Cu: less than 0.7 %, Cr: 0.1 % to 1.0 %, and V: 0.005 % to 0.05 % may be added to increase quench hardenability.

Cu: less than 0.7 %

[0054] Adding Cu improves the strength of the steel plate. If the content of Cu exceeds 0.7 %, however, the hot ductility deteriorates. Therefore, the Cu content is limited to 0.7 % or less. The Cu content is preferably 0.1 % to 0.6 %.

Cr: 0.1 % to 1.0 %

[0055] Cr is an element effective in increasing the strength of the steel plate. In order to obtain this effect, the Cr content is set to 0.1 % or more. An excessively high Cr content, however, adversely affects toughness. Therefore, the Cr content is preferably 0.1 % to 1.0 %, and more preferably 0.2 % to 0.8 %, when included.

V: 0.005 % to 0.05 %

[0056] V is an element that is effective in improving the strength and toughness of the steel plate at a content of 0.005 % or more. Setting the V content to exceed 0.05 %, however, leads to deterioration of toughness. Therefore, the V content is preferably 0.005 % to 0.05 % when included.

[0057] In this disclosure, in addition to the above-described essential elements, at least one selected from the group consisting of Mg: 0.0002 % to 0.0050 % and a REM: 0.0010 % to 0.0200 % may be added to increase the HAZ toughness.

[0058] Mg and REM are elements having the effect of improving the toughness of steel via the dispersion of oxides. In order to bring out such an effect, the Mg content is set to 0.0002 % or more, and the REM content to 0.0010 % or more. On the other hand, Mg content exceeding 0.0050 % and REM content exceeding 0.0200 % merely causes saturation of this effect. Accordingly, when adding these elements, the content is preferably set in the above-mentioned ranges. The Mg content is more preferably 0.0005 % to 0.0020 %, and the REM content is more preferably 0.0020 % to 0.0150 %.

[0059] Components other than the above-mentioned chemical composition are Fe and incidental impurities. In particular, when the steel plate is cooled from the austenite region, B exists in a segregated manner at austenite grain

boundaries, suppresses ferrite transformation, and generates bainite structures that include a large amount of M-A. Hence, B has the disadvantage of making the structure brittle particularly in the heat-affected zone. Accordingly, in this disclosure, the content of B in the steel plate needs to be kept below 0.0003 %.

[0060] In the steel plate, the size of precipitates needs to change little before and after PWHT, and the strength and toughness of the steel plate need to be maintained. FIG. 1 illustrates the relationship between (i) the precipitate size and precipitate composition after PWHT and (ii) the change in strength and toughness before and after PWHT (ΔTS , $\Delta vTrs$), and FIG. 2 illustrates TEM replica observation and EDX analysis results for precipitates in steel.

[0061] From the perspective of stability, the change in strength and in toughness before and after PWHT respectively need to satisfy the following ranges: ΔTS of 5 MPa to -15 MPa, and $\Delta vTrs$ of 10 °C to -5 °C. In order to satisfy these ranges, it is clear from FIG. 1 that the mean size of the precipitates needs to be kept to 20 nm or less, and that the Ti content (represented as [Ti]), the Nb content (represented as [Nb]), and the Mo content (represented as [Mo]) in the precipitates need to satisfy the relationship $[Nb]/([Ti] + [Nb] + [Mo]) > 0.3$.

[0062] As is clear from Table 1, which lists the EDX analysis results for precipitates in steel in FIG. 2, the above-mentioned precipitates are Ti, Nb, and Mo precipitates. Since it suffices for the relationship $[Nb]/([Ti] + [Nb] + [Mo]) \geq 0.3$ to be satisfied, the precipitates need to be at least Nb precipitates, whereas Ti and Mo precipitates may be included in any range that satisfies this relationship. In this disclosure, having an excellent PWHT property refers to ΔTS being in the range of 5 MPa to -15 MPa, and $\Delta vTrs$ being in the range of 10 °C to -5 °C. The precipitates in this disclosure (complex precipitates) are Mo, Ti, and Nb precipitates. Specifically, these precipitates are carbides, nitrides, or carbonitrides of Mo, Ti, and Nb, or a mixture thereof.

[0063] [Table 1]

Table 1

Measuring Point No.	Ti (mass%)	Nb (mass%)	Mo (mass%)	$[Nb]/([Ti]+[Nb]+[Mo])$
5	15.51	82.1	2.39	0.821
6	27.63	72.37	-	0.724
7	20.52	79.48	-	0.795
8	20.92	65.6	13.48	0.656

[Method of Determining Precipitate Particle Size]

[0064] The method of determining the precipitate particle size in this disclosure is in accordance with the TEM replica method. Specifically, after appropriately collecting the precipitate zone of Ti, Nb, and Mo carbides from the steel, the average equivalent circular diameter was determined using image processing on an observation made at 100,000X over four fields of view and taken as the particle size of the precipitates. In this disclosure, the lower limit on the measurement target for precipitate size was set to 2 nm. The reason is that precipitates with a precipitate size smaller than 2 nm are hard to measure.

[0065] Next, our method for manufacturing steel will be described. Our steel is preferably manufactured with the method of manufacturing described below.

[0066] Molten steel adjusted to have a chemical composition within the above-described ranges is prepared by steel-making with an ordinary method using a converter, an electric heating furnace, a vacuum melting furnace, or the like. Next, after forming the molten steel into a slab by continuous casting, the slab is hot rolled to a desired plate thickness. The result is then cooled, and as necessary, tempered. During the hot rolling of this disclosure, the slab heating temperature and rolling reduction are prescribed.

[0067] In this disclosure, unless otherwise noted, the temperature conditions on the steel plate are prescribed by the temperature at the central portion in the plate thickness direction of the steel plate. The temperature at the central portion in the plate thickness direction is determined from the plate thickness, the surface temperature, the cooling conditions, and the like by simulation calculation or the like. For example, the temperature at the central portion in the plate thickness direction may be determined by calculating the temperature distribution in the plate thickness direction using the finite difference method.

Slab reheating temperature: 950 °C to 1150 °C

[0068] The slab reheating temperature is set to 950 °C or higher in order to remove casting defects in the slab reliably with hot rolling. If the slab is reheated to a temperature exceeding 1150 °C, however, the austenite crystallized grains coarsen, causing the toughness of the steel plate to degrade. Hence, the upper limit on the slab reheating temperature

is set to 1150 °C.

Cumulative rolling reduction of hot rolling in a temperature range of 900 °C or higher: 30 % or higher

[0069] In order to render casting defects harmless by pressure bonding and to provide austenite grains with a fine microstructure by recrystallization, the cumulative rolling reduction of hot rolling in a temperature range of 900 °C or higher is set to 30 % or higher. The reason is that if the cumulative rolling reduction is less than 30 %, coarse grains formed during reheating remain and adversely affect the toughness of the steel plate. No upper limit is placed on the cumulative rolling reduction of hot rolling in a temperature range of 900 °C or higher, but in industrial terms, the upper limit is approximately 95 %.

Cumulative rolling reduction of hot rolling in a temperature range of less than 900 °C: 30 % to 70 %

[0070] In this temperature range, the rolled austenite grains do not sufficiently recrystallize. Therefore, austenite grains that remain flattened after rolling constitute a state of high internal distortion that includes numerous defects, such as an internal distortion zone. These austenite grains act as the driving force for ferrite transformation and encourage phase transformation.

[0071] If the cumulative rolling reduction is less than 30 %, however, accumulation of internal energy in the distortion zone is insufficient, making it difficult for ferrite transformation to occur and reducing the toughness of the steel plate. Conversely, if the cumulative rolling reduction exceeds 70 %, generation of polygonal ferrite is encouraged, making high strength and high toughness incompatible. Accordingly, in this disclosure, the cumulative rolling reduction of hot rolling in a temperature range of less than 900 °C is set in the range of 30 % to 70 %.

Cooling rate at least to 500 °C: 1.0 °C/s or higher

[0072] After hot rolling, accelerated cooling is performed at least to 500 °C at a cooling rate of 1.0 °C/s or higher. The reason is that if the cooling rate is less than 1.0 °C/s, sufficient strength of the steel plate is not obtained. Furthermore, if cooling is stopped at a higher temperature than 500 °C, the proportion of ferrite and pearlite structure increases, making high strength and high toughness of thick material incompatible. While no lower limit is placed on the stop temperature of accelerated cooling, the steel is preferably cooled to room temperature.

Tempering temperature: 450 °C to 650 °C

[0073] In this disclosure, when tempering the steel, a sufficient tempering effect is not obtained if the tempering temperature is less than 450 °C. On the other hand, when tempering at a temperature exceeding 650 °C, the precipitates may coarsen, lowering the toughness and the strength of the steel. Hence, a temperature exceeding 650 °C is not preferable.

[0074] The tempering in this disclosure more preferably uses induction heating, which suppresses coarsening of carbides during tempering. In this case, the tempering is preferably performed so that the temperature at the center of the steel plate calculated by a simulation using the finite difference method or the like is from 450 °C to 650 °C.

[0075] In this disclosure, when desired properties of the steel plate have been obtained, as for a TMCP steel plate or the like, the above-described tempering need not be performed.

[0076] Thick material in this disclosure has a thickness of 15 mm or greater. Accordingly, the thickness in this disclosure refers to a steel thickness of 15 mm or greater, but the effects of this disclosure are best obtained when the steel thickness is in a range of 40 mm to 100 mm. Manufacturing conditions other than the above-described manufacturing conditions on thick, high tensile strength steel may be in accordance with conventional methods.

[0077] In the thick, high tensile strength steel of this disclosure, coarsening of austenite grains in the heat-affected zone is suppressed while finely dispersing nuclei for ferrite transformation that do not dissolve even at high temperatures, thereby refining the structure of the heat-affected zone. High toughness is thus obtained. Also in an area reheated to a dual phase by the thermal cycle at the time of multilayer welding, the structure of the heat-affected zone due to initial welding is refined. Therefore, in the dual phase reheating area, the toughness of the non-transformed area can be improved, the austenite grains that undergo retransformation can be refined, and the extent of reduction in toughness can be reduced. Additionally, by generating fine complex precipitates of Ti, Nb, and Mo, a thick, high tensile strength steel plate is provided with an excellent CTOD property and PWHT property.

EXAMPLES

Examples according to this disclosure are now described.

[0078] Using continuously-cast slabs with steel codes A to Z having the chemical composition listed in Table 2 as raw material, the hot rolling and heating listed in Table 3 were performed to produce thick steel plates with a thickness of 50 mm to 150 mm. The steel plates were evaluated by a tensile test in which JIS No. 4 test pieces were collected from the 1/2 position along the thickness of the steel plates, so that the longitudinal direction of each test piece was perpendicular to the rolling direction of the steel plate. The yield stress (YS) and tensile strength (TS) were then measured.

[0079] A Charpy impact test was also performed by collecting JIS No. 4 V-notch test pieces measuring 2 mm from the 1/2 position along the thickness of the steel plates, so that the longitudinal direction of each test piece was perpendicular to the rolling direction of the steel plate. The absorbed energy $vE_{-40}^{\circ C}$ at $-40^{\circ C}$ was then measured. In these Examples, the base metal properties were evaluated as being good when all of the following relationships were satisfied: $YS \geq 460$ MPa, $TS \geq 570$ MPa, and $vE_{-40}^{\circ C} \geq 200$ J.

[0080] The toughness of the weld zone was evaluated by producing a multilayer fill weld joint, using a single bevel groove, by submerged arc welding having a welding heat input of 35 kJ/cm and measuring the absorbed energy $vE_{-40}^{\circ C}$ at $-40^{\circ C}$ with a Charpy impact test, using the weld bond on the straight side at the 1/2 position along the thickness of the steel plates as the notch position for the test. The toughness of the weld zone was determined to be good when the mean for three tests satisfied the relationship $vE_{-40}^{\circ C} \geq 150$ J.

[0081] Using the weld bond at the straight side as the notch position for the CTOD test pieces, the CTOD value at $-10^{\circ C}$, i.e. $\delta-10^{\circ C}$, was measured. The CTOD property of the weld joint was determined to be good when the minimum of the CTOD value ($\delta-10^{\circ C}$) over three tests was 0.5 mm or greater.

[0082] Furthermore, the precipitate zone in the steel was collected by the TEM replica method, and the average equivalent circular diameter was determined using image processing on an observation made at 100,000X over four fields of view. Precipitates with a particle size near the mean were selected by EDX, the precipitate composition thereof was determined, and $[Nb]/([Ti]+[Nb]+[Mo])$ was determined as the mean of three precipitates.

[0083] Regarding the change in base metal properties after PWHT, ΔTS ($TS(\text{after PWHT}) - TS(\text{before PWHT})$) and $\Delta vTrs$ ($vTrs(\text{after PWHT}) - vTrs(\text{before PWHT})$) were determined. PWHT was performed by maintaining the steel at $580^{\circ C}$ for four hours, with temperature raising and lowering rates of $70^{\circ C/h}$.

[0084] Table 3 illustrates the hot rolling conditions, heat treatment conditions, base metal properties, the results of the above-described Charpy impact test and CTOD test on the weld zone, the precipitate size/composition, and the change in base metal properties after PWHT.

[Table 2]

Steel No	C	Si	Mn	P	S	Al	Ni	Mo	Nb	Tr	B	N	Ca	Cu	Cr	V	O	Mg	REM	Ceq	[Ti]N	Formels ACR	Formels Ceq ^a
A	0.054	0.09	1.75	0.003	0.0026	0.034	0.63	0.41	0.021	0.012	0.0001	0.0037	0.0035	0.28	-	-	0.0012	-	-	0.488	3.24	0.84	3.17
B	0.065	0.23	1.52	0.005	0.00309	0.030	1.45	0.26	0.026	0.009	0.0001	0.0030	0.0015	-	-	-	0.0026	-	-	0.468	3.00	0.47	3.17
C	0.072	0.20	1.54	0.004	0.0017	0.025	1.29	0.16	0.009	0.012	0.0001	0.0036	0.0021	-	0.12	0.012	0.0012	0.0008	-	0.473	3.33	0.73	2.60
D	0.076	0.07	1.53	0.003	0.0020	0.024	0.51	0.40	0.016	0.012	0.0002	0.0034	0.0019	0.14	-	-	0.0011	-	0.0010	0.454	3.53	0.57	2.88
E	0.078	0.09	1.45	0.005	0.0014	0.017	1.31	0.24	0.028	0.010	0.0002	0.0038	0.0023	-	0.14	-	0.0022	-	-	0.483	2.63	0.71	3.17
F	0.105	0.10	1.60	0.005	0.0014	0.035	0.52	0.40	0.016	0.010	0.0001	0.0031	0.0020	0.24	-	0.012	0.0024	-	-	0.505	3.23	0.54	3.06
G	0.078	0.42	1.65	0.004	0.0016	0.027	1.36	0.22	0.029	0.009	0.0002	0.0032	0.0012	0.33	-	-	0.0015	-	0.510	2.81	0.35	3.35	
H	0.067	0.08	1.18	0.004	0.0026	0.013	1.14	0.19	0.012	0.008	0.0001	0.0025	0.0016	-	-	-	0.0017	-	0.377	3.20	0.29	2.37	
I	0.064	0.14	2.17	0.005	0.0022	0.018	0.77	0.20	0.017	0.009	0.0001	0.0027	0.0014	-	0.36	-	0.0016	0.0007	-	0.623	3.33	0.70	3.58
J	0.060	0.20	1.53	0.014	0.0011	0.039	0.58	0.42	0.026	0.012	0.0002	0.0034	0.0012	0.18	-	-	0.0012	-	-	0.450	3.53	0.58	3.28
K	0.064	0.19	1.47	0.006	0.0007	0.029	1.05	0.24	0.016	0.011	0.0001	0.0032	0.0011	-	0.24	-	0.0018	-	-	0.475	3.44	0.59	2.81
L	0.076	0.10	1.69	0.004	0.0013	0.024	1.45	0.27	0.030	0.013	0.0011	0.0038	0.0015	0.10	-	-	0.0020	-	-	0.528	3.42	0.46	3.45
M	0.052	0.06	1.54	0.006	0.0016	0.018	0.84	0.35	0.029	0.010	0.0001	0.0034	0.0023	1.24	0.82	-	0.0015	-	-	0.600	2.94	0.79	3.21
N	0.041	0.16	1.46	0.005	0.0017	0.039	1.30	0.26	0.056	0.007	0.0002	0.0025	0.0012	0.22	-	-	0.0013	-	0.0064	0.437	2.80	0.36	3.61
P	0.079	0.23	1.66	0.005	0.0008	0.017	0.51	0.33	0.025	0.017	0.0002	0.0047	0.0011	-	0.22	0.012	0.0021	-	-	0.501	2.87	0.43	3.24
Q	0.030	0.21	1.65	0.003	0.0010	0.035	0.66	0.16	0.027	0.009	0.0001	0.0028	0.0027	0.15	0.19	-	0.0010	-	-	0.490	3.15	1.24	3.20
R	0.043	0.14	1.46	0.003	0.0008	0.040	1.04	0.28	0.023	0.009	0.0001	0.0030	0.0012	-	1.14	0.018	0.0021	0.0013	-	0.683	3.00	0.49	2.91
S	0.026	0.23	1.45	0.004	0.0023	0.020	0.88	0.84	0.018	0.009	0.0002	0.0034	0.0034	-	-	-	0.0011	-	-	0.494	2.65	0.65	3.62
T	0.057	0.24	1.58	0.005	0.0026	0.031	0.67	0.10	0.012	0.009	0.0002	0.0038	0.0011	0.14	0.49	-	0.0019	-	-	0.506	2.37	0.24	2.62
U	0.079	0.09	1.41	0.006	0.0006	0.021	0.71	0.16	0.020	0.011	0.0002	0.0021	0.0006	0.22	-	-	0.0028	-	-	0.408	5.24	0.16	2.83
V	0.072	0.21	1.42	0.004	0.0007	0.033	1.13	0.26	0.021	0.011	0.0001	0.0025	0.0032	0.36	0.22	-	0.0025	-	-	0.505	4.40	1.95	2.92
W	0.079	0.14	1.88	0.006	0.0009	0.040	0.78	0.45	0.025	0.008	0.0002	0.0025	0.0029	0.45	0.36	0.015	0.0016	-	-	0.640	3.20	1.79	3.55
X	0.071	0.15	1.43	0.003	0.0019	0.017	1.31	0.17	0.014	0.006	0.0001	0.0048	0.0019	0.16	-	-	0.0020	-	-	0.495	1.25	0.44	2.78
Y	0.088	0.21	1.45	0.006	0.0021	0.014	1.29	0.23	0.019	0.014	0.0002	0.0025	0.0017	-	-	0.010	0.0017	-	-	0.444	5.60	0.39	2.91
Z	0.087	0.08	1.97	0.006	0.0018	0.024	1.47	0.45	0.018	0.007	0.0002	0.0023	0.0023	-	-	-	0.0013	-	-	0.601	3.04	0.75	4.03

Formels (3) ACR: $[(Ca) - (0.18 + 130 \cdot (C) - (O) \cdot 1.25) / S]$ Formels (4) Ceq: $5.5 \cdot [C]^{1.3} + 15[P] + 0.90[Mo] + 0.12[Ni] + 7.9[Nb]^{1.3} + 0.5[NiO]$

Underlined values are outside the range of the disclosure

[Table 3]

Sample No	Steel No	Rolling Conditions					Cooling Conditions		Base Metal Properties			Heat Affected Zone Properties		Pre-oxidation			Base Metal Properties After PWHT					Notes
		Reheating Temperature (°C)	Cumulative Rolling Reduction at 910 °C or Higher (%)	Cumulative Rolling Reduction at Less Than 910 °C (%)	Final ThicKness (mm)	Fresh Rolling Temperature (°C)	Cooling Rate (°C/s)	Cooling Stop Temperature (°C)	Temperature (°C)	YS (MPa)	TS (MPa)	ΔE -10 °C (J)	CTOD 6-10 °C (mm)	Mean Particle Size (nm)	[Nb]/[Ti] + [Nb]/[Mo]	YS (MPa)	TS (MPa)	ΔTS (MPa)	ΔTS (°C)			
1	A	1104	39	59	75	700	5	210	520	524	634	239	238	1.73	8.7	0.55	529	617	-18	5	Example	
2	A	1128	45	58	70	720	5	220	530	531	638	208	230	1.28	5.3	0.64	530	621	-17	6	Example	
3	A	1210	50	67	30	760	10	330	590	541	651	98	216	1.15	13.1	0.53	531	605	-46	19	Comparative Example	
4	B	1149	18	80	50	710	10	330	510	551	651	55	241	0.82	8.6	0.45	535	655	4	23	Comparative Example	
5	B	998	35	49	100	740	2	380	530	553	619	211	214	0.87	8.1	0.36	531	626	7	4	Example	
6	C	1104	44	38	750	750	5	180	580	515	638	232	209	1.13	9.7	0.59	517	605	-13	6	Example	
7	C	1064	30	31	150	740	0.8	250	590	445	588	46	221	1.73	13.5	0.25	459	521	-67	16	Comparative Example	
8	D	944	55	48	70	770	5	230	560	517	638	284	181	1.18	6.5	0.60	531	615	-23	8	Example	
9	D	1044	55	48	70	680	5	310	560	405	638	305	229	0.52	-	-	-	-	-	-	Comparative Example	
10	E	1025	61	57	50	750	10	280	680	432	651	297	217	0.53	-	-	-	-	-	-	Comparative Example	
11	E	1079	48	68	60	680	10	190	570	510	651	213	163	1.48	8.2	0.44	516	655	4	7	Example	
12	F	995	42	60	70	690	5	230	570	518	638	106	56	0.16	-	-	-	-	-	-	Comparative Example	
13	F	993	43	59	70	770	5	270	590	560	638	54	64	0.89	-	-	-	-	-	-	Comparative Example	
14	F	1003	48	55	70	720	5	290	590	412	638	248	-	-	-	-	-	-	-	-	Comparative Example	
15	F	1014	48	55	70	690	5	170	560	541	638	46	37	0.11	-	-	-	-	-	-	Comparative Example	
16	F	1128	55	48	70	720	5	250	540	501	638	257	56	0.10	-	-	-	-	-	-	Comparative Example	

ΔTS = TS after PWHT - TS before PWHT

ΔTS = TS after PWHT - 1 mm before PWHT

Underlined values are outside the range of the disclosure.

ΔTS = TS after PWHT - TS before PWHT
ΔYS = YS after PWHT - YS before PWHT

Sample No.		Steel No.	Rolling Conditions					Cooling Conditions		Tempering Temperature (°C)		Base Metal Properties			Heat Affected Zone Properties		Precipitates				Base Metal Properties After PWHT				Notes
			Reheating Temperature (°C)	Continuous Rolling Reduction at 900 °C or Higher (%)	Continuous Rolling Reduction at 900 °C or Less Than 900 °C (%)	Final Thickness (mm)	Finish Rolling Temperature (°C)	Cooling Rate (°C/s)	Cooling Stop Temperature (°C)	Y _P (MPa)	TS (MPa)	σ _E -40 °C (J)	σ _E -10 °C (J)	CTOD 6-10 °C (mm)	Mean Particle Size (nm)	{Nb}/ {Ti}+ {Nb}/ {Mo}	Y _P (MPa)	TS (MPa)	ΔTS (MPa)	Δσ _E (J)					
17		K	1121	47	69	50	700	10	320	500	546	651	83	-	-	-	-	-	-	-	-	Comparative Example			
18		L	976	54	64	50	720	10	260	540	598	651	35	56	0.12	-	-	-	-	-	-	Comparative Example			
19		M	1132	48	68	50	770	10	330	560	588	651	54	-	-	-	-	-	-	-	-	Comparative Example			
20		N	1043	58	66	50	770	10	340	500	497	651	35	66	0.31	-	-	-	-	-	-	Comparative Example			
21		P	987	49	35	100	770	2	340	510	503	619	56	-	-	-	-	-	-	-	-	Comparative Example			
22		Q	971	41	44	100	760	2	270	560	522	619	45	74	0.25	-	-	-	-	-	-	Comparative Example			
23		R	1106	45	39	100	740	2	260	540	546	619	65	40	0.16	-	-	-	-	-	-	Comparative Example			
24		S	1010	38	62	70	740	5	210	550	593	638	68	-	-	-	-	-	-	-	-	Comparative Example			
25		T	1064	45	58	70	740	5	220	500	589	638	51	-	-	-	-	-	-	-	-	Comparative Example			
26		U	976	51	52	70	680	5	300	510	534	638	245	52	0.41	-	-	-	-	-	-	Comparative Example			
27		V	1018	35	64	70	760	5	150	560	506	638	206	19	0.45	-	-	-	-	-	-	Comparative Example			
28		W	1088	50	53	70	680	5	260	560	574	638	64	-	-	-	-	-	-	-	-	Comparative Example			
29		X	1025	59	43	70	730	5	160	500	452	638	200	87	0.34	-	-	-	-	-	-	Comparative Example			
30		Y	1131	45	58	70	760	5	320	550	471	638	229	49	0.25	-	-	-	-	-	-	Comparative Example			
31		Z	1021	41	60	70	770	5	290	530	531	638	273	71	0.14	-	-	-	-	-	-	Comparative Example			
32		C	1120	60	55	40	710	12	420	-	592	702	215	197	1.54	88	0.55	578	673	-78	7	Example			

ΔTS = TS(after PWHT) - TS(before PWHT)
Δσ_E = σ_E(after PWHT) - σ_E(before PWHT)

Underlined values are outside the range of this disclosure

[0085] As listed in Table 2, steel codes A to E represent steel conforming to this disclosure, whereas steel codes F to Z represent comparative steel in which one of the steel components is outside of the range of this disclosure. Sample

Numbers 1, 2, 5, 6, 8, and 11 in Table 3 are all Examples for which the results of the Charpy impact test on the weld bond, the results of the three-point bending CTOD test on the weld bond, the precipitate size/composition in the steel plate, and the PWHT property all satisfy the targets.

[0086] Conversely, in Sample Numbers 3, 4, 7, 9, 10, and 12 to 31, at least one of the steel plate components, the manufacturing conditions, and the precipitate size/composition fall outside of the ranges of this disclosure, and one of the base metal properties, the results of the Charpy impact test on the weld bond, the results of the three-point bending CTOD test on the weld bond, and the PWHT property does not satisfy the target. In Table 3, entries marked with a horizontal line are entries for which measurement was not possible.

[0087] The steel of the Examples according to this disclosure has both excellent strength and toughness of the steel plate, as the yield stress (YS) of the steel plate is 460 MPa or higher, and the Charpy absorbed energy ($vE_{40}^{\circ C}$) is 200 J or higher. Furthermore, in the weld joint bond, $vE_{40}^{\circ C}$ is 150 J or higher, and the CTOD value is 0.5 mm or higher, thus also providing excellent toughness in the heat-affected zone. When the mean particle size of the precipitates is 20 μm or less and $[Nb]/([Ti]+[Nb]+[Mo]) \geq 0.3$, then the base metal property after PWHT is also excellent. By contrast, in the Comparative Examples outside of the ranges of this disclosure, only steel plates for which one of the above-described properties was inferior could be achieved.

Claims

1. A steel plate, comprising:

a chemical composition including, by mass%,

C: 0.020 % to 0.090 %,

Si: 0.01 % to 0.35 %,

Mn: 1.40 % to 2.00 %,

P: 0.008 % or less,

S: 0.0035 % or less,

Al: 0.010 % to 0.060 %,

Ni: 0.40 % to 2.00 %,

Mo: 0.05 % to 0.50 %,

Nb: 0.005 % to 0.040 %,

Ti: 0.005 % to 0.025 %,

N: 0.0020 % to 0.0050 %,

Ca: 0.0005 % to 0.0050 %, and

O: 0.0035 % or less,

Ceq specified by formula (1) below being in a range of 0.420 % to 0.520 %, formulas (2), (3), and (4) below being satisfied, B being controlled to be less than 0.0003 %, and a balance being Fe and incidental impurities; and precipitates including Ti, Nb, and Mo and having a mean particle size of 20 nm or less, wherein $[Nb]/([Ti] + [Nb] + [Mo]) \geq 0.3$, where [Ti] is the Ti content, [Nb] is the Nb content, and [Mo] is the Mo content:

$$Ceq = [C] + [Mn]/6 + ([Cu] + [Ni])/15 + ([Cr] + [Mo] + [V])/5$$

$$(1)$$

$$1.5 \leq [Ti]/[N] \leq 4.0 \quad (2)$$

$$0 < \{[Ca] - (0.18 + 130 \times [Ca]) \times [O]\} / 1.25 / [S] < 1.5 \quad (3)$$

$$5.5[C]^{(4/3)} + 15[P] + 0.90[Mn] + 0.12[Ni] + 7.9[Nb]^{(1/2)} + 0.53[Mo] \leq 3.70 \quad (4)$$

where [M] is the content of element M by mass%.

2. The steel plate of claim 1, wherein the chemical composition further includes, by mass%, at least one selected from the group consisting of Cu: less than 0.7 %, Cr: 0.1 % to 1.0 %, and V: 0.005 % to 0.05 %.

3. The steel plate of claim 1 or 2, wherein the chemical composition further includes, by mass%, at least one selected from the group consisting of Mg: 0.0002 % to 0.0050 % and a REM: 0.0010 % to 0.0200 %.

4. A method for manufacturing a steel plate, the method comprising:

heating steel having the chemical composition of any one of claims 1 to 3 to 950 °C to 1150 °C;
subsequently subjecting the steel to hot rolling at a cumulative rolling reduction of 30 % or higher in a temperature range of 900 °C or higher and a cumulative rolling reduction of 30 % to 70 % in a temperature range of less than 900 °C; and

subsequently cooling the steel at least to 500 °C with a cooling rate of 1.0 °C/s or higher.

5. The method of claim 4, further comprising tempering at 450 °C to 650 °C after the cooling.

FIG. 1

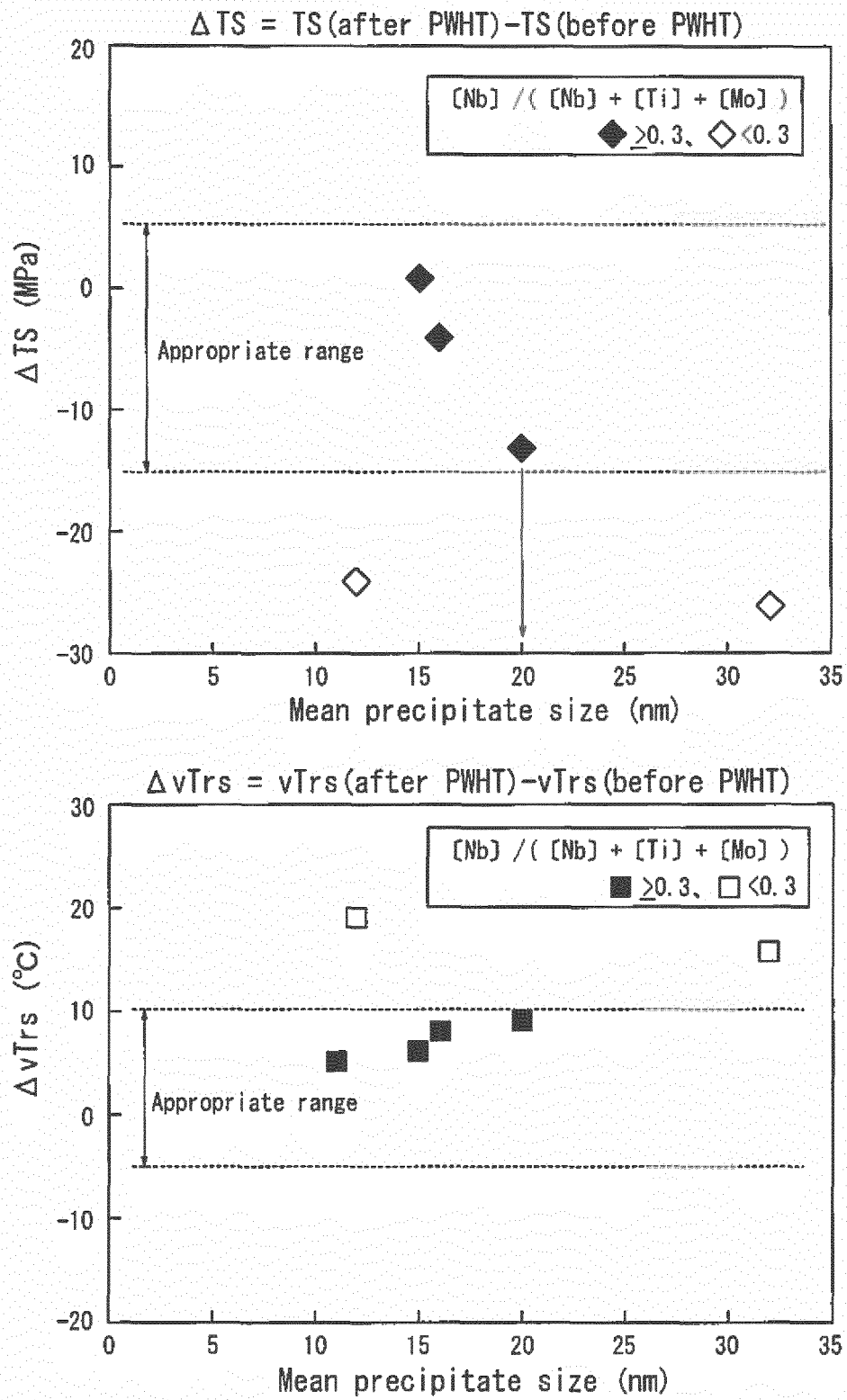
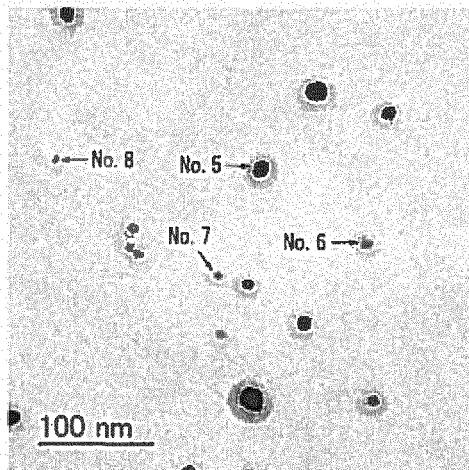
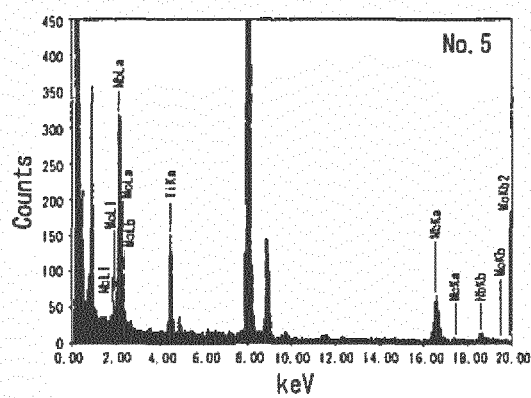
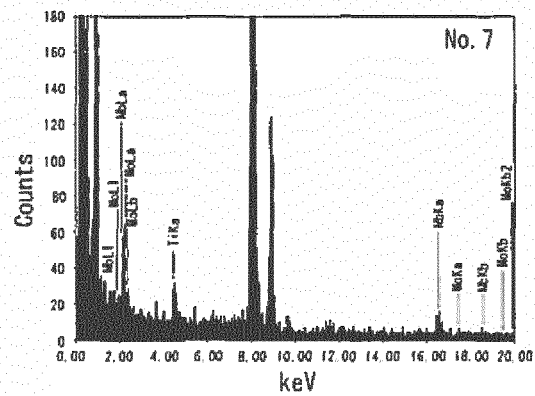
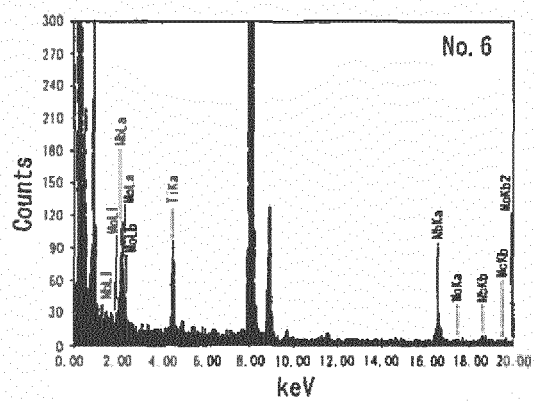
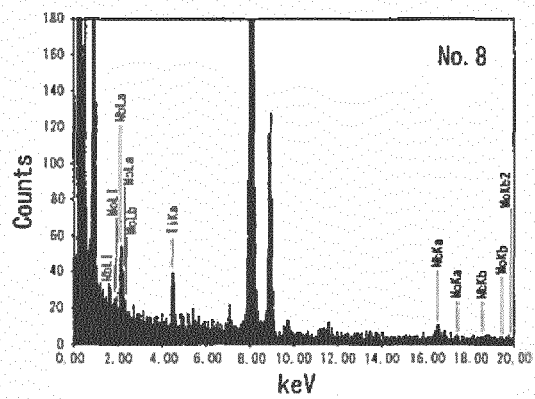


FIG. 2



INTERNATIONAL SEARCH REPORT

International application No.

PCT/JP2014/083321

A. CLASSIFICATION OF SUBJECT MATTER

C22C38/00(2006.01)i, C21D8/02(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)
C22C38/00-38/60, C21D8/00-8/10

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched
Jitsuyo Shinan Koho 1922-1996 Jitsuyo Shinan Toroku Koho 1996-2015
Kokai Jitsuyo Shinan Koho 1971-2015 Toroku Jitsuyo Shinan Koho 1994-2015

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

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	WO 2013/118313 A1 (JFE Steel Corp.), 15 August 2013 (15.08.2013), table 1 & EP 2813596 A1 & CN 104105810 A & KR 10-2014-0117560 A	1-5
A	JP 2006-291349 A (JFE Steel Corp.), 26 October 2006 (26.10.2006), table 1 (Family: none)	1-5
A	JP 2009-263777 A (JFE Steel Corp.), 12 November 2009 (12.11.2009), table 1-1 & WO 2009/123292 A1 & EP 2272994 A1 & CN 102124133 A & KR 10-2010-0116701 A	1-5

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

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23 March 2015 (23.03.15)

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INTERNATIONAL SEARCH REPORT

International application No.

PCT/JP2014/083321

C (Continuation). DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	JP 2001-247932 A (Nippon Steel Corp.), 14 September 2001 (14.09.2001), claims (Family: none)	1-5
A	CN 102691015 A (BAOSHAN IRON & STEEL), 26 September 2012 (26.09.2012), table 2 (Family: none)	1-5
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REFERENCES CITED IN THE DESCRIPTION

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