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(54) STEEL MEMBER AND PROCESS FOR PRODUCING SAME

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EP 2 918 694 B1

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Description

Technical Field

[0001] The present invention relates to a steel member and a method of manufacturing the steel member. Specifically, the invention relates to a steel member produced through performing welding and post-welding heat treatment (PWHT) on a thick steel plate, and particularly relates to a steel member of which the thicknesswise central portion has high strength and high toughness even after high-temperature and long PWHT, and relates to a method of manufacturing the steel member.

Background Art

[0002] In recent trends, higher temperature resistance and higher pressure resistance are required for a medium- or high-temperature pressure vessel for use in chemical industry including oil refining in order to improve operation efficiency. Hence, larger thickness or higher strength is required for a steel plate to be used for a steel member of the pressure vessel or the like. High-level toughness is also required for the steel member from the viewpoint of safety.

[0003] The steel plate is subjected to normalizing and/or quenching so as to have such properties including high strength. However, if the steel plate has a large thickness, the inside of the steel plate (particularly a thicknesswise central portion) is slowly cooled in the normalization or the quenching, and the steel plate is less likely to have the properties including high strength. The steel member for the pressure vessel or the like is produced through welding of the steel plate followed by stress relief annealing (post-welding heat treatment, hereinafter sometimes referred to as "PWHT") for relieving stress. If the steel plate has a large thickness, long PWHT is necessary for relieving stress. The steel plate subjected to long PWHT, however, is disadvantageously degraded in toughness or the like.

[0004] In a possible approach for solving such issues, quenching is performed in place of normalizing that has been performed so that the thicknesswise central portion is rapidly cooled. For a steel plate having a large thickness, however, such an approach also cannot sufficiently increase the cooling rate, i.e., does not sufficiently meet the demand of high strength and high toughness.

[0005] In an approach that allows the steel member to have high toughness, the amount of alloy elements is increased. Cr-Mo steel containing Cr and Mo as alloy elements is used for the steel member for the pressure vessel or the like. It is known that when 2.25Cr-1.0Mo steel is, for example, used as the Cr-Mo steel, good toughness is exhibited even in the thicknesswise central portion of a thick steel plate while such a portion is in general difficult to have good toughness. In recent years, however, there is an increased trend toward resources saving and cost reduction. Hence, there is a strong demand for developing a steel member of which the thicknesswise central portion has high strength and high toughness on the premise that the steel member is produced using Cr-Mo steel (for example, 1.25Cr-1.0Mo steel) having a lower amount of alloy elements than the 2.25Cr-1.0Mo steel.

[0006] To meet such an issue, there has been provided a technique for achieving high strength and high toughness by optimally adjusting the chemical composition while the amount of alloy elements is controlled to be small. For example, PTL 1 and PTL 2 each disclose a technique for improving low-temperature toughness of steel having a composition of 1.25Cr-0.5Mo level that is difficult to have good toughness.

[0007] PTL 1 discloses a technique for providing good hardenability by adding Nb and Ca, and suppressing degradation in properties during stress relief (SR) annealing. However, when this technique is used for an extremely thick steel plate that is mainly formed by an ingot casting process, the Ca forms coarse inclusions that may adversely affect toughness. It is therefore considered to be difficult that the thicknesswise central portion of a steel member having a larger thickness stably has good toughness.

[0008] PTL 2 discloses a technique for decreasing austenite grain size by performing controlled rolling or controlled rolling combined with accelerated cooling before quenching in a manufacturing process, and thereby providing good low-temperature toughness. This technique however is difficult to be practically used since the controlled rolling extremely lowers productivity of a rolling line for manufacturing an extremely thick steel plate having a thickness of more than 100 mm.

Citation List

Patent Literature

[0009]

PTL 1: Japanese Patent No. 2743765
PTL 2: Japanese Unexamined Patent Application Publication No. 2000-345281
PTL 3: European Patent Application EP1764423

Summary of Invention

Technical Problem

[0010] An object of the invention, which has been made in light of the above-described circumstances, is to provide a steel member produced using a thick steel plate, of which the inside (thicknesswise central portion) has high strength and high toughness even after being subjected to welding followed by long (particularly high-temperature and long) PWHT in the manufacturing process of the steel member, and provide a method of manufacturing the steel member.

Solution to Problem

[0011] A steel member of the invention, which has succeeded in solving the above-described problem, consisting of

C: 0.12 to 0.18%, by mass percent, the same applies to the following for the chemical components,
Si: 0.50 to 0.80%,
Mn: 0.40 to 0.70%,
P: 0.015% or less, not including 0%,
S: 0.005% or less, not including 0%,
Al: 0.040 to 0.080%,
Cu: 0.05 to 0.40%,
Ni: 0.05 to 0.40%,
Cr: 1.25 to 1.50%,
Mo: 0.45 to 0.65%,
N: 0.0030 to 0.0060%,
B: 0.0003 to 0.0010%, and optionally
V: more than 0% to 0.030%,

with the remainder consisting of Fe and inevitable impurities,
in which a microstructure of the thicknesswise central portion of the steel member satisfies all of the following (a) to (d).

(a) The microstructure is at least one of tempered bainite and tempered martensite.
(b) The mean equivalent circle diameter of grains is 20 μm or less, each grain being surrounded by a large-angle grain boundary having a crystal misorientation of 15° or more between two adjacent grains.
(c) The maximum size of grain boundary carbide is 0.8 μm or less.
(d) The fraction of the grain boundary carbide is 1.0 area% or more.

[0012] The invention also includes a method of manufacturing the steel member. The method includes performing hot rolling on a slab having a chemical composition of the above-described steel member, after the hot rolling, performing quenching under a condition of heating temperature of 900 to 950°C and holding time after the quenching, performing welding and post-welding heat treatment.
[0013] After the quenching, tempering may be further performed at a temperature of 620°C to A_{c1} point.
[0014] When the post-welding heat treatment is performed at a heating temperature and for a heating time such that a value P represented by Formula (1) is 20 or more, a steel member having good properties can also be produced.

$$\text{Value P} = T \times (20 + \log t) \times 10^{-3} \cdots (1)$$

where T is heating temperature (K), and t is heating time (hr).

Advantageous Effects of Invention

[0015] According to the invention, there is provided a steel member produced using a thick steel plate, of which the inside (thicknesswise central portion) has high strength and high toughness even after being subjected to welding followed by long (particularly high-temperature and long) PWHT in the manufacturing process of the steel member. Consequently, it is possible to provide a medium- or high- temperature pressure vessel or the like that is produced using a thick steel plate, and has high strength and high toughness even after being subjected to high-temperature and long PWHT.

[0016] Furthermore, the steel member of the invention is controlled to be low in amount of alloy elements, and therefore contributes to resources saving and cost reduction.

Description of Embodiments

[0017] The inventors have made earnest study to provide a steel member that is premised to be produced using a thick steel plate (hereinafter, sometimes simply referred to as "steel plate") that is composed of Cr-Mo steel (for example, 1.25Cr-0.5Mo steel) having a lower amount of alloy elements than the 2.25Cr-1.0Mo steel and has a thickness of 90 mm or more, the thicknesswise central portion of the steel plate having high toughness (low-temperature toughness) and high strength even if the thick steel plate is subjected to long PWHT.

[0018] As a result, they have found that the following approaches are specifically effective for providing high toughness of the thicknesswise central portion of the steel member.

- A fine microstructure is formed. In detail, (a) the microstructure is controlled to be at least one of tempered bainite and tempered martensite, and (b) the mean equivalent circle diameter (hereinafter, sometimes simply referred to as "large-angle grain boundary size") of grains, each grain being surrounded by a large-angle grain boundary having a crystal misorientation of 15° or more between two adjacent grains, is controlled to be 20 μm or less.
- Grain boundary carbide that tends to be coarsened and become a fracture origin is refined. In detail, (c) the maximum size of the grain boundary carbide is controlled to be 0.8 μm or less.
- Reduction in temper embrittlement sensitivity (hereinafter, sometimes referred to as "reduction in temper embrittlement" or "reduction in grain boundary fracture (grain boundary cracking)) is performed. In detail, the steel member is controlled to satisfy the composition described later.

[0019] In addition, they have found that the following approaches are specifically effective for providing high strength of the thicknesswise central portion of the steel member.

- A fine microstructure is formed. In detail, (a) the microstructure is controlled to be at least one of tempered bainite and tempered martensite.

The fraction of grain boundary carbide is controlled. In detail, (d) the fraction of the grain boundary carbide is controlled to be 1.0 area% or more.

[0020] The above-described (a) to (d) on the microstructure of the thicknesswise central portion of the steel member of the invention are now described.

[0021] In the following description, "microstructure of the thicknesswise central portion" is simply referred to as "microstructure". The following properties, i.e., strength and toughness (low-temperature toughness) mean properties of at least the thicknesswise central portion of the steel member (i.e., the thick steel plate subjected to welding and PWHT).

[(a) Microstructure is at least one of tempered bainite and tempered martensite, and (b) Mean equivalent circle diameter of grains is 20 μm or less, each grain being surrounded by a large-angle grain boundary having a crystal misorientation (crystal misorientation) of 15° or more between two adjacent grains]

[0022] Each of the tempered bainite and the tempered martensite is a fine microstructure, and is particularly effective for providing high strength and high toughness of the thicknesswise central portion of an extremely thick steel plate. The microstructure of the steel member of the invention is at least one of tempered bainite and tempered martensite, and does not substantially include other phases such as polygonal ferrite, retained austenite, and perlite. When the polygonal ferrite is contained, the microstructure mainly includes an upper bainite structure having a large grain size, so that good toughness cannot be provided.

[0023] As described above, the microstructure of the thicknesswise central portion is controlled to be at least one of tempered bainite and tempered martensite, thereby the microstructure can be refined. In the invention, a large-angle grain boundary size of the microstructure (i.e., at least one of tempered bainite and tempered martensite) of the thicknesswise central portion is controlled to be 20 μm or less to achieve high toughness through steady refinement of the microstructure.

[0024] A so-called large-angle grain boundary, which has a crystal misorientation (crystal misorientation) of 15° or more between two adjacent grains in most cases, has a large crystal misorientation between two adjacent grains. Hence, brittle fracture is curvedly propagated, and a surface unit of the brittle fracture is reduced, contributing to improvement in toughness in the microstructure including tempered bainite and tempered martensite. In the invention, the large-angle grain boundary size (the mean equivalent circle diameter of grains each being surrounded by the large-angle grain boundary) is controlled to be 20 μm or less as described above to increase the large-angle grain boundaries in a certain

region in order to sufficiently improve toughness. The large-angle grain boundary size can be determined by an electron back scattering pattern (EBSP) method as described later in an embodiment. The large-angle grain boundary size is preferably 15 μm or less, and more preferably 13 μm or less. The lower limit of the large-angle grain boundary size is roughly 10 μm due to manufacturing reasons.

[(c) Maximum size of grain boundary carbide is 0.8 μm or less, and (d) Fraction of grain boundary carbide of 1.0 area% or more]

[0025] As described above, the steel member of the invention is subjected to PWHT (particularly long PWHT, and further particularly high-temperature and long PWHT). When the Cr-Mo steel as a material of the steel member is subjected to PWHT, grain boundary carbide of $M_{23}C_6$ is typically formed. When the PWHT is performed under a severe condition such as high temperature and long time, the grain boundary carbide is coarsened and thus tends to be a fracture origin, causing degradation in toughness. In the invention, the maximum size of the grain boundary carbide is controlled to be 0.8 μm or less in the thicknesswise central portion of the steel member, thereby the steel member has good toughness. The maximum size of the grain boundary carbide is preferably 0.6 μm or less, and more preferably 0.5 μm or less. The lower limit of the maximum size of the grain boundary carbide is roughly 0.2 μm within a range of each of the composition and the manufacturing condition defined in the invention.

[0026] When the amount of the grain boundary carbide is too small, the steel member is difficult to have high strength. Hence, the fraction of the grain boundary carbide (the proportion of the grain boundary carbide in the entire microstructure of the thicknesswise central portion as described later in the embodiment) is controlled to be 1.0 area% or more. The fraction of the grain boundary carbide is preferably 2.0 area% or more. Although the fraction of the grain boundary carbide increases with an increase in the content of C, the increased C content coarsens the carbide, and tends to degrade toughness. Consequently, from the viewpoint of providing good toughness, the upper limit of the C content is defined as described later, and the upper limit of the fraction of the grain boundary carbide is about 5.0 area% within the range of the C content.

[0027] In the invention, while the microstructure of the thicknesswise central portion must be controlled as described above, the microstructure of any other region (for example, a thicknesswise surface portion) is not limited. A portion closer to the surface than the thicknesswise central portion is in general rapidly cooled in quenching compared with the thicknesswise central portion; hence, such a portion tends to have a finer microstructure than the thicknesswise central portion, and tends to be better in both strength and toughness than the thicknesswise central portion.

[0028] To form the fine microstructure as described in the (a) and (b) in the thicknesswise central portion, it is necessary that B is contained as a chemical component in the amount as described later so as to exist in a form of free B (dissolved B) to improve hardenability. To achieve this, it is important that Al is added in the amount as described later so that N, which is easily bonded to B and form BN, is fixed in a form of AlN (that is useful for suppressing coarsening of prior austenite (γ) grains during quenching to form a fine microstructure) in order to provide a sufficient amount of free B. Furthermore, as described in detail later, it is important to appropriately control the manufacturing condition such as heating temperature and heating retention time in quenching.

[0029] In addition, it is necessary to control the C content and the Cr content to achieve the size and the fraction of the grain boundary carbide as described in the (c) and (d).

[0030] Furthermore, it is necessary to control the content of each of elements including Si in order to control the temper embrittlement sensitivity to provide good toughness.

[0031] Description is now made on the (chemical) composition of the steel member necessary for providing the microstructure and the properties.

[C: 0.12 to 0.18%]

[0032] C is an element necessary for forming at least one of tempered bainite and tempered martensite during quenching of a thick steel plate even in the thicknesswise central portion of the steel plate while such a portion is slowly cooled in the quenching. Moreover, C is an element necessary for forming the grain boundary carbide to provide sufficient strength of a base metal. To allow such effects to be sufficiently exhibited, the C content is 0.12% or more. The C content is preferably 0.13% or more, and more preferably 0.15% or more. However, if the C content is excessive, the grain boundary carbide is coarsened after long PWHT, and toughness is degraded. In addition, weld cracking easily occurs during welding of the steel plate. Consequently, the C content is 0.18% or less. The C content is preferably 0.17% or less, and more preferably 0.16% or less.

[Si: 0.50 to 0.80%]

[0033] Si is an element effective for increasing strength of a base metal (i.e., strength of the thicknesswise central

portion) of the steel member. Si is also an element used as a deoxidizer. To allow such effects to be exhibited, the Si content is 0.50% or more. The Si content is preferably 0.55% or more, and more preferably 0.60% or more. However, if the Si content is excessive, temper embrittlement sensitivity increases, and toughness is degraded. Hence, the Si content is 0.80% or less. The Si content is preferably 0.75% or less, and more preferably 0.70% or less.

[Mn: 0.40 to 0.70%]

[0034] Mn is an element effective for stabilizing austenite and lowering transformation temperature, and thus improving hardenability and forming a fine microstructure, and consequently providing high strength and high toughness. To allow such an effect to be exhibited, 0.40% or more of Mn is contained. The Mn content is preferably 0.45% or more, and more preferably 0.48% or more. However, if Mn is excessively contained, the temper embrittlement sensitivity increases, and toughness is degraded. Consequently, the upper limit of the Mn content is 0.70%. The Mn content is preferably 0.65% or less, and more preferably 0.60% or less.

[P: 0.015% or less (not including 0%)]

[0035] P as an inevitable impurity adversely affects toughness of each of the base metal and the weld bead, and segregates in a grain boundary of the steel member, causing grain boundary cracking and degradation in toughness. The P content is controlled to be 0.015% or less to prevent such disadvantages. The P content is preferably 0.010% or less.

[S: 0.005% or less (not including 0%)]

[0036] S forms MnS and easily causes weld cracking during welding of a steel plate. Consequently, the S content is preferably small as much as possible, and is controlled to be 0.005% or less, preferably 0.003% or less.

[Al: 0.040 to 0.080%]

[0037] As described above, Al is an important element in the invention, and is necessary for fixing N in a form of AlN during quenching to provide good hardenability by free B. AlN is also useful for suppressing coarsening of prior γ grains during quenching and forming a fine microstructure. Furthermore, Al is an element necessary for deoxidation. To allow such effects to be exhibited, the Al content is 0.040% or more. The Al content is preferably 0.045% or more, and more preferably 0.050% or more. If the Al content is excessive, coarse alumina-based inclusions are formed, and toughness is degraded. Consequently, the Al content is 0.080% or less. The Al content is preferably 0.075% or less, and more preferably 0.071% or less.

[Cu: 0.05 to 0.40%, and Ni: 0.05 to 0.40%]

[0038] Cu and Ni are each an element effective for increasing strength without significantly degrading toughness. To allow such an effect to be sufficiently exhibited, Cu is contained in the amount of 0.05% or more (preferably 0.10% or more, more preferably 0.11% or more, and further preferably 0.20% or more). In addition, Ni is contained in the amount of 0.05% or more (preferably 0.10% or more, more preferably 0.15% or more, and further preferably 0.16% or more). However, when such elements are each added in a large amount, cost is increased; hence, the upper limit of the content of each of Cu and Ni is 0.40% or less. The Cu content is more preferably 0.37% or less, and further preferably 0.30% or less. The Ni content is more preferably 0.38% or less, and further preferably 0.30% or less.

[Cr: 1.25 to 1.50%]

[0039] Cr is an element effective for suppressing coarsening of carbide due to PWHT, and providing good toughness of the steel member. Cr is also an element effective for providing high strength in a medium- or high- temperature region, and further effective for improving corrosion resistance. To allow such effects to be exhibited, Cr is contained in the amount of 1.25% or more. The Cr content is preferably 1.35% or more, and more preferably 1.39% or more. If Cr is excessively contained, temper embrittlement sensitivity increases, and grain boundary fracture easily occurs after PWHT, leading to an adverse effect on toughness. In addition, excessive Cr degrades workability and weldability, and increases manufacturing cost. Consequently, the Cr content is 1.50% or less. The Cr content is preferably 1.45% or less, and more preferably 1.40% or less.

[Mo: 0.45 to 0.65%]

[0040] Mo is an element effective for improving hardenability and reducing temper embrittlement. To allow such effects to be exhibited, Mo is necessary to be contained in the amount of 0.45% or more. The Mo content is preferably 0.50% or more, and more preferably 0.55% or more. When the Mo content exceeds 0.65%, the effects are not so enhanced, and manufacturing cost is increased; hence, the upper limit of the Mo content is 0.65%. The Mo content is preferably 0.62% or less, and more preferably 0.60% or less.

[N: 0.0030 to 0.0060%]

[0041] N is an important element in addition to Al in the invention. N is fixed during quenching through formation of AlN, thereby the effect of improving hardenability by free B can be maximally exhibited. AlN is also useful for suppressing coarsening of prior Y grains during quenching and forming a fine microstructure. If the N content is less than 0.0030%, AlN becomes insufficient, and the prior Y grains are coarsened. As a result, the fine microstructure is not formed, and toughness is degraded. Consequently, the N content is 0.0030% or more. The N content is preferably 0.0035% or more, and more preferably 0.0040% or more. If the N content exceeds 0.0060%, the effect of fixing N by Al is substantially not exhibited, and BN is formed, thereby the effect of improving hardenability by free B is prevented. As a result, the microstructure is coarsened, and toughness is degraded. Consequently, the N content is 0.0060% or less. The N content is preferably 0.0055% or less, and more preferably 0.0050% or less.

[B: 0.0003 to 0.0010%]

[0042] As described above, B is contained in a form of free B (dissolved B) and thus improves hardenability, which in particular makes it possible to form a fine microstructure even in the thicknesswise central portion of the thick steel plate while such a portion is slowly cooled in quenching. As a result, the thicknesswise central portion is allowed to have good toughness. To allow such an effect to be exhibited, 0.0003% or more of B is necessary though it is premised that the Al content and the N content are controlled as described above, and a quenching condition is controlled as described later. The B content is preferably 0.0005% or more, and more preferably 0.0007% or more. If B is excessively contained, hardenability may be rather degraded, or weld cracking may be caused; hence, the upper limit of the B content is 0.0010%. The B content is preferably 0.0009% or less, and more preferably 0.0008% or less.

[0043] The steel member of the invention contains the above-described components with the remainder consisting of iron and inevitable impurities. The steel member may further contain V in an appropriate amount as described below in addition to such elements.

[V: more than 0% to 0.030%]

[0044] V is an element that contributes to increasing strength through formation of carbide and nitride, and is effective for improving hardenability and forming a fine microstructure. To allow such effects to be exhibited, V is preferably contained in the amount of 0.005% or more. The V content is more preferably 0.010% or more. Excessive addition of V causes an increase in cost; hence, the upper limit of the V content is preferably 0.030%. The V content is more preferably 0.028% or less, and further preferably 0.020% or less.

[0045] A method of manufacturing the steel member of the invention is now described.

[0046] A slab having the above-described chemical composition of the steel member is hot-rolled in a usual manner to produce a thick steel plate. Subsequently, the thick steel plate is subjected to hardening (and tempering as necessary). The thick steel plate has a thickness of 90 mm or more (particularly 100 mm or more, and further particularly 120 mm or more).

[0047] To form the fine microstructure of the steel member defined by the (a) and (b), the thick steel plate to be used for the steel member must be subjected to hardening under the following condition.

[Heating Temperature of 900 to 950°C and Heating Retention Time of 60 min or more in Hardening]

[0048] Heating temperature in hardening is controlled to be 900 to 950°C (in particular, controlled to be 900°C or more), and heating retention time therein is controlled to be 60 min or more, thereby the prior Y grains can be somewhat grown. As a result, hardenability is improved, and a fine microstructure can be formed.

[0049] If the heating temperature in hardening is below 900°C, the prior Y grains are still fine during the hardening; hence, the fine microstructure is not formed in a slowly cooled portion such as the thicknesswise central portion of the thick steel plate, and good toughness cannot be provided. Consequently, the heating temperature in hardening is 900°C or more. The heating temperature is preferably 910°C or more. If the heating temperature exceeds 950°C, some of N

that has been fixed in a form of AlN is dissolved, and is bonded to B and formed into BN, so that the effect of improving hardenability by free B is not exhibited. As a result, the fine microstructure is not formed, and toughness is degraded. Consequently, the heating temperature in hardening is 950°C or less. The heating temperature is preferably 940°C or less.

[0050] Even if the heating temperature is within the above-described range, the retention time at the heating temperature (heating retention time) of shorter than 60 min allows the prior Y grains to be still fine. Hence, sufficient hardenability is not provided even if the predetermined amount of B is contained. As a result, the microstructure is coarsened and toughness is degraded. Consequently, the heating retention time is 60 min or more. The heating retention time is preferably 80 min or more. The upper limit of the heating retention time is about 150 min from the viewpoint of productivity or the like.

[0051] When the condition of hardening is controlled as described above so that the prior Y grain size is within a range roughly from 50 to 100 μm, the fine microstructure is preferably easily formed.

[0052] When tempering is performed following the hardening, the tempering is recommended to be performed under the following condition.

[Tempering Temperature: 620°C to A_{c1} point]

[0053] In the hardening, the neighborhood of the surface is rapidly cooled regardless of thickness, and thus hardness of the surface is easily increased. Hence, workability such as bendability of the steel plate can be improved through tempering after the hardening. Consequently, tempering is preferably performed to lower the hardness of the surface in the manufacturing process of the steel member from the viewpoint of improving the workability of the steel plate. In the tempering condition, the tempering temperature is preferably 620°C to A_{c1} point. The tempering temperature of 620°C or more allows the hardness of the surface to be sufficiently lowered, and allows good workability to be maintained. The tempering temperature is more preferably 700°C or more. When the tempering temperature exceeds the A_{c1} point, some of the microstructure is reversely transformed and then air-cooled; hence, polygonal ferrite is mixedly formed in the microstructure. As a result, strength is lowered, and toughness is also degraded due to the coarse microstructure of the reversely transformed region. Consequently, the upper limit of the tempering temperature is preferably equal to the A_{c1} point. The tempering temperature is more preferably 750°C or less.

[0054] The A_{c1} point is calculated from the expression of A_{c1} point = 723 - 14 × [Mn] + 22 × [Si] - 14.4 × [Ni] + 23.3 × [Cr] (where the [Mn], [Si], [Ni], and [Cr] represent the contents (by mass percent) of Mn, Si, Ni, and Cr, respectively).

[0055] The steel member of the invention is produced as follows: the thick steel plate produced through the hardening (and tempering as necessary) is subjected to welding in a usual manner, and further subjected to post-welding heat treatment (PWHT) for removing strain as described above to produce the steel member. In a condition of the PWHT, heating temperature is 600 to 690°C, and heating time is 5 to 22 hours. In particular, when the steel member of the invention is subjected to PWHT under a severe condition of high temperature and long time, which allows the value P (a value referred to as Hollomon-Jaffe parameter) represented by Formula (1) to be 20 or more (for example, the value P is 20.3 for temperature of 680°C or more and heating time of 20 hours or more), the effects of the invention are sufficiently exhibited.

$$\text{Value P} = T \times (20 + \log t) \times 10^{-3} \cdots (1)$$

(where T is heating temperature (K), and t is heating time (hours)).

[0056] The invention covers a thick steel plate of which the thicknesswise central portion is in general difficult to have high strength and high toughness after PWHT (particularly high-temperature and long PWHT). The invention therefore also covers a steel member, which is produced by such a thick steel plate, having a thickness of 90 mm or more (particularly 100 mm or more, and further particularly 120 mm or more).

[0057] For example, the steel member of the invention can be used for a medium- or high- temperature pressure vessel and the like for use in chemical industry including oil refining.

[0058] This application claims the benefit of Japanese Priority Patent Application JP 2012-247775 filed on November 9, 2012, the entire contents of which are incorporated herein by reference.

Embodiment

[0059] Although the invention is now described in detail with an embodiment, the invention should not be limited thereto, and modifications or alterations thereof may be made within the scope without departing from the gist described before and later, all of which are included in the technical scope of the invention.

[0060] A slab satisfying the (chemical) composition shown in Table 1 (the remainder consisting of iron and inevitable

impurities, and each blank in Table 1 indicating no element being added) was hot-rolled in a usual manner, and then subjected to hardening under the condition shown in Table 2 to produce steel plates each having a thickness (also being a thickness of a test specimen simulating the steel member) shown in Table 2. In examples other than steel No. A1-13 in Tables 2 and 3, each steel plate was further subjected to tempering under the condition shown in Table 2 or 3. The heating temperature at each of hardening and tempering refers to the temperature of the thicknesswise central portion of the steel plate, which was calculated by calculus of finite differences based on the furnace atmosphere temperature of the heat treatment furnace and in-furnace time, or was measured using a thermocouple inserted into a dummy steel plate having the same thickness in an experimental furnace.

[0061] Furthermore, the steel plate was heat-treated under a condition of heating temperature of 690°C and heating retention time of 22 hours (an extremely severe condition among currently-practiced conditions, the value P is 20.6 at the condition) in a truck-type electric furnace (air atmosphere) as a simulation of the PWHT after welding, so that a test specimen simulating the steel member was produced. The heating rate from room temperature to the heating temperature and the cooling rate from the heating temperature to room temperature were each 55 °C/hr or less.

[0062] In manufacture of the steel member, the steel plate is subjected to welding followed by PWHT. For example, when multilayer welding is performed as the welding, the welding is less likely to adversely affect the properties (particularly toughness) of the steel member (including a weld heat-affected zone). In this embodiment, therefore, each test specimen was produced without being subjected to heat treatment following welding.

[0063] The test specimen produced in the above manner was used to evaluate a microstructure, and perform a tensile test and a Charpy impact test according to the following procedure. In addition, surface hardness was measured using the steel plate before being subjected to PWHT in order to evaluate workability (properties to be required in the manufacturing process of the steel member) of the steel plate.

[Observation of Microstructure]

[0064] The microstructure was observed as follows.

(1) A sample was taken from the steel plate to allow observation of a thicknesswise section including both sides of the steel plate in a direction parallel to a rolling direction and perpendicular to the surface of the steel plate.
(2) The observation surface was mirror-finished by polishing with wet emery papers (#150 to #1000) or by a polishing method having similar polishing capability (such as polishing using an abrasive including a diamond slurry).
(3) The polished sample was etched using a 3%-nital solution to allow crystal grain boundaries to be shown.
(4) Photographs of the shown microstructure were each taken at a magnification of 400 times (a 6 cm $\times$ 8 cm photograph was taken in this embodiment) in a portion of t/2 (half the thickness). Subsequently, polygonal ferrite formed in a prior austenite grain boundary was determined in each photograph, and was blacked out. Subsequently, the photographs were loaded into an image analyzer (the region of each photograph corresponds to 150 μm $\times$ 200 μm for 400 times). The loading into the image analyzer was performed in each magnification such that the total size of the regions was equal to or larger than 1 mm $\times$ 1 mm (i.e., at least 35 photographs were loaded for 400 times).
(5) The image analyzer calculated a black area ratio for each photograph, and determined, assuming the average of the black area ratios in all the photographs as the polygonal ferrite (PF) fraction, the total fraction minus the PF fraction as the fraction of at least one of tempered bainite and tempered martensite (B + M).

[0065] The tempered bainite described herein refers to a microstructure formed through tempering of upper bainite, lower bainite, or bainitic ferrite. Such phases, including tempered martensite, are typically difficult to be sorted out, and the microstructure is sufficiently tempered after PWHT. Hence, any of the phases other than polygonal ferrite was defined as at least one of tempered bainite and tempered martensite (B + M). It was also found that no perlite phase was contained in any of the test specimens used in this embodiment.

[Measurement of Large-Angle Grain Boundary Size by EBSP Method]

[0066] The mean equivalent circle diameter (large-angle grain boundary size) of grains, each grain being surrounded by a large-angle grain boundary having a crystal misorientation of 15° or more between two adjacent grains, was measured using the EBSP method. The measurement procedure was as follows.

(1) A sample was taken from the steel plate to allow observation of a thicknesswise section including both sides of the steel plate in a direction parallel to a rolling direction and perpendicular to the surface of the steel plate.
(2) The observation surface was mirror-finished by polishing with wet emery papers (#150 to #1000) or by a polishing method having similar polishing capability (such as polishing using an abrasive including a diamond slurry).
(3) A boundary having a crystal misorientation of 15° or more was assumed as a grain boundary, and size of each

grain (large-angle grain) surrounded by the grain boundary was measured using an EBSP system TexSEM from Laboratories Inc. in the thicknesswise t/2 portion within a measurement range of 200 × 200 μm and with a pitch of 0.5 μm. A measurement point having a confidence index of less than 0.1, the confidence index indicating reliability of the measuring orientation, was excluded from the analysis object.

(4) The sizes of the grains each being surrounded by a large-angle grain boundary were measured in this way, and the average of the sizes was calculated and assumed as "mean equivalent circle diameter of grains each being surrounded by a large-angle grain boundary having a crystal misorientation of 15° or more between two adjacent grains (of at least one of tempered bainite and tempered martensite)" of the invention. When the size of the grain surrounded by a large-angle grain boundary was 1.0 μm or less, such a grain was determined as measurement noise, and was excluded from the object of the average calculation.

[Measurement of Size and Fraction of Grain Boundary Carbide]

[0067] The size and the fraction of the grain boundary carbide were measured as follows.

(1) A sample was taken from the steel plate to allow observation of a thicknesswise section including both sides of the steel plate in a direction parallel to a rolling direction and perpendicular to the surface of the steel plate.

(2) The observation surface was mirror-finished by polishing with wet emery papers (#150 to #1000) or by a polishing method having similar polishing capability (such as polishing using an abrasive including a diamond slurry).

(3) The polished sample was etched using a 3%-nital solution to allow crystal grain boundaries to be shown.

(4) Photographs of the shown microstructure were each taken at a magnification of 1000 times (a 6 cm × 8 cm photograph was taken in this embodiment) in a portion of t/2 (half the thickness). Subsequently, the photographs were loaded into an image analyzer (the region of each photograph corresponds to 60 μm × 80 μm for 1000 times). The loading into the image analyzer was performed such that the total size of the regions was equal to or larger than 0.4 mm × 0.4 mm (i.e., at least 35 photographs were loaded for 1000 times).

(5) The image analyzer calculated size (minor axis length) and area ratio of grain boundary carbide for each photograph, and calculated the maximum size of the grain boundary carbide in all the photographs, and determined the average of the area ratios of the grain boundary carbide as the fraction of the grain boundary carbide.

[Tensile Test (Evaluation of Tensile Characteristics)]

[0068] A round-bar tensile test piece was sampled from the portion of t/2 (half the thickness) in a direction perpendicular to the rolling direction, and was subjected to a tensile test according to the procedure of ASTM A370 so that yield strength and tensile strength were measured. A sample having a yield strength of 310 MPa or more and a tensile strength of 515 MPa or more was evaluated to have high strength (good tensile characteristics).

[Charpy Impact Test (Evaluation of Impact Characteristics)]

[0069] A full-size V-notch test piece was sampled from the portion of t/2 (half the thickness) in a direction perpendicular to the rolling direction, and was subjected to a Charpy impact test at a test temperature of -10°C according to the procedure of ASTM A370 so that absorbed energy was measured. The average of absorbed energy values of three test pieces was determined as that absorbed energy. A sample having an absorbed energy of 100 J or more was evaluated to have good toughness (good impact characteristics).

[Measurement of Surface Hardness (Evaluation of Workability of Steel Plate)]

[0070] To evaluate workability of the steel plate, a steel plate before being subjected to PWHT was subjected to a Brinell hardness test at a depth position of 1 mm from the surface of the steel plate according to the procedure of ASTM 370. A sample showing up to 250 HB was evaluated to be excellent (○) in workability, and a sample showing higher than 250 HB was evaluated to be normal (Δ) in workability.

[0071] Such results are shown in Tables 2 and 3.

Table 1

Billet No.	Composition (mass%)													A_{C1} point (°C)
	C	Si	Mn	P	S	Al	Cu	Ni	Cr	Mo	V	B	N	
A1	0.16	0.50	0.45	0.007	0.002	0.056	0.10	0.15	1.44	0.62		0.0008	0.0040	759
A2	0.12	0.54	0.59	0.007	0.003	0.055	0.37	0.4	1.39	0.62		0.0005	0.0052	753
A3	0.18	0.60	0.51	0.007	0.003	0.055	0.10	0.15	1.39	060		0.0009	0.0050	759
A4	0.17	0.55	0.60	0.015	0.005	0.058	0.37	0.38	1.44	0.62	0.030	0.0010	0.0045	755
A5	0.17	0.55	0.40	0.007	0.003	0.058	0.40	0.38	1.25	0.62	0.028	0.0008	0.0045	753
A6	0.16	0.55	0.46	0.007	0:001	0.080	0.11	0.16	1.44	0.65		0.0007	0.0060	760
A7	0.16	0.55	0.47	0.007	0.002	0.057	0.05	0.06	1.43	0.62		0.0008	0.0044	761
A8	0.16	0.55	0.47	0.007	0.001	0.071	0.11	0.17	1.44	0.62		00008	0.0030	760
A9	0.16	0.55	0.47	0.007	0,002	0.040	0.11	0.16	1.50	0.62		0.0007	0.0054	761
A10	0.16	0.60	0.51	0.007	0.003	0.056	0.11	0.16	1.39	0.59		0.0003	0.0051	759
A11	0.17	0.55	0.40	0.007	0.003	0.058	0.40	0.38	1.25	0.45	0.028	0.0008	0.0045	753
A12	0.17	0.55	0.70	0.007	0.003	0.058	0.37	0.38	1.39	0.62	0.005	0.0008	0.0045	752
A13	0.16	0.80	0.45	0.007	0.003	0.058	0.11	0.15	1.39	0.62	0.027	0.0008	0.0045	765
A14	0.17	0.55	0.45	0.001	0.001	0.058	0.37	0.38	1.44	0.62	0.030	0.0010	0.0045	757
B1	0.08	0.54	0.63	0.007	0.003	0.057	0.37	0.37	1.44	0.62		0.0007	0.0054	754
B2	0.20	0.64	0.48	0.006	0.002	0.058	0.13	0.19	1.48	0.65		0.0010	0.0042	762
B3	0.17	0.55	0.60	0.020	0.008	0.058	0,31	0.38	1.44	0.62	0.030	0.0010	0.0045	755
84	0.16	0.60	0.50	0.007	0.003	0.040	0.10	0.15	1.40	0.60		0.0002	0.0045	760
B5	0.16	0.60	0.51	0.007	0.003	0.058	0.10	0.16	1.20	0.60		0.0008.	0.0049	755
B6	0.16	0.55	0.46	0.007	0.002	0.038	0.11	0.16	1.44	0.63		0.0009	0.0030	760
B7	0.16	0.60	0.51	0.007	0.003	0.040	0.11	0.15	2.00	0.60		0.0003	0.0040	774
B8	0.16	0.11	0.51	0.007	0.003	0.042	0.10	0.15	1.40	0.59		0.0003	0.0052	749
B9	0.16	0.55	0.46	0.007	0.001	0.090	0.11	0.16	1.44	0.65		0.0007	0.0060	760
B10	0.16	0.55	0.47	0.007	0.002	0.040	0.11	0.16	150	0.62		0.0007	0.0070	761

(continued)

Billet No.	Composition (mass%)													A_{C1} point (°C)
	C	Si	Mn	P	S	Al	Cu	Ni	Cr	Mo	V	B	N	
B11	0.16	0.55	0.47	0.007	0.001	0.071	0.11	0.17	1.44	0.62		0.0008	0.0025	760
B12	0.16	0.90	0.51	0.007	0.003	0.042	0.10	0.15	1.40	0.59		0.0003	0.0052	766
B13	0.16	0.50	0.80	0.007	0.002	0.055	0.10	0.15	1.44	0.62		0.0008	0.0040	754
B14	0.16	0.50	0.45	0.007	0.002	0.055	0.10	0.15	1.44	0.35		0.0008	0.0040	759
B15	0.16	0.50	0.45	0.007	0.002	0.055	0.10	0.15	1.44	0.62		0.0020	0.0040	759
A_{C1} point = 723 -14 × [Mn] + 22 × [Si] - 14.4 × [Ni] + 23.3 × [Cr] where [Mn], [Si], [Ni], and [Cr] represent the contents (by mass percent) of Mn, Si, Ni, and Cr, respectively.														

Table 2

Steel No.	Slab No.	Thickness t (mm)	Quenching temperature (°C)	Holding time at 900°C or more (minutes)	Tempering temperature (°C)	Microstructure				Grain boundary carbide		Tensile characteristics		Impact characteristics vE_{-10} (°C)	Workability evaluation
						Prior γ grain size (μm)	Faction of B + M (%)	PF fraction (%)	Microstructure size (μm)	Maximum size (μm)	Fraction (area%)	YS (MPa)	TS (MPa)		
A1-1	A1	120	930	60	730	75	100	0	19	0.6	3.1	411	579	266	○
A1-2	A1	120	930	100	730	90		0	16	0.5	3.0	405		277	○
A1-3	A1	120	930	30	730	40	100	0	25	0.8	3.0	400	580	80	○
A1-4	A1	120	930	80	620	70	100	0	16	0.6	2.9	410	585	270	○
A1-5	A1	120	930	80	750	70	100	0	16	0.6	3.0	400	590	250	○
A1-6	A1	120	930	80	780	70		15	20	0.6	3.5	330	510	90	○
A1-7	A1	120	890	0*	730	45	100	0	30	0.6	3.0	400	580	80	○
A1-8	A1	120	900	65	730	55	100	0	15	0.6	3.0	418	600	210	○
A1-9	A1	120	950	65	730	80	100	0	15	0.6	3.0	418	600	210	○
A1-10	A1	120	1000	70	730	105	100	0	28	0.6	3.0	372	540	74	○
A1-11	A1	90	930	60	730	70	100	0	13	0.6	3.0	420	600	288	○
A1-12	A1	160	930	60	730	75	100	0	20	0.6	3.0	385	575	145	○
A1-13	A1	120	930	60	Not performed	75	100	0	15	0.5	2.9	415	585	145	Δ
A2	A2	120	930	80	730	70	100	0	20	0.3	1.0	397	549	230	○
A3	A3 A4	120	930	80	730	65	100	0	15	0.6	5.0	414	579	150	○
A4	A4	120	930	80	730	70	100	0	14	0.6	3.8	414	591	205	○
A5	A5	120	930	80	730	70	100	0	18	0.6	3.9	403	573	189	○
A6	A6	120	930	80	730	80	100	0	14	0.6	2.8	414	580	190	○
A7	A7	120	930	80	730	70	100	0	16	0.6	3.0	399	568	210	○
A8	A8	120	930	80	730	65	100	0	16	0.6	3.2	402	579	180	○

(continued)

Steel No.	Slab No.	Thickness t (mm)	Quenching temperature (°C)	Holding time at 900°C or more (minutes)	Tempering temperature (°C)	Microstructure				Grain boundary carbide		Tensile characteristics		Impact characteristics vE_{-10} (°C)	Workability evaluation
						Prior γ grain size (μm)	Faction of B + M (%)	PF fraction (%)	Microstructure size (μm)	Maximum size (μm)	Fraction (area%)	YS (MPa)	TS (MPa)		
A9	A9	120	930	80	730	70	100	0	15	0.6	3.0	400	575	185	○

* Holding time at 890°C is 80 min.

Table 3

Steel No.	Slab No.	Thickness t (mm)	Quenching temperature (°C)	Holding time at 900°C or more (minutes)	Tempering temperature (°C)	Micros tructure				Grain bound dary carbide Tensile chara acteristics				Impact charac-teristics vE_{-10} (°C)	Workability evaluation
						Prior γ grain size (μm)	Fraction of B + M (%)	PF frac-tion (%)	Microstructure size (μm)	Maximum size (μm)	Fraction (area%)	YS (MPa)	TS (MPa)		
A10	A10	120	930	80	730	70	100	0	17	0.6	3.0	407	571	190	○
A11	A11	120	930	80	730	75	100	0	18	0.6	3.7	420	580	188	○
A12	A12	120	930	80	730	70	100	0	16	0.6	3.8	415	575	175	○
A13	A13	120	930	80	730	65	100	0	15	0.6	3.0	425	585	180	○
A14	A14	120	930	80	730	75	100	0	15	0.5	2.8	425	602	201	○
B1	B1	120	930	80	730	75	85	15	30	0.3	0.7	372	507	350	○
B2	B2	120	930	80	730	70	100	0	12	1.0	5.5	440	620	78	○
B3	B3	120	930	80	730	80	100	0	15	0.6	3.3	400	588	75	○
B4	B4	120	930	80	730	70	65	35	40	0.6	3.0	355	550	50	○
B5	B5	120	930	80	730	70	100	0	19	1.0	3.0	415	590	70	○
B6	B6	120	930	80	730	110	100	0	25	0.6	3.0	402	592	65	○
B7	B7	120	930	80	730	70	100	0	18	0.3	3.0	411	602	75	○
B8	B8	120	930	80	730	70	100	0	19	0.6	3.0	360	510	140	○
B9	B9	120	930	80	730	75	100	0	15	0.6	3.0	398	582	55	○
B10	B10	120	930	80	730	70	83	17	40	0.6	3.0	370	566	68	○
B11	B11	120	930	80	730	115	100	0	25	0.6	3.0	404	602	72	○
B12	B12	120	930	80	730	80	100	0	18	0.6	3.0	400	630	84	○
B13	B13	120	930	80	730	70	100	0	18	0.6	3.0	400	598	77	○
B14	B14	120	930	80	730	70	100	0	18	0.6	3.0	400	560	72	○
B15	B15	120	930	80	730	70	100	0	18	0.6	3.0	435	620	77	○

[0072] Tables 1 to 3 reveal the following. Specifically, in each of examples of the invention of A1-1, A1-2, A1-4, A1-5, A1-8, A1-9, A1-11 to A1-13, and A2 to A14, the steel member was composed of steel satisfying the defined composition, and was produced under the defined condition. Hence, the resultant steel member satisfied the specification of the microstructure, and exhibited high strength and high toughness at the thicknesswise central portion despite a large thickness of the steel member.

[0073] Comparison of A1-13 with any of other examples of the invention shows that the steel member is preferably subjected to tempering under the defined condition in order to provide good workability.

[0074] In contrast, each of example Nos. other than the above-described examples did not satisfy one of the composition and the manufacturing condition, and was therefore inferior in at least one of the tensile characteristics and the impact characteristics of the thicknesswise central portion.

[0075] Specifically, in A1-3, since the heating retention time in hardening was too short, prior austenite grain size was still small, and sufficient hardenability was not provided. As a result, the microstructure was coarsened, and toughness was degraded.

[0076] In A1-6, since the tempering temperature was too high, polygonal ferrite was formed, and the microstructure was softened. As a result, both of strength and toughness were bad.

[0077] In A1-7, since the hardening temperature was too low, prior Y grain size was still small during hardening. As a result, a fine microstructure was not formed, and good toughness was not provided.

[0078] In A1-10, since the hardening temperature was too high, some of N fixed in a form of AlN was dissolved and bonded to B, and therefore the effect of improving hardenability by free B was not provided. As a result, a fine microstructure was not formed, and toughness was degraded.

[0079] B1 to B15 are examples that each do not satisfy the defined composition as described in detail below.

[0080] In B1, since the C content was insufficient, the microstructure had neither tempered bainite nor tempered martensite, and the grain boundary carbide was not sufficiently provided, and consequently strength was insufficient. In B2, since the C content was excessive, coarse grain boundary carbide was formed, and toughness was degraded.

[0081] In B3, since the P content and the S content were each excessive, grain boundary cracking occurred, and toughness was degraded. In B4, since the B content was insufficient, hardenability was not sufficient. As a result, a fine microstructure was not formed, and toughness was degraded.

[0082] In B5, since the Cr content was insufficient, coarse grain boundary carbide was formed, and toughness was degraded. In B6, since the Al content was insufficient, the effect of suppressing coarsening of prior Y grains by AlN was not exhibited during hardening, and a fine microstructure was not formed. As a result, toughness was degraded. In B7, since the Cr content was excessive, grain boundary fracture was caused by temper embrittlement, and good toughness was not provided.

[0083] In B8, since the Si content was insufficient, high strength was not provided. In B9, since the Al content was excessive, coarse inclusions were formed, and toughness was degraded. In B10, since the N content was excessive, the effect of fixing N by Al was not exhibited, and BN was formed, and consequently the effect of improving hardenability by free B was not sufficiently exhibited. As a result, the microstructure was coarsened, and toughness was degraded.

[0084] In B11, since the N content was insufficient, the effect of suppressing coarsening of prior Y grains by AlN was not exhibited during hardening, and a fine microstructure was not formed. As a result, toughness was degraded.

[0085] B12 was excessive in Si content. B13 was excessive in Mn content. B14 was insufficient in Mo content. B15 was excessive in B content. Hence, any of them was increased in temper embrittlement sensitivity and degraded in toughness.

Claims

1. A steel member, consisting of:

C: 0.12 to 0.18%, by mass percent, the same applies to the following for the chemical components:

Si: 0.50 to 0.80%;
Mn: 0.40 to 0.70%;
P: 0.015% or less, not including 0%;
S: 0.005% or less, not including 0%;
Al: 0.040 to 0.080%;
Cu: 0.05 to 0.40%;
Ni: 0.05 to 0.40%;
Cr: 1.25 to 1.50%;
Mo: 0.45 to 0.65%;

N: 0.0030 to 0.0060%;
B: 0.0003 to 0.0010%; and optionally
V: more than 0% to 0.030%,

with the remainder consisting of Fe and inevitable impurities,
wherein a microstructure of the thicknesswise central portion of the steel member satisfies all of the following (a) to (d)

(a) the microstructure is at least one of tempered bainite and tempered martensite,
(b) the mean equivalent circle diameter of grains is 20 μm or less, each grain being surrounded by a large-angle grain boundary having a crystal misori-entation of 15° or more between two adjacent grains,
(c) the maximum size of grain boundary carbide is 0.8 μm or less, and
(d) a fraction of the grain boundary carbide is 1.0 area% or more.

2. A method of manufacturing the steel member according to claim 1, the method comprising:

performing hot rolling on a slab having a chemical composition of the steel member according to claim 1;
after the hot rolling, performing quenching under a condition of heating temperature of 900 to 950°C and holding time at the heating temperature of 60 min or more; and
after the quenching, performing welding and post-welding heat treatment.

3. The method of manufacturing the steel member according to claim 2, further comprising, after the quenching, performing tempering at a temperature of 620°C to A_{c1} point.

4. The method of manufacturing the steel member according to claim 2, wherein the post-welding heat treatment is performed at a heating temperature and for a heating time such that a value P represented by Formula (1) is 20 or more,

$$\text{Value P} = T \times (20 + \log t) \times 10^{-3} \quad \cdots \quad (1)$$

where T is heating temperature (K), and t is heating time (hr).

Patentansprüche

1. Stahlelement, bestehend aus:

C: 0,12 bis 0,18%, in Massenprozent, wobei das gleiche für die nachstehenden chemischen Komponenten gilt:

Si: 0,50 bis 0,80%;
Mn: 0,40 bis 0,70%;
P: 0,015% oder weniger, 0% nicht einschließend;
S: 0,005% oder weniger, 0% nicht einschließend;
Al: 0,040 bis 0,080%;
Cu: 0,05 bis 0,40%;
Ni: 0,05 bis 0,40%;
Cr: 1,25 bis 1,50%;
Mo: 0,45 bis 0,65%;
N: 0,0030 bis 0,0060%;
B: 0,0003 bis 0,0010%; und gegebenenfalls
V: mehr als 0% bis 0,030%,

wobei der Rest aus Fe und unvermeidlichen Verunreinigungen besteht,
wobei eine Mikrostruktur des zentralen Dickenabschnitts des Stahlelements allen der nachstehenden (a) bis (d) genügt:

(a) die Mikrostruktur ist mindestens eine von getempertem Bainit und getempertem Martensit,
(b) der mittlere äquivalente Kreisdurchmesser von Körnern beträgt 20 μm oder weniger, wobei jedes Korn von einer großwinkligen Korngrenze mit einer Kristallfehlorientierung von 15° oder mehr zwischen zwei benachbarten Körnern umgeben ist,
(c) die maximale Größe von Korngrenzencarbid beträgt 0,8 μm oder weniger und
(d) ein Anteil des Korngrenzencarbids beträgt 1,0 Flächen-% oder mehr.

2. Verfahren zur Herstellung des Stahlelements nach Anspruch 1, wobei das Verfahren umfasst:

das Durchführen von Warmwalzen auf einer Bramme mit einer chemischen Zusammensetzung des Stahlelements nach Anspruch 1;
nach dem Warmwalzen, das Durchführen von Abschrecken unter einer Bedingung einer Erwärmungstemperatur von 900 bis 950°C und einer Haltezeit bei der Erwärmungstemperatur von 60 min oder mehr; und
nach dem Abschrecken, das Durchführen von Schweißen und einer Wärmebehandlung nach dem Schweißen.

3. Verfahren zur Herstellung des Stahlelements nach Anspruch 2, weiter umfassend, nach dem Abschrecken, das Durchführen von Tempern bei einer Temperatur von 620°C bis zum A_{c1}-Punkt.

4. Verfahren zur Herstellung des Stahlelements nach Anspruch 2, wobei die Wärmebehandlung nach dem Schweißen bei einer Erwärmungstemperatur und für eine Erwärmungszeit durchgeführt wird, so dass ein Wert P, dargestellt durch Formel (1), 20 oder mehr beträgt,

$$\text{Wert P} = T \times (20 + \log t) \times 10^{-3} \quad \cdots \quad (1)$$

wobei T die Erwärmungstemperatur (K) ist und t die Erwärmungszeit (hr) ist.

Revendications

1. Élément en acier, constitué de :

C: 0,12 à 0,18%, en pourcentage en masse, la même chose s'applique dans ce qui suit pour les composants chimiques :

Si : 0,50 à 0,80 % ;
Mn : 0,40 à 0,70 % ;
P : 0,015 % ou moins, sans inclure 0 % ;
S : 0,005 % ou moins, sans inclure 0 % ;
Al : 0,040 à 0,080 % ;
Cu : 0,05 à 0,40 % ;
Ni : 0,05 à 0,40 % ;
Cr: 1,25 à 1,50 % ;
Mo : 0,45 à 0,65 % ;
N : 0,0030 à 0,0060 % ;
B : 0,0003 à 0,0010 % ; et facultativement V : plus de 0 % à 0,030 %,
le reste consistant en Fe et les impuretés inévitables,

dans lequel une microstructure de la partie centrale dans le sens de l'épaisseur de l'élément en acier satisfait tous les points (a) à (d) suivants

(a) la microstructure est au moins une parmi la bainite revenue et la martensite revenue,
(b) le diamètre moyen du cercle équivalent des grains est de 20 μm ou moins, chaque grain étant entouré d'un joint de grain à grand angle ayant une désorientation cristalline de 15° ou plus entre deux grains adjacents,
(c) la taille maximale du carbure de joint de grain est de 0,8 μm ou moins, et
(d) une fraction du carbure de joint de grain est de 1,0 % surfacique ou plus.

2. Procédé de fabrication de l'élément en acier selon la revendication 1, le procédé comprenant :

la mise en oeuvre d'un laminage à chaud sur une brame ayant une composition chimique de l'élément en acier selon la revendication 1 ;
après le laminage à chaud, la mise en oeuvre d'une trempe dans une condition de température de chauffage de 900 à 950 °C et d'une durée de maintien à la température de chauffage de 60 minutes ou plus ; et
après la trempe, la mise en oeuvre d'un soudage et d'un traitement thermique post-soudage.

3. Procédé de fabrication de l'élément en acier selon la revendication 2, comprenant en outre, après la trempe, la mise en oeuvre d'un revenu à une température de 620 °C jusqu'au point A_{c1}.

4. Procédé de fabrication de l'élément en acier selon la revendication 2, dans lequel le traitement thermique post-soudage est réalisé à une température de chauffage et pendant une durée de chauffage de façon qu'une valeur P représentée par la formule (1) soit de 20 ou plus,

$$\text{Valeur P} = T \times (20 + \log t) \times 10^{-3} \qquad (1)$$

où T est la température de chauffage (K), et t est la durée de chauffage (heure).

REFERENCES CITED IN THE DESCRIPTION

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