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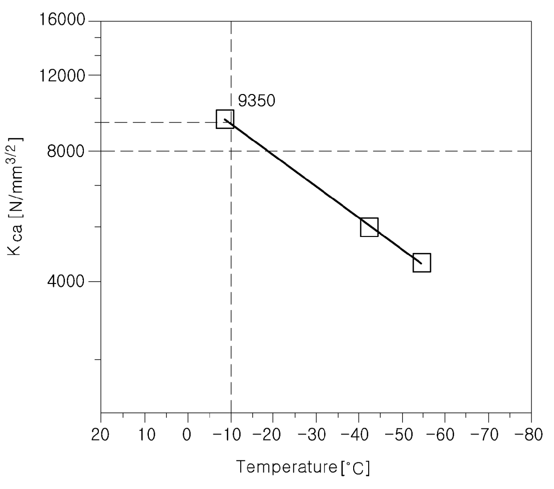
(54) HIGH-STRENGTH STEEL MATERIAL FOR POLAR REGION ENVIRONMENT HAVING EXCELLENT ANTI-FRACTURE CHARACTERISTICS AT LOW TEMPERATURES AND METHOD FOR MANUFACTURING SAME

(57) Provided are a high-strength steel material for a polar region environment having excellent anti-fracture characteristics at low temperatures and a method for manufacturing the same. A high-strength steel material having excellent anti-fracture characteristics at low temperatures according to the present invention comprises, in terms of weight%: 0.005-0.07% of C; 0.005-0.3% of Si; 1.7-3.0% of Mn; 0.001-0.035% of Sol.Al; 0.02% or less (0% excluded) of Nb; 0.01% or less (0% excluded) of V; 0.001-0.02% of Ti; 0.01-1.0% of Cu; 0.01-2.0% of Ni; 0.01-0.5% of Cr; 0.001-0.5% of Mo; 0.0002-0.005% of Ca; 0.001-0.008% of N; 0.02% or less (0% excluded) of P; 0.003% of less (0% excluded) of S; 0.003% or less (0% excluded) of O; the remainder of Fe; and unavoidable impurities. The high-strength steel material having excellent anti-fracture characteristics at low temperatures according to the present invention satisfies relation 1 and relation 2, and the microstructure thereof comprises at least 70 area% of a total of polygonal ferrite and acicular ferrite and comprises 3.5 area% or less of MA phase (martensite-austenite composite phase).

【FiG.1】



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Description

[Technical Field]

[0001] The present disclosure relates to a high-strength steel material for polar environments having excellent anti-fracture characteristics at low temperature, able to be preferably applied to shipbuilding and marine structural steels, and a method for producing the same.

[Background Art]

[0002] Due to global warming, ice in the Arctic regions continues to be decreased. Accordingly, interest in an Arctic route connecting Europe and East Asia is significantly increasing. Recently, cargo ships have experimentally run only during the summer months, and it has been reported that the time and costs were reduced by up to 30% or more, as compared with the existing route through Southeast Asia. In addition, it is expected that if Arctic ice disappears completely, a straight route through the Arctic Point will be opened within 20 to 30 years . Therefore, a need for a ship for passing through the Arctic region has been realized. Accordingly, there is an increasing need for design of a safe ship for such a polar region environment and a steel material for the polar region environment.

[0003] Existing structural steel materials are vulnerable to fracture in polar region environments, for example, environments exposed to low temperatures of down to -60°C, drifting ice, and the like. Therefore, a novel high-strength steel material having excellent anti-fracture characteristics at low temperature is required to overcome the vulnerability to fractures.

[0004] In general, a high-strength thick steel material, used in large-sized ships or oil-extraction platforms, is vulnerable to fracturing at low temperatures. This is because a high-strength ultra-thick steel material should include large amounts of alloy elements such as Mn, Mo, and the like, to secure strength and a coarse hard-phase microstructure such as granular bainite, M-A, or the like, may be easily formed due to a low rolling reduction rate and a low accelerated cooling rate during manufacturing of the ultra-thick steel material. Such a microstructure causes a steel material to be significantly vulnerable to fracture at low temperature. Accordingly, it is necessary to refine the microstructure and to significantly reduce a hard microstructure such as granular bainite, M-A, or the like, such that an ultra-thick steel plate has high strength and excellent anti-fracture characteristics at low temperature.

[0005] To address the above-mentioned issues, conventional methods have been used, such as: ① a slab reheating temperature is extremely decreased and controlled-rolled at low temperature to refine a structure, ② tempering is performed at low temperature by adding copper (Cu) in an amount of 1% or more to improve strength with a fine Cu precipitate, ③ a large amount of nickel (Ni) is added to improve low-temperature toughness against granular bainite, a hard phase, or the like, and ④ a promoting element such as carbon (C), or the like, is significantly decreased to extremely reduce an M-A structure. However, as a structure such as a ship is increased in size, and a use environment is changing to a polar region environment, it may be difficult to sufficiently secure fraction initiation and propagation resistance at low temperature by simply applying the above-mentioned methods.

[0006] Accordingly, there is a need for development of a high-strength steel material having more improved fracture initiation and propagation resistance at low temperature and a method for producing the same.

(Prior Art Document)

[0007] (Patent Document 1) Korean Patent Publication No. 10-2002-0028203

[Disclosure]

[Technical Problem]

[0008] An aspect of the present disclosure is to provide a high-strength steel having excellent fracture initiation and propagation resistance at low temperature and a method for producing the same.

[0009] On the other hand, the objective of the present disclosure is not limited to the above description. The objective of the present disclosure maybe understood from the content of the present specification. Those skilled in the art have no difficulty in understanding additional objectives of the present disclosure.

[Technical Solution]

[0010] According to an aspect of the present disclosure, a high-strength steel material having excellent anti-fracture characteristics at low temperature includes, by weight percentage (wt%): 0.005 to 0.07% of carbon (C); 0.005 to 0.3%

of silicon (Si) ; 1.7 to 3.0% of manganese (Mn) ; 0.001 to 0.035% of soluble aluminum (Sol.Al); 0.02% or less (excluding 0%) of niobium (Nb); 0.01% or less (excluding 0%) of vanadium (V); 0.001 to 0.02% of titanium (Ti); 0.01 to 1.0% of copper (Cu); 0.01 to 2.0% of nickel (Ni) ; 0.01 to 0.5% of chromium (Cr) ; 0.001 to 0.5% of molybdenum (Mo); 0.0002 to 0.005% of calcium (Ca) ; 0.001 to 0.008% of nitrogen (N); 0.02% or less (excluding 0%) of phosphorus (P); 0.003% or less (excluding 0%) of S; 0.003% or less (excluding 0%) of oxygen (O) ; a remainder of iron (Fe) ; and unavoidable impurities, and satisfies Relational Expression 1 and Relational Expression 2. A microstructure of the high-strength steel material includes 70 area% or more of a sum of polygonal ferrite and acicular ferrite and includes 3.5 area% or less of a MA(martensite-austenite constituents) (MA) phase,

Relational Expression 1:

$$Mn + 0.5 \times (Ni + Cu) \geq 2.5\text{wt\%}$$

Relational Expression 2:

$$Mo + Cr + 1.5 \times Si + 10 \times Nb \leq 0.5\text{wt\%}$$

(where in Relational Expressions 1 and 2, contents of elements are in wt%).

[0011] According to another aspect of the present disclosure, a method for producing a high-strength steel material having excellent anti-fracture characteristics at low temperature includes preparing a steel slab satisfying the above-mentioned alloy composition; heating the steel slab to a temperature within a range of 1000 to 1200°C; finishing hot rolling the heated steel slab at a temperature of 650°C or higher to obtain a finishing hot-rolled steel sheet such that a total reduction ratio in a non-recrystallization temperature range is 30% or more (excluding a reduction ratio of a recrystallization region); and cooling the finishing hot-rolled steel sheet to a cooling end temperature within a range of 200 to 550°C at a cooling rate of 2 to 30°C/sec.

[Advantageous Effects]

[0012] According to the present disclosure, a steel material having significantly improved fracture initiation and propagation resistance may be effectively provided.

[Description of Drawings]

[0013]

FIG. 1 is a graph showing a measured Kca value of a steel material in Inventive Example 1 in the present embodiment.
FIG. 2 is an image illustrating a microstructure of a steel material of Inventive Example 3 in the present embodiment.

[Best Mode for Invention]

[0014] Hereinafter, example embodiments of the present disclosure will be described. The disclosure may, however, be exemplified in many different forms and should not be construed as being limited to the specific embodiments set forth herein. Embodiments of the present disclosure are also provided to more fully describe the present disclosure to those skilled in the art.

[0015] The present inventors have repeatedly conducted research and experiments to further improve fracture initiation and propagation resistance at low temperatures, and confirmed that precise control is required in such a manner of significantly increasing the amount of an alloying element effective in improving both strength and toughness of a ferrite matrix while significantly decreasing amounts of alloying elements, in particular, C, Mo, Cr, Nb, and the like, forming a carbide. Due the above precise control, a microstructure of a steel material may include polygonal ferrite and acicular ferrite in the total of 70 area% or more, and a MA(martensitic-austenitic constituents) phase of 3.5 area% or less. Accordingly, the present inventors discovered that fracture initiation and propagation resistance at low temperatures may be significantly improved and provide the present disclosure, based on the discovery.

[0016] For example, according to the present disclosure, a high-strength steel material having excellent anti-fracture characteristics at low temperature includes, by weight percentage (wt%): 0.005 to 0.07% of carbon (C); 0.005 to 0.3% of silicon (Si) ; 1.7 to 3.0% of manganese (Mn) ; 0.001 to 0.035% of soluble aluminum (Sol.Al); 0.02% or less (excluding 0%) of niobium (Nb); 0.01% or less (excluding 0%) of vanadium (V); 0.001 to 0.02% of titanium (Ti); 0.01 to 1.0% of copper (Cu); 0.01 to 2.0% of nickel (Ni) ; 0.01 to 0.5% of chromium (Cr) ; 0.001 to 0.5% of molybdenum (Mo); 0.0002 to 0.005% of calcium (Ca) ; 0.001 to 0.008% of nitrogen (N); 0.02% or less (excluding 0%) of phosphorus (P); 0.003% or less (excluding 0%) of S; 0.003% or less (excluding 0%) of oxygen (O) ; a remainder of iron (Fe) ; and unavoidable impurities, and satisfies Relational Expression 1 and Relational Expression 2. A microstructure of the high-strength steel material includes 70 area% or more of a sum of polygonal ferrite and acicular ferrite and includes 3.5 area% or less of a MA(martensitic-austenitic constituents) phase.

[0017] Hereinafter, an alloy composition of a steel material according to the present disclosure and reasons and reasons why the contents thereof are limited will be described in detail. Hereinafter, the content of each element is in weight percentage (wt%) .

C: 0.01 to 0.07%

[0018] Carbon (C) is an element promoting formation of acicular ferrite or lath bainite and playing an important role in securing strength by forming cementite or pearlite. When the content of C is less than 0.01%, C is not almost diffused. Accordingly, since transformation occurs relatively early, a steel material may be transformed into a coarse ferrite structure to deteriorate strength and toughness thereof. On the other hand, when the content of is greater than 0.07%, cementite or an MA phase may be excessively formed, and may be coarsely formed to deteriorate fracture initiation resistance at low temperatures. Therefore, the content of C may be in the range of, in detail, 0.01 to 0.07%. The content of C may be in the range of, in more detail, 0.01 to 0.06% and, even more detail, 0.01 to 0.05%.

Si: 0.005 to 0.3%

[0019] Silicon (Si) is an element generally added for solid solution strengthening together with deoxidization and desulfurization effects. However an effect of increasing yield strength and tensile strength is negligible, and silicon may significantly deteriorate fraction initiation resistance at low temperature as stability of austenite is significantly improved in a heat-affected zone (HAZ) to increase a fraction of an MA phase, so that the content of Si may be in the range of 0.3% or less. Meanwhile, to control the content of Si to be 0.005% or less, processing time in a steelmaking process may be significantly increased to increase manufacturing costs and productivity may be reduced. Therefore, the content of Si may be in the range of 0.005 to 0.3%. The content of Si may be in the range of, in more detail, 0.005 to 0.25% and, in even more detail, 0.005 to 0.2%.

Mn: 1.7 to 3.0%

[0020] Manganese (Mn) is added in an amount of 1.7% or more to secure sufficiently high strength because a strength increasing effect, caused by solid solution strengthening, is significant and a decrease in toughness at low temperature is not significant. However, when Mn is excessively added, segregation may severely occur in the center in a thickness direction of the steel sheet, and at the same time formation of MnS, a non-metallic inclusion, may be promoted together with segregated S. A MnS inclusion, formed in the center, may be elongated by subsequent rolling, and a segregation site may significantly deteriorate fracture initiation and propagation resistance at low temperature because a low-temperature structure having high strength is easily formed due to high hardenability. For this reason, an upper limit of the content of Mn may be, in detail, 3.0%. Therefore, the content of Mn may be in the range of, in detail, 1.7 to 3.0%. The content of Mn may be in the range of, in more detail, 1.7 to 2.8%.

Sol.Al: 0.005 to 0.035%

[0021] Soluble aluminum (Sol.Al) is used as a strong deoxidizer together with Si and Mn in a steelmaking process. Such an effect may be sufficiently achieved when at least 0.005% or more of Sol.Al should be added during sole or composite deoxidation. However, when the content of Sol.Al is greater than 0.035%, the above-mentioned effect may saturated and a fraction of Al_2O_3 in an oxidative inclusion, formed as a result of deoxidation, may be increased more than necessary, such that a size of the inclusion is coarsened and the inclusion is not easily removed during refinement. Thus, low-temperature toughness of the steel material may be significantly reduced. In addition, similarly to Si, formation of the MA phase may be promoted in the heat-affected zone to significantly deteriorate fracture initiation and propagation resistance at low temperatures. Therefore, the content of Sol.Al may be in the range of 0.005 to 0.035%. The content of Sol.Al may be in the range of, in more detail, 0.005 to 0.03% and, in even more detail, 0.005 to 0.02%.

Nb: 0.02% or less (excluding 0%)

[0022] Nb is solid-solubilized in austenite during reheating of the slab to increase hardenability of the austenite, and precipitates as a fine carbonitride (Nb,Ti)(C,N) during hot rolling such that recrystallization is suppressed during rolling or cooling to make final microstructure fine. However, when Nb is added in an excessively large amount, hardenability of the heat-affected zone may be excessively increased such that formation of the MA phase is promoted to significantly deteriorate fracture initiation and propagation resistance at low temperatures. Therefore, in the present disclosure, the content of Nb may be limited to 0.02% or less (excluding 0%). The content of Nb may be, in more detail, 0.015% or less and, in even more detail, 0.012% or less.

V: 0.01% or less (excluding 0%)

[0023] Vanadium (V) is almost all solid-solubilized again when the slab is reheated, and thus, almost precipitates during cooling after rolling to improve strength. However, in a heat-affected zone, V is dissolved at high temperature to significantly increase hardenability to promote formation of an MA phase. Therefore, in the present disclosure, the content of V may be limited to 0.01% or less (excluding 0%). The content of V may be, in detail, 0.008% or less and, in more detail, 0.005% or less.

Ti: 0.001 to 0.02%

[0024] Titanium (Ti) is mainly present as a hexagonal precipitate in the form of fine TiN at high temperature or, when added with Nb or the like, forms a (Ti,Nb)(C,N) precipitate to suppress grain growth of the base material and the heat-affected zone. To sufficiently secure the above-mentioned effect, Ti may be added in an amount of, in detail, 0.001% or more. To significantly the effect, the amount of Ti may be preferably increased according to the content of N added. On the other hand, when the content of Ti is greater than 0.02%, a coarse carbonitride may be formed more than necessary to act as an initiation point of fracture cracking, and thus, may significantly deteriorate impact characteristics of the heat-affected zone. Therefore, the content of Ti may be in the range of, in detail, 0.001 to 0.02%. The content of Ti may be in the range of, in more detail, 0.001 to 0.017% and, in even more detail, 0.001 to 0.015%.

Cu: 0.01 to 1.0%

[0025] Copper (Cu) is an element solid-solubilized or precipitated to significantly improve strength without significantly deteriorating fraction initiation and propagation resistance. When the content of Cu is less than 0.01%, the above-mentioned effect is insufficient. On the other hand, when the content of Cu is greater than 1.0%, cracking may occur in a surface of a steel sheet. In addition, Cu is an expensive element to cause an increase in costs. Therefore, the content of Cu may be in the range of 0.01 to 1.0%. The content of Cu may be in the range of, in detail, 0.01 to 0.6% and, in even more detail, 0.01 to 0.4%.

Ni: 0.01 to 2.0%

[0026] Ni has little effect of increasing strength, but is effective in increasing fracture initiation and propagation resistance at low temperature . In particular, when Cu is added, Ni has an effect of suppressing surface cracking caused by selective oxidation occurring when the slab is reheated. Addition of Ni may improve toughness at low temperature even when a coarse hard structure is formed in the heat-affected zone by a high temperature and a high cooling rate. When the content of Ni is less than 0.01%, the above-mentioned effect is insufficient. Since Ni is an expensive element, an increase in costs may occur when the content of Ni is greater than 2.0%. Therefore, the content of Ni may be in the range of, in detail, 0.01 to 2.0%. The content of Ni may be in the range of, in more detail, 0.2 to 1.8% and, in even more detail, 0.3 to 1.2%.

Cr: 0.01 to 0.5%

[0027] Chromium (Cr) has a small effect of increasing yield strength and tensile strength when it is solid-solubilized, but has an effect of improving strength and toughness by allowing a fine microstructure to be formed in a thick steel plate at a low cooling rate due to high hardenability. When the content of Cr is less than 0.01%, the above-mentioned effect is insufficient. On the other hand, when the content of Cr is greater than 0.5%, costs may be increased and the low-temperature toughness of the heat-affected zone may be deteriorated. Therefore, the content of Cr may be in the range of, in detail, 0.01 to 0.5%. The content of Cr may be in the range of, in more detail, 0.01 to 0.4% and, in even more detail, 0.01 to 0.25%.

Mo: 0.01 to 0.65%

[0028] Molybdenum (Mo) is an element effective in retarding phase transformation in an accelerated cooling process to significantly increase strength and preventing toughness from being degraded by grain boundary segregation of an impurity such as P. When the content of Mo is less than 0.01%, the above-mentioned effect is insufficient. On the other hand, when the Mo content is greater than 0.65%, the formation of the MA phase may be promoted in the heat-affected zone due to high hardenability to significantly deteriorate the fracture initiation and propagation resistance at low temperatures. Therefore, the content of Mo may be in the range of, in detail, 0.01 to 0.65%. The content of Mo may be in the range of, in more detail, 0.01 to 0.5% and, in even more detail, 0.01 to 0.4%.

Ca: 0.0002 to 0.005%

[0029] When calcium (Ca) is added to molten steel during a steelmaking process after deoxidation of Al, Ca binds to S, mainly present as MnS, to suppress formation of MnS and to form spherical CaS to suppress cracking in a center of the steel material. Therefore, in the present disclosure, Ca should be added in an amount of 0.0002% or more to sufficiently form the added S into CaS. However, when Ca is excessively added, excessive Ca may bind to oxygen (O) to form a coarse and hard oxidative inclusion, which may be elongated and fractured in subsequent rolling to act as an initiation point of cracking at low temperatures. Accordingly, an upper limit of the content of Ca may be, in detail, 0.005%. Therefore, the content of Ca may be in the range of, in detail, 0.0002 to 0.005%. The content of Ca may be in the range of, in more detail, 0.0005 to 0.003% and, in even more detail, 0.0005 to 0.0025%.

N: 0.001 to 0.006%

[0030] Nitrogen (N) is an element forming a precipitate together with added Nb, Ti, and Al such that grains of steel are refined to improve strength and toughness of a base material. However, since N is present in a state of excessive atom when excessively added, it is known as the most representative element reducing low-temperature toughness by causing aging after cold deformation. In addition, N is known to promote surface cracking due to embrittlement at high temperatures when a slab is produced by continuous casting. Therefore, in the present disclosure, considering that the content of Ti is 0.001 to 0.02%, the amount of added N may be limited to a range of 0.001 to 0.006%. The content of N may be in the range of, in more detail, 0.001 to 0.005% and, in even more detail, 0.001 to 0.0045%.

P: 0.02% or less (excluding 0%)

[0031] Phosphorus (P) is an element serving to increase strength, but deteriorating low-temperature toughness. In particular, in heat-treated steel, P may cause grain boundary segregation to significantly deteriorate low-temperature toughness. Therefore, an amount of P may be preferably controlled as low as possible. However, since significant costs may be required to excessively remove P in the steelmaking process, the content of P is limited to be 0.02% or less. The content of P may be, in more detail, 0.015% or less and, in even more detail, 0.012% or less.

S: 0.003% or less (excluding 0%)

[0032] Sulfur (S) is a representative factor bonded to Mn to produce a MnS inclusion in the center of the steel plate in the thickness direction, thereby deteriorating low-temperature toughness. Therefore, the content of S may be maintained as low as possible to secure low-temperature strain aging impact properties. However, since significant costs may be required to excessively remove S, the content of S may be limited to be 0.003% or less. The content of S may be, in more detail, 0.002% or less and, in even more detail, 0.0015% or less.

O: 0.003% or less (excluding 0%)

[0033] Oxygen (O) is formed as an oxidative inclusion to be removed by adding a deoxidizer such as Si, Mn, Al, or the like, in the steelmaking process. When the amount of the added deoxidizer and an inclusion removal process are insufficient, the amount of the oxidative inclusion remaining in the molten steel may be increased and a size of the inclusion may be significantly increased. Such an unremoved coarse oxidative inclusion may remain in the form of a fractured or spherical shape therein during a rolling process in the steel material producing process, and may serve as a fracture initiation point or a cracking propagation path at low temperature. Therefore, to secure impact characteristics and CTOD characteristics at low temperatures, the coarse oxidative inclusion should be suppressed as much as possible. To this end, the content of is limited to be 0.003% or less. The content of O may be, in more detail, 0.0025% or less and, in even more detail, 0.0022% or less.

[0034] In the present disclosure, the other component of the steel is iron (Fe). However, since impurities in raw materials or manufacturing environments may be inevitably included in the steel, they cannot be excluded. For example, 5 ppm or less of boron (B), or the like, may be included. Such impurities are well-known to those of ordinary skill in the art to which the present disclosure pertains, and thus descriptions thereof will not be given in the present disclosure.

[0035] In addition, the alloy composition of the present disclosure is required to include Mn, Ni, Cu, Cr, and Nb to satisfy not only the content of each of the above-described elements but also Relational Expressions 1 and 2.

Relational Expression 1:

$$Mn + 0.5 \times (Ni + Cu) \geq 2.5wt\%$$

Relational Expression 2:

$$Mo + Cr + 1.5 \times Si + 10 \times Nb \leq 0.5wt\%$$

(where in Relational Expressions 1 and 2, contents of elements are in wt%).

[0036] In Relational Expression 1, Mn, Ni, and Cu are representative face-centered cubic metals, and are elements increasing strength through solid solution strengthening when added to a steel material and not significantly deteriorating toughness even at low temperature. The present inventors designed Relational Expression 1 in consideration of an effect of the above elements on strength and toughness of the steel. As a value of Relational Expression 1 is increased, an effect of the solid solution strengthening may be improved to increase strengths of the steel and the heat-affected zone. Therefore, to obtain sufficient strength, the value of Relational Expression 1 may be controlled to be, in detail, 2.5 or more.

[0037] Relation Expression 2 is designed considering the degree of an influence of elements promoting formation of an MA phase, a representative microstructure significantly deteriorating the toughness of the steel material and the heat-affected zone. As a value of Relational Expression 2 is increased, a fraction of the MA phase may be significantly increased. As a result, a ductile-brittle transition temperature, a low-temperature impact characteristic of the steel material, may be increased. For example, as the value of Relational Expression 2 is increased, the low-temperature toughness tends to be decreased. Therefore, to sufficiently secure the low-temperature impact characteristics, particularly, a CTOD value of the steel material, the value of Relational Expression 2 may be controlled to be, in detail, 0.5 or less. A sub-critically reheated heat-affected zone (SC-HAZ), the most important location for guaranteeing a low-temperature CTOD value of a welding zone, has a microstructure almost similar to a microstructure of the base material because a temperature during welding is lower than or equal to a two-phase region temperature. Therefore, the value of Relational Expression 2 may be controlled to be 0.5 or less to sufficiently secure the low-temperature impact characteristic and the CTOD value of the welding zone. The value of Relational Expression 2 may be, in detail, 0.48 or less and, in even more detail, 0.45 or less.

[0038] The microstructure of the steel material according to the present disclosure includes polygonal ferrite and acicular ferrite in a total amount of 70 area% or more and a MA(martensitic-austenitic constituents) phase phase in an amount of 3.5 area% or less.

[0039] The acicular ferrite is the most important and basic microstructure for not only increasing the strength due to an fine grain size effect but also preventing propagation of cracking occurring at low temperatures. Since the polygonal ferrite is coarse as compared with the acicular ferrites, the polygonal ferrite contributes relatively little to an increase in strength, but significantly contributes to suppression of cracking propagation at low temperature due to low dislocation density and high angle grain boundary thereof.

[0040] When a total of the polygonal ferrite and acicular ferrite is less than 70 area%, it maybe difficult to suppress initiation and propagation of cracking at low temperatures, and it may be difficult to secure high strength. Therefore, the total of the polygonal ferrite and the acicular ferrite may be, in detail, 70 area% or more, in more detail, 85 area% or more and, in even more detail, 90 area% or more.

[0041] In the present disclosure, the polygonal ferrite and the acicular ferrite have a ratio of high angle grain boundary in which a crystal orientation difference between grains is defined to be 15° or more. The ratio of high angle grain boundary may occupy, in detail, 40% or more of the total grain boundary. In addition, a length of the high angle grain boundary per unit area may be, in detail, 300 mm/mm^2 or more.

[0042] In addition, the MA phase does not accept deformation due to high hardness thereof, so that above the critical point, the deformation is concentrated on a peripheral soft ferrite matrix and an interface between MA phase and the peripheral soft ferrite matrix is separated or the MA phase itself is fractured to act as a cracking initiation point. Since the MA phase is the most important cause of deteriorating low-temperature fracture characteristics of the steel, it is preferably necessary to control the MA phase as low as possible, and the MA phase may be controlled to be, in detail, 3.5 area% or less.

[0043] In this case, the MA phase may have an average size of 2.5 μm or less, measured as a circle equivalent diameter. This is because stress is further concentrated when the average size of the MA phase is greater than 2.5 μm, so the MA phase is easily fractured to act as a cracking initiation point.

[0044] In addition, the polygonal ferrite and the acicular ferrite may not be work-hardened by hot rolling. For example, the polygonal ferrite and the acicular ferrite may not be elongated by hot rolling. The polygonal ferrite and the acicular ferrite may be formed after hot rolling.

[0045] The microstructure of the steel material according to the present disclosure may include not only the above-described polygonal ferrite, acicular ferrite, and MA phase but also bainitic ferrite, cementite, and the like.

[0046] The bainitic ferrite is a structure, transformed at low temperature, in which many dislocations are present, but has relatively coarse characteristics as compared with various ferrites and includes an MA phase therein. Therefore, the bainitic ferrite is vulnerable to cracking initiation and propagation, and thus, should be controlled as low as possible.

[0047] The steel material according to the present disclosure may include inclusions, each having a size of 10 μm or more, in the range of 11/cm^2 or less. The size is measured as a circle equivalent diameter. When the number of the inclusions, each having a size of 10 μm or more, is greater than 11/cm^2, the inclusions may act as a cracking initiation point at low temperature. To control such coarse inclusions, Ca or a Ca alloy may be preferably introduced in a final step of secondary refinement and bubbling and circulation with an argon (Ar) gas may be preferably performed for 3 minutes or more after the introduction of the Ca or the Ca alloy.

[0048] In addition, the steel material according to the present disclosure may have yield strength of 460 MPa or more, an impact energy value of 300 J or more at temperature of -60°C, and a CTOD value of 0.2 mm or more at temperature of -20°C. The steel material according to the present disclosure may have a tensile strength of 570 MPa or more, and may have a ductile-brittle transition temperature (DBTT) of -80°C or less.

[0049] Hereinafter, a method for manufacturing a high-strength steel material having excellent fracture initiation and propagation resistance at low temperatures according to the present disclosure will be described.

[0050] The method for manufacturing a steel material according to the present disclosure includes preparing a steel slab satisfying the above-described alloy composition, heating the steel slab to a temperature within a range of 1000 to 1200°C, finishing hot rolling the heated steel slab at a temperature of 650°C or higher to obtain a finishing hot-rolled steel sheet, and cooling the finishing hot-rolled steel sheet to a cooling end temperature within a range of 200 to 550°C at a cooling rate of 2 to 30°C/sec.

Preparing Steel Slab

[0051] A steel slab, satisfying the above-described alloy composition, is prepared.

[0052] In this case, the preparing a steel slab may preferably include introducing Ca or a Ca alloy into molten steel in a final step of secondary refinement of the molten steel and performing bubbling and circulation with an argon (Ar) gas for 3 minutes or more after the introduction of the Ca or the Ca alloy. This is aimed at controlling coarse inclusions.

Heating Steel Slab

[0053] The heat slab is heated to a temperate within a range of 1000 to 1200°C.

[0054] When the slab heating temperature is less than 1000°C, it may be difficult to solid-solubilize again a carbide or the like, formed in the steel slab during continuous casting, and a homogenization treatment of a segregated elements is insufficient. Therefore, the steel slab may be heated to a temperature of, in detail, 1000°C or more, a temperature at which 50% or more of added Nb may be solid-solubilized again.

[0055] Meanwhile, when the slab heating temperature is higher than 1200°C, an austenite grain may be grown too coarsely and may be insufficiently refined even by subsequent rolling, and thus, mechanical properties such as tensile strength, low-temperature toughness, and the like, of a steel sheet may be significantly deteriorated.

[0056] The heating temperature of the steel slab may be, in more detail, 1000 to 1160°C and, in even more detail, 1000 to 1140°C.

Hot-Rolling

[0057] The heated steel slab is finishing hot-rolled at a temperature of 650°C or higher, an initiation temperature of bainite formation, to obtain a finishing hot-rolled steel sheet.

[0058] When the finishing hot rolling temperature is less than 650°C, coarse bainite may be formed, so that the steel sheet may be work-hardened during rolling to excessively increase strength more than necessary and to significantly reduce impact toughness at low temperature. Therefore, a finish rolling temperature may be limited to, in detail, 650°C or more. This is because when the hot rolling temperature is low, coarse pro-eutectoid ferrite is formed before the hot rolling is finished, and then elongated by rolling to be work-hardened, and retained austenite remains in the form of a band and, simultaneously, is transformed into a high-density structure of an MA hard phase to deteriorate low-temperature toughness.

[0059] In addition, the hot rolling may be preferably performed in a non-recrystallization temperature range at a total reduction ratio of 30% or more (excluding a recrystallization reduction ratio), such that sufficient transformation energy may be accumulated in austenite to sufficiently form polygonal ferrite and acicular ferrite, advantageous in low-temperature toughness, during subsequent transformation and to secure a ratio and density of high angle grain boundary.

[0060] The reduction ratio may be, in more detail, 40% or more and, in even more detail, 45% or more.

Cooling

[0061] The finishing hot-rolled steel sheet is cooled.

[0062] The hot-rolled steel sheet may be cooled to a cooling end temperature within a range of, in detail, 200 to 550°C at a cooling rate of 2 to 30°C/sec. When the cooling rate is less than 2°C/sec, the cooling rate is too low to avoid coarse ferrite, pearlite, and bainite transformation ranges, and thus, strength and low-temperature toughness may be deteriorated. When the cooling rate is greater than 30°C/sec, granular bainite or martensite may be formed to increase the strength and to significantly deteriorate low-temperature toughness.

[0063] In addition, when the cooling end temperature is higher than 550°C, it may be difficult to form a microstructure such as acicular ferrite, or the like, and there may be high possibility that coarse bainite or pearlite is formed. On the other hand, when the temperature is less than 200°C, there is no disadvantage in a microstructure but productivity may be significantly reduced due to excessive time required for cooling.

[0064] The cooling end temperature may be, in more detail, 200 to 500°C and, in even more detail, 200 to 450°C.

[0065] As necessary, the method may further include performing a tempering process in which the cooled hot-rolled steel sheet is heated to a temperature within a range of 450 to 650°C and maintained for (1.3 x t + 5) minutes to (1.3 x t + 200) minutes (where t denotes a value obtained by measuring a thickness of the hot-rolled steel sheet in mm), and then the maintained hot-rolled steel sheet is cooled. This is aimed at further improve yield strength or low-temperature toughness by decomposing MA or martensite when the MA or the martensite is excessively formed, removing internal high dislocation density, and precipitating a small amount of solid-solubilized Nb, or the like.

[0066] However, when the heating temperature is less than 450°C, a ferrite matrix is insufficiently softened and embrittlement occurs due to P segregation, or the like, and thus, toughness may be deteriorated. On the other hand, when the heating temperature is higher than 650°C, grains may be rapidly recovered and grown. In addition, when the heating temperature reaches a higher temperature, grains are reversely transformed into austenite to significantly yield strength and to deteriorate low-temperature toughness.

[0067] In addition, when the maintenance time is less than (1.3 x t + 5) minutes, a structure is insufficiently homogenized. When the maintenance time is greater than (1.3 x t + 200) minutes, productivity is reduced.

[Most for Invention]

[0068] Hereinafter, the present disclosure will be described more specifically according to an example.

(EXAMPLE)

[0069] A steel material is manufactured by heating, hot rolling, and cooling a steel slab, having the composition listed in Table 1, under conditions listed in Table 2.

[0070] A microstructure of the manufactured steel material was observed, and physical properties of the microstructure were measured and listed in Table 3.

[0071] After the manufactured steel material is welded, a CTOD value (-20°C) of a subcritical heat-affected zone (SCHAZ) was measured and listed in Table 3. Since a CTOD value (-20°C) of the steel material was greater than that of the heat-affected zone, the CTOD value (-20°C) of the steel material was not additionally measured.

[0072] For the microstructure of the steel material, a cross section was polished with a mirror surface, and etched with Nital or Lepera, as desired, to measure an image for a certain area of a specimen at 100x to 5000x magnification with an optical or scanning electron microscope. A fraction of each image was measured from the measured image using an image analyzer. To obtain a statistically significant value, the measurement was repeated for the same specimen but, at the changed position, and the average value was calculated.

[0073] To observe characteristics of the formed structure in more detail, the specimen etched with Nital was subjected to electron back scatter diffraction (EBSD) measurement with a scanning electron microscope to quantitatively measure grain boundary characteristics of the manufactured steel material.

[0074] The physical properties of the steel material were measured and listed from an engineering strain-engineering stress curve obtained by a normal tensile test.

[0075] The impact energy value (-60°C) of the heat-affected zone was measured by performing a Charpy V-notch impact test.

[0076] To obtain the CTOD value (-20°C), the specimen was processed to a size of B (thickness) x B (width) x 5B (length) in a direction, perpendicular to a rolling direction, according to BS 7448 standard and fatigue cracking was inserted such that a fatigue cracking length was approximately 50% of the width of the specimen, and a CTOD test was then performed at temperature of -20°C. Here, B denotes a thickness of the manufactured steel material.

[0077] The Kca (crack arrest K) value was obtained from a K value when a temperature is -10°C after performing a test three times using an ESSO test method to obtain a graph of a crack propagation stop temperature and the K value measured in each test. A crack arrest temperature (CAT) was calculated from a value of a Nil-ductility transition temperature (NDTT) measured in an NRL test and then calculated from a transform formula of Equation 1.

Equation 1:

$$CAT=[NDTT+10^{\circ}\ C]+[\frac{\ln\sigma}{0.046}-105]+[153\cdot(B-5)^{\frac{1}{13}}-190]$$

where B denotes a thickness of a steel material.

Table 1

Cl	St	Alloy Composition (wt%)	Re	Re

		C	Si	Mn	P	S	Sol.Al	Cu	Ni	Cr	Mo	Ti	Nb	V	N	Ca		
Inventive Steel	a	0.035	0.18	2.2	0.005	0.0011	0.005	0.28	1.3	0.05	0.02	0.012	0.005	0.003	0.0035	0.002	3.0	0.39
	b	0.046	0.01	1.9	0.005	0.0008	0.02	0.31	0.99	0.03	0.29	0.012	0.011	0.003	0.0042	0.0025	2.6	0.45
	c	0.015	0.008	2.5	0.005	0.001	0.005	0.09	0.61	0.02	0.24	0.012	0.003	0.005	0.0026	0.0016	2.9	0.30
Comparative Steel	d	0.081	0.015	1.9	0.005	0.0012	0.022	0.31	1.13	0.03	0.2	0.012	0.012	0.003	0.0035	0.0028	2.6	0.37
	e	0.021	0.011	3.2	0.005	0.0013	0.015	0.03	0.25	0.03	0.02	0.012	0.007	0.003	0.0035	0.0016	3.3	0.14
	f	0.029	0.01	2.1	0.005	0.0009	0.011	0.29	0.75	0.12	0.15	0.012	0.038	0.003	0.0035	0.0011	2.6	0.67
	g	0.025	0.015	2.4	0.005	0.0012	0.005	0.12	0.65	0.23	0.35	0.012	0.018	0.003	0.0035	0.0021	2.8	0.78
	h	0.061	0.051	1.7	0.005	0.0018	0.028	0.15	0.25	0.03	0.2	0.012	0.005	0.003	0.0035	0.0018	1.9	0.36
	i	0.034	0.017	1.9	00.057	0.0008	0.008	0.22	0.53	0.33	0.19	0.007	0.012	0.004	0.0039	0.0015	2.3	0.67

* In Table 1, Relational Expression 1 is Mn + 0.5 x (Ni + Cu) and Relational Expression 2 is Mo + Cr + 1.5 x Si + 10 x Nb.

Table 2

Classification	Steel Type	Product Thickness (mm)	Slab Heating Temperature(°C)	Finish Rolling Temperature (°C)	Non-recrystallization Region Overall Reduction Rate (%)	Cooling Rate (°C/s)	Cooling End Temperature (°C)	Tempering
IE1	a	85	1060	790	45	4.0	390	x
IE2	b	100	1100	750	40	3.4	340	x
IE3	c	51	1120	800	57	6.9	280	x
IE4	b	76	1100	810	50	4.0	260	o
CE1	d	76	1110	780	48	4.5	380	x
CE2	e	51	1140	840	55	7.5	330	x
CE3	f	85	1120	805	35	3.1	280	x
CE4	g	80	1100	800	45	4.5	440	x
CE5	h	80	1100	780	40	4.3	360	x
CE6	i	100	1180	770	35	3.3	310	x
CE7	a	90	1170	820	20	3.5	270	x
CE8	c	76	1160	800	50	-	-	x

IE: Inventive Example / CE: Comparative Example

Table 3

Classification	Steel Type	Ferrite Phase Fraction (%)	High-Angle Grain Boundary Ratio (%)	High-Angle Grain Boundary Length Density (mm/mm^2)	Yield Strength (MPa)	Tensile Strength (MPa)	Impact Energy Value(-60 °C Base material):J	Kca (-10°C)	CAT (°C)	CTOD Value of HAZ (-20 °C):mm
IE1	a	78	53	390	480	620	315	9350	-16	0.82
IE2	b	73	42	320	525	679	279	-	-17	0.36
IE3	c	83	63	490	488	627	332	-	-20	1.21
IE4	b	85	50	390	533	637	322	-	-26	0.76
CE1	d	56	43	380	596	766	75	-	11	0.18
CE2	e	57	35	220	520	668	56	-	-4	0.14
CE3	f	55	33	190	540	696	90	5860	3	0.08
CE4	g	46	42	210	587	754	103	-	2	0.12
CE5	h	75	51	390	428	554	268	-	-25	0.35
CE6	i	55	38	220	471	610	85	-	-9	0.08
CE7	a	61	23	180	475	615	112	-	-2	0.15
CE8	c	86	56	480	435	545	185	-	-36	0.29

IE: Inventive Example / CE: Comparative Example

* In Table 3, ferrite type refers to a sum of polygonal ferrite and acicular ferrite.

[0078] As can be seen from Tables 1 to 3, Inventive Examples 1 to 4, satisfying both the alloy composition and the manufacturing conditions proposed in the present disclosure, have fracture toughness resistance at low temperature and a large CTOD value in a heat-affected zone in consideration of yield strength, tensile strength, impact energy value, Kca, CAT, and the like. In particular, as can be seen from FIG. 1, a Kca value measured in Inventive Example 1 represents a value significantly greater than 8000, a required value. As can be seen from FIG. 2, such excellent strength and low-temperature toughness characteristics are a result obtained from a sufficiently formed fine polygonal and acicular ferrite structure.

[0079] Meanwhile, Comparative Example 1 is a case in which the content of C was greater than a range of the present disclosure, and added C is the most powerful element that promoting granular bainite and MA. Therefore, C was excessively added to significantly reduce a fraction of ferrite, advantageous for toughness, such that strength in the base material was high but low-temperature toughness such as an impact energy value, or the like, was poor and, in particular, a CTOD value of the heat-affected zone was significantly reduced.

[0080] Comparative Example 2 is a case in which the content of added Mn was greater than a range of the present disclosure. In this case, since the content of Mn was high, probability of segregation in the center of a steel material was significantly increased. Thus, impact energy in a center in a thickness direction of a steel material was significantly deteriorated. In addition, a hard structure having partially high hardness was formed in a central segregation zone to cause pop-in fracture, and thus, a CTOD value was significantly reduced.

[0081] Comparative Example 3 is a case in which the content of Nb, generally used for strength improvement and structure refinement, is greater than a range of the present disclosure. In general, the addition of Nb refines a structure to be advantageous in increasing both strength and toughness. However, when the content of added Nb is greater than necessary, formation of polygonal and acicular ferrites advantageous for toughness is suppressed and a structure such as granular bainite, or the like, is promoted. Therefore, a ratio and density of high angle grain boundaries of 15° or more, advantageous in suppressing propagation of cracking, are significantly reduced to result in relatively easy propagation of cracking. As shown in Table 3, a Kca value measured in Comparative Example 3 was 5860, which was significantly less than 8000, a required value. In addition, formation of an M-A structure, particularly disadvantageous in low-temperature toughness, was significantly promoted in a heat-affected zone to significantly reduce CTOD.

[0082] Comparative Examples 4, 5 and 6 are cases in which a range of the content of each element satisfied a range of the present disclosure, but values of Relational Expressions 1 and 2 were out of the range of the present disclosure. In Comparative Examples 4, 5, and 6, strength was low or low-temperature toughness was significantly deteriorated.

[0083] Specifically, Comparative Example 4 is a case in which Relational expression 1, including components advantageous in improving low-temperature toughness, satisfied a range of the present disclosure but Relational Expression 2, including components reducing low-temperature toughness, was greater than the range of the present disclosure. As a result, strength was sufficiently high, but an impact energy value in the base material or a CTOD value in a heat-affected zone was poor.

[0084] Comparative Example 5 is a case in which Relational Expression 2 satisfied a range of the present disclosure, but Relational Expression 1 was out of the range of the present disclosure. Overall, the amounts of added components were insufficient to secure strength of a steel material, and thus, strength of a base material was significantly reduced.

[0085] Comparative Example 6 is a case in which both Relational Expressions 1 and 2 were out of a range of the present disclosure. For example, Comparative Example 6 is a case in which components advantageous for low-temperature toughness were insufficient, while components disadvantageous for low-temperature toughness were sufficient. As a result, all low-temperature toughness characteristic values were poor.

[0086] Comparative Example 7 is a case in which components of a steel material satisfied all ranges of the present disclosure, but were below a total rolling reduction amount of a non-recrystallization region of the present disclosure. For example, a reduction amount of the non-recrystallization region was insufficient, and thus, a fraction of ferrite inhibiting propagation of cracking in a microstructure of a steel material was also low. In addition, a ratio and density of high angle grain boundaries were significantly reduced, and thus, low-temperature toughness characteristic values were poor.

[0087] Comparative Example 8 is a case in which components of a steel material satisfied all ranges of the present disclosure, but the steel material was manufacture without applying accelerated cooling after controlled rolling in a manufacturing process of the steel material. Ferrite, advantageous in low-temperature toughness due to a low cooling rate, was sufficiently formed, but strength was significantly reduced due to coarseness.

[0088] While example embodiments have been shown and described above, it will be apparent to those skilled in the art that modifications and variations could be made without departing from the scope of the present disclosure as defined by the appended claims.

Claims

1. A high-strength steel material having excellent anti-fracture characteristics at low temperature, the high-strength

steel material comprising, by weight percentage (wt%): 0.005 to 0.07% of carbon (C) ; 0.005 to 0.3% of silicon (Si) ; 1.7 to 3.0% of manganese (Mn) ; 0.001 to 0.035% of soluble aluminum (Sol.Al) ; 0.02% or less (excluding 0%) of niobium (Nb) ; 0.01% or less (excluding 0%) of vanadium (V); 0.001 to 0.02% of titanium (Ti) ; 0.01 to 1.0% of copper (Cu) ; 0.01 to 2.0% of nickel (Ni); 0.01 to 0.5% of chromium (Cr); 0.001 to 0.5% of molybdenum (Mo); 0.0002 to 0.005% of calcium (Ca); 0.001 to 0.008% of nitrogen (N) ; 0.02% or less (excluding 0%) of phosphorus (P); 0.003% or less (excluding 0%) of S; 0.003% or less (excluding 0%) of oxygen (O); a remainder of iron (Fe); and unavoidable impurities, and satisfies Relational Expression 1 and Relational Expression 2,
wherein a microstructure of the high-strength steel material includes at least 70 area% of a sum of polygonal ferrite and acicular ferrite and includes 3.5 area% or less of a MA(martensitic-austenitic constituents) phase,

Relational Expression 1:

$$Mn + 0.5 \times (Ni + Cu) \geq 2.5wt\%$$

Relational Expression 2:

$$Mo + Cr + 1.5 \times Si + 10 \times Nb \leq 0.5wt\%$$

where in Relational Expressions 1 and 2, contents of elements are in wt%.

2. The high-strength steel material of claim 1, wherein the polygonal ferrite and the acicular ferrite have a ratio of high angle grain boundary in which a crystal orientation difference between grains is defined to be 15° or more, and the ratio of high angle grain boundary occupies 40% or more of the total grain boundary, and
a length of the high angle grain boundary per unit area is 300 mm/mm^2 or more.

3. The high-strength steel material of claim 1, wherein the steel material has yield strength of 460 MPa or more, an impact energy value of 250 J or more at temperature of -60°C, and a Kca value of 8000 N/mm$^{3/2}$ or more measured in an ESSO test, or a crack arrest temperature (CAT) less than -10°C calculated from a Nil-ductility transition temperature (NDTT) measured in an NRL test.

4. The high-strength steel material of claim 1, wherein the steel material has tensile strength of 570 MPa or more, and a ductile-brittle transition temperature (DBTT) of -80°C or less.

5. The high-strength steel material of claim 1, wherein the steel material includes inclusions, each having a size of 10 μm or more, in the range of 11/cm^2 or less.

6. A method for manufacturing a steel material having excellent anti-fracture characteristics at low temperature, the method comprising:

preparing a steel slab satisfying Relational Expression 1 and Relational Expression 2, the steel slab comprising, by weight percentage (wt%): 0.005 to 0.07% of carbon (C); 0.005 to 0.3% of silicon (Si); 1.7 to 3.0% of manganese (Mn); 0.001 to 0.035% of soluble aluminum (Sol.Al); 0.02% or less (excluding 0%) of niobium (Nb); 0.01% or less (excluding 0%) of vanadium (V); 0.001 to 0.02% of titanium (Ti); 0.01 to 1.0% of copper (Cu); 0.01 to 2.0% of nickel (Ni); 0.01 to 0.5% of chromium (Cr); 0.001 to 0.5% of molybdenum (Mo); 0.0002 to 0.005% of calcium (Ca) ; 0.001 to 0.008% of nitrogen (N) ; 0.02% or less (excluding 0%) of phosphorus (P) ; 0.003% or less (excluding 0%) of S; 0.003% or less (excluding 0%) of oxygen (O) ; a remainder of iron (Fe); and unavoidable impurities;
heating the steel slab to a temperature within a range of 1000 to 1200°C;
finishing hot rolling the heated steel slab at a temperature of 650°C or higher to obtain a finishing hot-rolled steel sheet such that a total reduction ratio in a non-recrystallization temperature range is 30% or more, excluding a reduction ratio of a recrystallization region; and
cooling the finishing hot-rolled steel sheet to a cooling end temperature within a range of 200 to 550°C at a cooling rate of 2 to 30°C/sec,

Relational Expression 1:

$$Mn + 0.5 \times (Ni + Cu) \geq 2.5wt\%$$

Relational Expression 2:

$$Mo + Cr + 1.5 \times Si + 10 \times Nb \leq 0.5wt\%$$

where in Relational Expressions 1 and 2, contents of elements are in wt%.

7. The method of claim 6, further comprising:
performing a tempering process in which the cooled hot-rolled steel sheet is cooled after being heated to a temperature within a range of 450 to 650°C and then maintained for (1.3 x t + 5) minutes to (1.3 x t + 200) minutes.

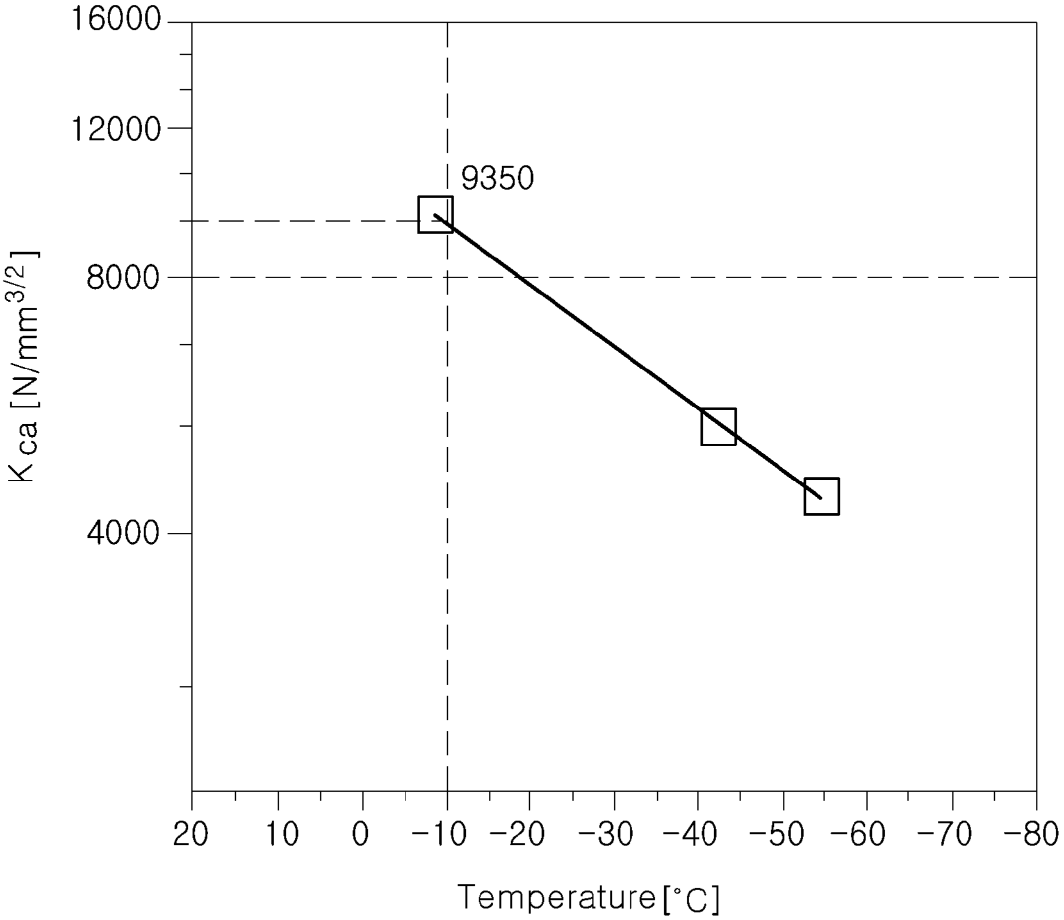
8. The method of claim 6, wherein the preparing a steel slab comprises:

introducing Ca or a Ca alloy into molten steel in a final step of secondary refinement of the molten steel; and
performing bubbling and circulation with an argon (Ar) gas for 3 minutes or more after the introduction of the Ca or the Ca alloy.

9. The method of claim 6, wherein a microstructure of the cooled steel material includes 70 area% or more of a sum of polygonal ferrite and acicular ferrite, and includes 3.5 area% or less of a MA(martensitic-austenitic constituents) phase.

10. The method of claim 9, wherein the polygonal ferrite and the acicular ferrite have a ratio of high angle grain boundary in which a crystal orientation difference between grains is defined to be 15° or more, and the ratio of high angle grain boundary occupies 40% or more of the total grain boundary, and
a length of the high angle grain boundary per unit area is 300 mm/mm^2 or more.

【FiG.1】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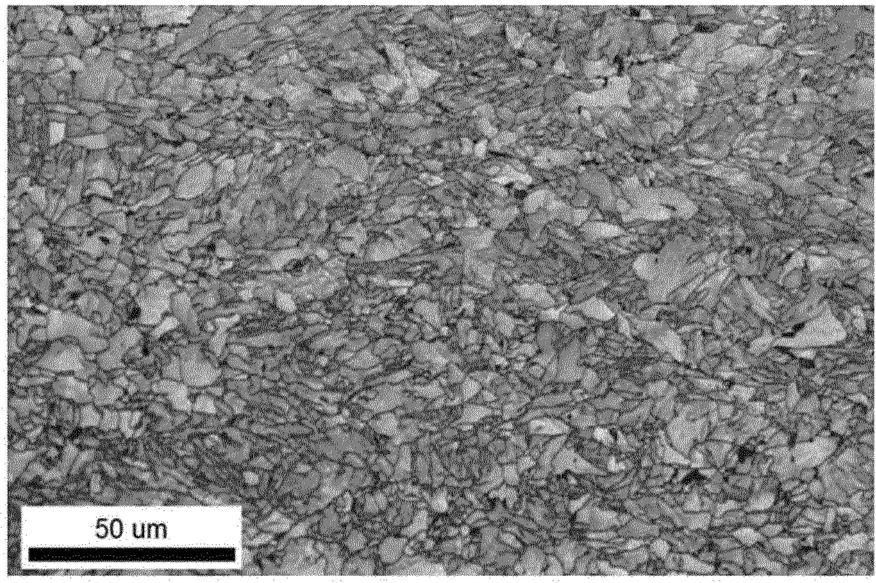
Temperature [°C]

【FiG.2】



INTERNATIONAL SEARCH REPORT

International application No.
PCT/KR2018/016155

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/58(2006.01)i, C22C 38/46(2006.01)i, C22C 38/42(2006.01)i, C22C 38/44(2006.01)i, C22C 38/50(2006.01)i, C22C 38/48(2006.01)i, C22C 38/00(2006.01)i, C21D 8/02(2006.01)i, C21D 9/46(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)

C22C 38/58; C21D 8/00; C21D 8/02; C21D 9/14; C22C 38/00; C22C 38/04; C22C 38/46; C22C 38/42; C22C 38/44; C22C 38/50; C22C 38/48; C21D 9/46

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched
Korean Utility models and applications for Utility models: IPC as above
Japanese Utility models and applications for Utility models: IPC as above

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

eKOMPASS (KIPO internal) & Keywords: nickel, copper, manganese, molybdenum, chrome, niobium, polygonal ferrite, acicular ferrite, cooling

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X	JP 2002-194488 A (NKK CORP.) 10 July 2002 See paragraphs [0040]-[0044], [0052] and claims 4-6.	1-10
A	KR 10-2016-0078714 A (POSCO) 05 July 2016 See paragraphs [0121], [0130], [0131] and claims 1, 2, 4, 9, 10.	1-10
A	JP 2015-183273 A (NIPPON STEEL & SUMITOMO METAL) 22 October 2015 See paragraphs [0058]-[0060] and claims 1, 5.	1-10
A	KR 10-2012-0074705 A (POSCO) 06 July 2012 See paragraphs [0107], [0118] and claims 1, 3-5, 8, 9.	1-10
A	KR 10-1730756 B1 (NIPPON STEEL & SUMITOMO METAL CORPORATION) 26 April 2017 See paragraphs [0150]-[0155] and claims 1, 2.	1-10

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

* Special categories of cited documents:
"A" document defining the general state of the art which is not considered to be of particular relevance
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"L" document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified)
"O" document referring to an oral disclosure, use, exhibition or other means
"P" document published prior to the international filing date but later than the priority date claimed
"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention
"X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone
"Y" document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art
"&" document member of the same patent family

Date of the actual completion of the international search 29 MARCH 2019 (29.03.2019)	Date of mailing of the international search report **29 MARCH 2019 (29.03.2019)**
Name and mailing address of the ISA/KR Korean Intellectual Property Office Government Complex Daejeon Building 4, 189, Cheongsa-ro, Seo-gu, Daejeon, 35208, Republic of Korea Facsimile No. +82-42-481-8578	Authorized officer Telephone No.

Form PCT/ISA/210 (second sheet) (January 2015)

INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
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