(19) Europäisches Patentamt
European Patent Office
Office européen des brevets

(11) EP 3 889 298 A1

(12) **EUROPEAN PATENT APPLICATION**
published in accordance with Art. 153(4) EPC

(43) Date of publication:
06.10.2021 Bulletin 2021/40

(21) Application number: **19889604.5**

(22) Date of filing: **26.11.2019**

(51) Int Cl.:
C22C 38/38 *(2006.01)* ***C22C 38/26*** *(2006.01)*
C22C 38/28 *(2006.01)* ***C22C 38/04*** *(2006.01)*
C22C 38/02 *(2006.01)* ***C22C 38/06*** *(2006.01)*
C22C 38/00 *(2006.01)* ***C21D 8/02*** *(2006.01)*
C21D 9/46 *(2006.01)*

(86) International application number:
PCT/KR2019/016299

(87) International publication number:
WO 2020/111702 (04.06.2020 Gazette 2020/23)

(84) Designated Contracting States:
AL AT BE BG CH CY CZ DE DK EE ES FI FR GB GR HR HU IE IS IT LI LT LU LV MC MK MT NL NO PL PT RO RS SE SI SK SM TR
Designated Extension States:
BA ME
Designated Validation States:
KH MA MD TN

(30) Priority: **26.11.2018 KR 20180147124**

(71) Applicant: **POSCO**
Pohang-si, Gyeongsangbuk-do 37859 (KR)

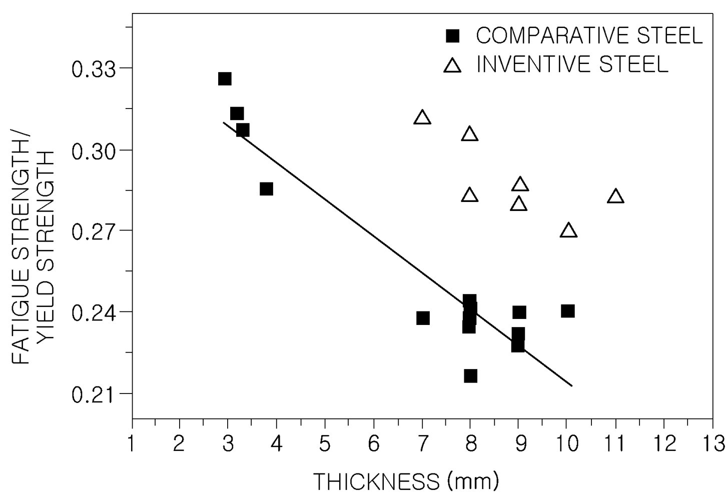
(72) Inventors:
- **KIM, Sung-Il**
  Gwangyang-si, Jeollanam-do 57807 (KR)
- **NA, Hyun-Taek**
  Gwangyang-si, Jeollanam-do 57807 (KR)

(74) Representative: **Zech, Stefan Markus**
Meissner Bolte Patentanwälte
Rechtsanwälte Partnerschaft mbB
Postfach 86 06 24
81633 München (DE)

(54) **HIGH-STRENGTH STEEL WITH EXCELLENT DURABILITY AND METHOD FOR MANUFACTURING SAME**

(57) The present invention relates to steels used for members of chassis parts and wheel discs of commercial vehicles, or the like and, more specifically, to a high-strength steel with excellent durability and a method for manufacturing same.

【FIG. 1】



EP 3 889 298 A1

Description

[Technical Field]

[0001] The present disclosure relates to a high-strength steel material having excellent durability and a method for manufacturing the same.

[Background Art]

[0002] Conventionally, members of chassis parts and wheel discs of commercial vehicles have used high-strength steel plates having a thickness of 5 mm or more and a yield strength of 450 to 600 MPa in order to secure high rigidity due to characteristics of the vehicles. However, in order to achieve a reduction in weight and an increase in strength of the vehicles, high-strength steel materials having a tensile strength of 650 MPa or more are currently used therein.

[0003] When manufacturing parts using such high-strength steel materials, such a part goes through manufacturing by pressing sheet materials which have had shear molding and punch molding performed thereon. In performing the shear molding and punch molding, there may be a disadvantage that micro-cracks may be created in sheared parts of steel plates, to shorten durability and lifespan of a final product (part).

[0004] As a solution for solving such a problem, Patent Document 1 has proposed that conventional hot-rolling be performed in an austenite region and coiling then be performed at a high temperature, to form a ferrite phase as matrix structure and a precipitate as a fine structure. In addition, Patent Document 2 has proposed that a technique in which a coiling temperature is cooled to a temperature at which a bainite phase is formed as a matrix structure so as not to generate a coarse pearlite structure, and coiling is then performed. Patent Document 3 discloses a technique of miniaturizing austenite grains by using titanium (Ti), niobium (Nb), and the like to roll at a reduction ratio of 40% or more in a non-recrystallized region during hot-rolling.

[0005] For manufacturing high-strength steel, alloy components such as Si, Mn, Al, Mo, and Cr may be mainly used. In this case, although strength of a hot-rolled steel sheet may be effectively improved, when a large amount of alloy components are added, some components may be segregated in the steel or may cause microstructure non-uniformity to deteriorate shear formability, and microcracks generated in a sheared surface may be easily propagated under a fatigue environment, to break parts.

[0006] In particular, as a thickness of the steel material increases, microstructure non-uniformity between a surface portion and a central portion in a thickness direction may increase, and occurrence of cracks in the sheared surface may increase, and a propagation speed of the cracks may increase under a fatigue environment, to deteriorate durability.

[0007] However, the preceding technologies (Patent Documents 1 to 3) do not take into account the fatigue characteristics of thick steel materials having high strength.

[0008] In addition, in order to miniaturize crystal grains of the thick steel materials and obtain a precipitation strengthening effect, in using precipitate-forming elements such as Ti, Nb, V, or the like, when it is coiled at a high temperature of about 500 to 700°C in which precipitates are easily formed, or a cooling rate of the steel sheet is not controlled during cooling after hot-rolling, coarse carbides may be formed in a central portion of the thick steel materials in a thickness direction, to deteriorate quality of the sheared surface. In addition, applying a large pressure drop of 40% in a non-recrystallized region during hot-rolling may deteriorate shape quality of a rolled plate and may cause a load on equipment, which may be difficult to apply in practice.

(Patent Document 1) Japanese Laid-Open Patent Publication No. 2002-322541
(Patent Document 2) Korean Registered Patent No. 10-1528084
(Patent Document 3) Japanese Laid-Open Patent Publication No. 1997-143570

[Disclosure]

[Technical Problem]

[0009] An aspect of the present disclosure is to provide a steel material having a certain thickness, high strength, and excellent durability, as a thick steel material, and a method of manufacturing the same.

[0010] The technical problem of the present disclosure is not limited to the aforementioned matters. Those of ordinary skill in the art to which the present disclosure pertains will not have any difficulty in understanding the additional problems of the present disclosure from the contents described in the disclosure of the present disclosure.

[Technical Solution]

[0011] According to an aspect of the present disclosure, a high-strength steel material having excellent durability, includes, by weight, carbon (C): 0.05 to 0.15%, silicon (Si): 0.01 to 1.0%, manganese (Mn): 1.0 to 2.3%, aluminum (Al): 0.01 to 0.1%, chromium (Cr): 0.005 to 1.0%, phosphorus (P): 0.001 to 0.05%, sulfur (S): 0.001 to 0.01%, nitrogen (N): 0.001 to 0.01%, niobium (Nb): 0.005 to 0.07%, titanium (Ti): 0.005 to 0.11%, a balance of Fe, and other inevitable impurities,

[0012] wherein a sum of a fraction of a ferrite phase and a fraction of a bainite phase in a microstructure is 90% or more, and a fraction of a crystal grain, in which an aspect ratio (a ratio of short side/long side) of the crystal grain in a central portion (a portion ranging from a t/4 point to a t/2 point in a thickness direction) is 0.3 or less, is less than 50%, and a length of a grain boundary observed in a unit area (1 mm^2) in the central portion is 700 mm or more.

[0013] According to another aspect of the present disclosure, a method of manufacturing a high-strength steel material having excellent durability includes heating a steel slab satisfying the above-described alloy composition in a temperature within a range of 1200 to 1350°C; hot-rolling the heated steel slab to prepare a hot-rolled steel sheet; cooling the hot-rolled steel sheet to a temperature within a range of 400 to 500°C and then coiling (CT); and air-cooling to a temperature within a range of room temperature to 200°C after the coiling,

[0014] wherein the hot-rolling includes finish hot-rolling performed at a temperature (FDT (°C)) satisfying the following Relationship 1, and the cooling includes first cooling and second cooling, wherein the first cooling is performed at a cooling rate (CR_1) satisfying the following Relationship 2, and the second cooling is performed at a cooling rate (CR_2) satisfying the following Relationship 3:

[Relationship 1]

$$Tn-50 \leq \text{FDT (hot-rolling end temperature (°C))} \leq Tn$$

where Tn = 730 + 92×[C] + 70×[Mn] + 45×[Cr] + 650×[Nb] + 410×[Ti] - 80×[Si] - 1.4×(t-5) (where, an element refers to wt% of the element, and t refers to a thickness (mm) of the final hot-rolled steel sheet)

[Relationship 2]

$$CR_1 \geq 196 - 300\times[C] + 4.5\times[Si] - 71.8\times[Mn] - 59.6\times[Cr] + 187\times[Ti] + 852\times[Nb]$$

where an element refers to wt% of the element

[Relationship 3]

$$CR_{Min} \leq CR_2 \leq CR_{Max}$$

where CR_{Max} = 76.6 - 157×[C] - 25.2×[Si] - 14.1×[Mn] - 27.3×[Cr] + 61×[Ti] + 448×[Nb], CR_{Min}= 27.4 - 45.3×[C] + 5.28×[Si] - 11×[Mn] - 7.33×[Cr] + 42.3×[Ti] + 82×[Nb] (where, an element refers to wt% of the element).

[Advantageous Effects]

[0015] According to an aspect of the present disclosure, it may be possible to provide a thick steel material excellent in cross-section quality during forming with high strength, to excellently secure a ratio of fatigue limit and yield strength of the steel material after forming.

[0016] The steel material of the present disclosure has an effect of being suitably applied to members of chassis parts and wheel discs of vehicles, or the like.

[Description of Drawings]

[0017] FIG. 1 is a graph illustrating ratios of fatigue strength and yield strength of Inventive Steel and Comparative Steel, depending on a thickness, in an embodiment of the present disclosure.

[Best Mode for Invention]

[0018] The present inventors have conducted intensive research to fundamentally solve a problem of lowering durability during forming of existing thick steel materials for vehicles.

[0019] In particular, the present inventors have investigated changes in crack distribution and durability in a sheared surface after forming, according to a composition and a microstructure of existing thick steel materials, and have found that the durability were changed according to shape control of crystal grains in a central portion of the steel materials in a thickness direction.

[0020] As a result, the present inventors have confirmed that a steel material excellent in cross-section quality during forming with high strength to have a target durability may be provided, and the present disclosure has been then accomplished.

[0021] Hereinafter, the present disclosure will be described in detail.

[0022] A high-strength steel material having excellent durability, according to an aspect of the present disclosure, may include, by weight, carbon (C): 0.05 to 0.15%, silicon (Si) : 0.01 to 1.0%, manganese (Mn): 1.0 to 2.3%, aluminum (Al) : 0.01 to 0.1%, chromium (Cr): 0.005 to 1.0%, phosphorus (P): 0.001 to 0.05%, sulfur (S): 0.001 to 0.01%, nitrogen (N): 0.001 to 0.01%, niobium (Nb): 0.005 to 0.07%, titanium (Ti): 0.005 to 0.11%.

[0023] Hereinafter, the reason for limiting the alloy composition of the hot-rolled steel sheet provided by the present disclosure as described above will be described in detail.

[0024] In this case, unless otherwise specified, an amount of each component means weight %, and a ratio of a structure is based on an area.

Carbon (C): 0.05 to 0.15%

[0025] Carbon (C) may be the most economical and effective element for strengthening steel. When an amount thereof increases, a precipitation strengthening effect may increase or a fraction of a bainite phase may increase, to improve tensile strength. In addition, as a thickness of a hot-rolled steel material increases, a cooling rate in a central portion in a thickness direction during cooling after hot-rolling becomes slower, and when an amount of C is large, coarse carbide or pearlite may be likely to be formed.

[0026] In the present disclosure, when an amount of C is less than 0.05%, it may be difficult to obtain a sufficient strengthening effect of the steel. When an amount of C exceeds 0.15%, there may be a problem that a pearlite phase or coarse carbide is formed in the central portion in a thickness direction to deteriorate shear formability and durability.

[0027] Therefore, in the present disclosure, C may be included in an amount of 0.05 to 0.15%, and, more advantageously, may be included in an amount of 0.06 to 0.12%.

Silicon (Si): 0.01 to 1.0%

[0028] Silicon (Si) may be advantageous in deoxidizing molten steel, having a solid solution strengthening effect, and delaying formation of coarse carbides to improve formability.

[0029] When an amount of Si is less than 0.01%, the solid solution strengthening effect may be small, the effect of delaying the formation of carbides may be also low, and it may be difficult to improve the formability. When an amount of Si exceeds 1.0%, red scale may be formed on a surface of the steel sheet during hot-rolling, to greatly deteriorate surface quality of the steel sheet, as well as ductility and weldability.

[0030] Therefore, in the present disclosure, Si may be included in an amount of 0.01 to 1.0%, and, more advantageously, may be included in an amount of 0.2 to 0.7%.

Manganese (Mn): 1.0 to 2.3%

[0031] Similarly to Si, manganese (Mn) may be an element, effective for solid solution strengthening of steel, and may increase hardenability of steel to facilitate formation of a bainite phase during cooling after hot-rolling.

[0032] When an amount of Mn is less than 1.0%, the above-described effects may not be sufficiently obtained. When an amount of Mn exceeds 2.3%, hardenability may increase significantly and martensite phase transformation may be likely to occur. In addition, in a continuous casting process, a segregation portion may be greatly developed in a central portion in a thickness direction, when casting a slab. Further, a microstructure may be formed unevenly in a thickness direction during cooling after hot-rolling, to deteriorate shear formability and durability.

[0033] Therefore, in the present disclosure, Mn may be included in an amount of 1.0 to 2.3%, and, more advantageously, may be included in an amount of 1.1 to 2.0%.

Aluminum (Al): 0.01 to 0.1%

[0034] Aluminum (Al) may be an element mainly added for deoxidation. When an amount thereof is less than 0.01%, an effect of the addition may not be sufficiently obtained. When an amount thereof exceeds 0.1%, it may be easy to cause corner cracks in a slab during a continuous casting process by combining with nitrogen (N) in steel to form AlN, and there may be a risk of generating defects due to formation of inclusions.
[0035] Therefore, in the present disclosure, Al may be included in an amount of 0.01 to 0.1%.
[0036] In the present disclosure, aluminum may refer to soluble aluminum (Sol. Al).

Chromium (Cr): 0.005 to 1.0%

[0037] Chromium (Cr) may plays a role in helping solid-solution strengthening of steel, and formation of bainite at a coiling temperature by delaying transformation of a ferrite phase during cooling. In order to obtain the above-described effects, Cr may be included in an amount of 0.005% or more. When an amount thereof exceeds 1.0%, the ferrite transformation may be excessively delayed and a martensite phase may be formed, to deteriorate elongation. In addition, similarly to Mn, a segregation portion may be largely developed in a central portion in a thickness direction. Further, a microstructure may be formed unevenly in a thickness direction, to deteriorate shear formability and durability.
[0038] Therefore, in the present disclosure, Cr may be included in an amount of 0.005 to 1.0%, and, more advantageously, may be included in an amount of 0.3 to 0.9%.

Phosphorus (P): 0.001 to 0.05%

[0039] Phosphorus (P) may be an element simultaneously promoting solid solution strengthening and ferrite transformation. In order to manufacture such a P in an amount of less than 0.001%, manufacturing costs may be excessive, which may be economically disadvantageous, and it may be difficult to secure a target level of strength. When an amount of P exceeds 0.05%, brittleness may occur due to grain boundary segregation, and fine cracks tend to occur during forming, to greatly deteriorate shear formability and durability.
[0040] Therefore, in the present disclosure, P may be included in an amount of 0.001 to 0.05%.

Sulfur (S): 0.001 to 0.01%

[0041] Sulfur (S) may be an impurity present in steel. When an amount thereof exceeds 0.01%, it may combine with Mn to form non-metallic inclusions. Therefore, there may be a problem that fine cracks are likely to occur during cutting processing of the steel, to greatly deteriorate shear formability and durability. In order to manufacture to include S in an amount of less than 0.001%, an excessive amount of time may be consumed during a steelmaking operation, lowering productivity.
[0042] Therefore, in the present disclosure, S may be included in an amount of 0.001 to 0.01%.

Nitrogen (N): 0.001 to 0.01%

[0043] Nitrogen (N) may be a representative solid solution strengthening element along with C, and may combine with Ti, Al, or the like to form coarse precipitates. In general, the solid solution strengthening effect of N may be superior to that of carbon, but there may be a problem that toughness of steel decreases as an amount of N in the steel increases. In consideration of the above, N may be included in an amount of 0.01% or less. In order to manufacture to include N in an amount of less than 0.001%, it may take a large of time during a steelmaking process, to lower productivity.
[0044] Therefore, in the present disclosure, N may be included in an amount of 0.001 to 0.01%.

Niobium (Nb): 0.005 to 0.07%

[0045] Niobium (Nb) may be a precipitation strengthening element, and may be effective in improving strength and impact toughness of steel due to an effect of refining grains by precipitation during hot-rolling and delayed recrystallization. In order to sufficiently obtain the above-described effect, it may be contained in an amount of 0.005% or more. When an amount thereof exceeds 0.07%, formability and durability may be deteriorated due to formation of elongated crystal grains and formation of coarse complex precipitates by excessive recrystallization delay during hot-rolling.
[0046] Therefore, in the present disclosure, Nb may be included in an amount of 0.005 to 0.07%, and, more advantageously, may be included in an amount of 0.01 to 0.06%.

Titanium (Ti): 0.005 to 0.11%

[0047] Titanium (Ti) may be a representative precipitation strengthening element along with Nb, and may have strong affinity with N to form coarse TiN in steel. The TiN may have an effect of suppressing growth of crystal grains during a heating process for hot-rolling. In addition, remaining Ti after reacting with N may be dissolved in the steel to form TiC precipitates by bonding with carbon, which may be useful for improving strength of the steel.
[0048] In order to sufficiently obtain the above-described effect, it is necessary to contain more than 0.005% of Ti. When an amount thereof exceeds 0.11%, there may be a problem that collision resistance property may be deteriorated during forming due to generation of coarse TiN and coarsening of precipitates.
[0049] Therefore, in the present disclosure, Ti may be included in an amount of 0.005 to 0.11%, and, more advantageously, may be included in an amount of 0.01 to 0.1%.
[0050] The remainder of the present disclosure may be iron (Fe). In the conventional steel manufacturing process, since impurities which are not intended from raw materials or the surrounding environment may be inevitably incorporated, the impurities may not be excluded. All of these impurities are not specifically mentioned in this specification, as they are known to anyone of ordinary skill in the manufacturing process.
[0051] The steel material of the present disclosure having the above-described alloy composition may be comprised of a composite structure of a ferrite phase and a bainite phase in a microstructure.
[0052] In this case, a sum of a fraction of the ferrite phase and a fraction of the bainite phase may be 90% or more, by area, of which the bainite phase may have, by area fraction, 50% or more.
[0053] When a fraction of the bainite phase may be less than 50%, by area, it may be difficult to secure the target strength, and in this case, when a coarse ferrite phase increases, a non-uniform microstructure may be generated, such that fine cracks during shear deformation or punching deformation may easily occur.
[0054] In this case, the ferrite phase may refer to a polygonal ferrite phase, a high-temperature ferrite phase, and the bainite phase may refer to both an acicular ferrite phase and a bainitic ferrite phase, a low-temperature ferrite phase.
[0055] A remainder structure excluding the composite structure may include an MA phase (a mixture of martensite and austenite) and a martensite phase. In this case, the two phases may be combined and included in an area fraction of 1 to 10%, of which the MA phase may be included in an amount of less than 3%.
[0056] When combined fractions of the MA phase and the martensite phase exceed 10%, tensile strength may increase, whereas a hardness value may be higher than that of the surrounding structure phase to occur cracks at a interface of the MA phase and the martensite phase during shear deformation or punching deformation, to deteriorate fatigue characteristics. In particular, the MA phase has an average size of 1/10 compared to the martensite phase, but crack generation tendency at the phase interface may be similar to that of the martensite phase, to increase a propagation speed when exposed to a fatigue environment. Therefore, the MA phase may be included in an area fraction of less than 3%.
[0057] In this manner, fractions of the coarse MA phase and the martensite phase in a matrix structure may be minimized to obtain an effect of eliminating structure non-uniformity.
[0058] Even when a steel material of the present disclosure contains 3% or less (including 0%) of a pearlite phase in addition to the above-described structure, there may be no great difficulty in securing intended properties.
[0059] In particular, in a steel material of the present disclosure, a fraction of a crystal grain in which an aspect ratio (a ratio of short side length (short axis)/long side length (long axis)) of the crystal grain in a central portion, corresponding to a portion ranging a t/4 point to a t/2 point in a thickness direction, is 0.3 or less may be less than 50%, and a length of a grain boundary observed in a unit area (1 mm^2) in the central portion may be 700 mm or more.
[0060] In a case that a fraction of a crystal grain in which an aspect ratio of the crystal grain in a central portion is 0.3 or less is 50% or more, when a crack occurs, growth of the crack may be facilitated to deteriorate durability. In addition, when a length of a grain boundary in the central portion is less than 700 mm, strength of the central portion may decrease and a crack may easily propagate, to deteriorate durability.
[0061] Methods of analyzing an aspect ratio of crystal grains and a length of a grain boundary are not particularly limited, but may be analyzed using electron back scattered diffraction (EBSD), for example. Specifically, in measurement results of rolled cross-section by EBSD, for a crystal grain having a large tilt angle grain boundary of 15° or more, it may be calculated as a length of a grain boundary observed in a unit area (1 mm^2), and an aspect ratio may be calculated as a ratio of a short axis and a long axis of a grain size.
[0062] A steel material of the present disclosure having the above-described alloy composition and microstructure may be a thick steel material having a thickness of 5 to 12 mm, a tensile strength thereof may be 650 MPa or more, and a ratio of fatigue limit and yield strength (fatigue limit/yield strength) thereof may be 0.25 or more. As a result, excellent durability as well as high strength may be secured.
[0063] Hereinafter, a method of manufacturing a high-strength steel material having excellent durability, which may be another aspect of the present disclosure, will be described in detail.
[0064] A high-strength steel material according to the present disclosure may be manufactured by performing a series

of processes of [heating - hot-rolling - coiling - cooling] of a steel slab satisfying the alloy composition proposed in the present disclosure.

[0065] Hereinafter, each of the above process conditions will be described in detail.

Heating of Steel Slab

[0066] In the present disclosure, prior to hot-rolling, a steel slab may be heated to perform a homogenization treatment. In this case, the heating may be performed at 1200 to 1350°C.

[0067] When a temperature of the heating is less than 1200°C, a precipitate may not be sufficiently re-dissolved, to decrease formation of a precipitate in a process after hot-rolling, and there may be a problem that coarse TiN remains. When a temperature of the heating exceeds 1350°C, strength may decrease due to abnormal grain growth of austenite grains, which is not allowable.

Hot-Rolling

[0068] The reheated steel slab may be hot-rolled to produce a hot-rolled steel sheet, and, in this case, performed at a temperature within a range of 800 to 1150°C, and finish hot-rolling may be performed under conditions satisfying the following Relationship 1.

[0069] When the hot-rolling is performed at a temperature higher than 1150°C, a temperature of the hot-rolled steel sheet increases, to have a coarse grain size and poor surface quality of the hot-rolled steel sheet. When the hot-rolling is performed at a temperature lower than 800°C, elongated crystal grains may develop due to excessive recrystallization delay, to increase severe anisotropy, deteriorate formability, and severely develop a non-uniform microstructure due to rolling at a temperature below a temperature within a range of austenite.

[0070] Particularly, in the hot-rolling of the present disclosure, when rolling ends at a temperature higher than the temperature range proposed in the following Relationship 1 (a temperature exceeding Tn), a microstructure of steel may be coarse and non-uniform, and phase transformation may be delayed. Therefore, excessive microcracks may be formed during shear forming and punch forming due to formation of a coarse MA phase and a martensite phase, to deteriorate durability. When rolling ends at a temperature lower than the temperature range suggested by the following Relationship 1 (a temperature less than Tn-50), transformation of a ferrite phase may be promoted to increase a phase fraction of fine ferrite at a t/4 point in a thickness direction directly under a surface layer in which a temperature is relatively low, in a thick steel material having a thickness of 5 mm or more. In this case, a crystal grain shape may be elongated, which causes rapid propagation of cracks, and ununiform microstructure may remain in a central portion in a thickness direction, which may be disadvantageous in securing durability.

[Relationship 1]

$$Tn-50 \leq \text{FDT (hot-rolling end temperature (°C))} \leq Tn$$

where Tn = 730 + 92×[C] + 70×[Mn] + 45×[Cr] + 650× [Nb] + 410× [Ti] - 80×[Si] - 1.4× (t-5) (where, an element refers to wt% of the element, and t refers to a thickness (mm) of the final hot-rolled steel sheet)

Cooling and Coiling

[0071] As described above, the hot-rolled steel sheet manufactured by the hot-rolling may be cooled to a temperature within a range of 400 to 500°C, and then coiling may be performed at that temperature.

[0072] The cooling may include first cooling and second cooling, the first cooling may be performed at a cooling rate (CR_1) satisfying the following Relationship 2, and the second cooling may be performed at a cooling rate (CR_2) satisfying the following Relationship 3.

[0073] Specifically, the first cooling may end in a temperature within a range at which phase transformation of ferrite occurs during cooling. A temperature at which the phase transformation of ferrite occurs may be changed according to the alloy composition proposed in the present disclosure. More specifically, the first cooling may be carried out to a temperature at which transformation of a hard phase such as a bainite phase, an MA phase, and a martensite phase does not occur. Even more preferably, the first cooling may be performed until a temperature of the hot-rolled steel sheet obtained by hot-rolling reaches 600°C.

[0074] In a case in which a thickness of a rolled plate is 5 mm or more as in the present disclosure during the first cooling in the temperature section, since a cooling rate in a central portion of the rolled plate in a thickness direction may be slower than a cooling rate in a region of a subsurface to the t/4 point, a coarse ferrite phase in the central portion in a thickness direction may be formed to have a non-uniform microstructure.

[0075] Therefore, in the present disclosure, it may be cooled at a cooling rate faster than the specific cooling rate (CR_1) represented by the following Relationship 2, such that excessive ferrite phases are not formed or ferrite phases are not coarsened during the first cooling.

[Relationship 2]

$$CR_1 \geq 196 - 300\times[C] + 4.5\times[Si] - 71.8\times[Mn] - 59.6\times[Cr] + 187\times[Ti] + 852\times[Nb]$$

where an element refers to wt% of the element.

[0076] The second cooling may be performed immediately after completion of the first cooling under the above-described conditions, and the second cooling may end at a coiling temperature (CT (°C)).

[0077] During the second cooling in the temperature section, in order to transform a non-transformed phase into a bainite phase over an entire thickness of a steel material to form 90% (area fraction) of a matrix structure into ferrite and bainite phases, the cooling may be performed at a specific cooling rate (CR_2) indicated by the following Relationship 3. In this case, when a cooling rate is slower than CR_{Min}, carbides may be formed and grown to be coarse, as compared to a bainite phase, and the coarse carbides may be mainly present at a grain boundary of a ferrite phase, and when a cooling rate is further slower, there may be a problem that a pearlite phase may be formed, to be easy to form a crack during shear forming and punch forming, and the crack may propagate along the grain boundary even with a small external force. When a cooling rate exceeds CR_{Max}, an MA phase or a martensite phase increasing a difference in hardness between phases may be excessively formed to deteriorate durability.

[0078] Therefore, it is necessary to perform cooling at a cooling rate satisfying the following Relationship 3 during the second cooling in the temperature section.

[Relationship 3]

$$CR_{Min} \leq CR_2 \leq CR_{Max}$$

where CR_{Max}= 76.6 - 157× [C] - 25.2×[Si] - 14.1× - 27.3×[Cr] + 61×[Ti] + 448× [Nb], CR_{Min}= 27.4 - 45.3× [C] + 5.28×[Si] - 11×[Mn] - 7.33×[Cr] + 42.3×[Ti] + 82×[Nb] (where, an element refers to wt% of the element).

[0079] When a coiling temperature exceeds 500°C during coiling after the above-described cooling process is completed, a pearlite phase may be formed and strength of steel may be insufficient. When a coiling temperature is less than 400°C, a martensite phase may be excessively formed to deteriorate shear formability, punch formability, and durability.

[0080] The present disclosure may control process conditions to satisfy the above-described Relationships 1 to 3 in manufacturing an intended steel material, such that an area fraction of a crystal grain in which an aspect ratio of the crystal grain formed in a central portion of the steel material in a thickness direction is 0.3 or less is secured to be less than 50%, and a length of a grain boundary observed in a unit area (1 mm^2) in the central portion is secured to be 700 mm or more.

[0081] In manufacturing a thick steel material having a thickness of 5 mm or more, it may be difficult to uniformly secure a microstructure in a central portion in a thickness direction, when performing by conventional hot-rolling. In particular, when hot-rolling is performed at an excessively low temperature to obtain delay of recrystallization in a central portion in a thickness direction, deformed structure may be strongly developed in a region from directly below a surface layer to a t/4 point of the rolled plate in the thickness direction. From this, due to an increase in non-uniformity of phase with the central portion in the thickness direction, micro-cracks tend to occur at a non-uniform part during shear deformation or punching deformation, and durability of the part may be also deteriorated. Therefore, as shown in the above Relationship 1, it is necessary to complete hot-rolling between Tn (°C) and Tn-50 (°C), where Tn may be a temperature at which the delay of recrystallization starts.

[0082] When hot-rolling ends at a temperature higher than the temperature suggested in the above Relationship 1, a coarse ferrite phase and a coarse polygonal ferrite phase may be formed, to greatly decrease an area fraction of a crystal grain having an aspect ratio of 0.3 or less. Meanwhile, there may be a risk of significantly reducing a size of a grain boundary to deteriorate strength of a central portion, and there may be a problem of facilitating growth of a crack when the crack is formed. In addition, when hot-rolling ends at a temperature lower than the temperature suggested in the above Relationship 1, severely elongated grains may increase to significantly increase an area fraction of a crystal grain having an aspect ratio of 0.3 or less, and coarse carbide or a martensite phase may be formed at a grain boundary to easily propagate a crack formed during shear forming by external force, to deteriorate durability.

[0083] In addition, the above Relationships 2 and 3 may correspond to cooling conditions for optimizing a microstructure such that strength and durability of steel are improved by a phase transformation process during cooling. That is, since a type and a fraction of a structure phase, as well as an aspect ratio of a crystal grain and a size of a grain boundary vary depending on the cooling conditions, cooling may be performed under the conditions satisfying the above Relationships 2 and 3.

Air cooling

[0084] A coil obtained by completing the cooling and coiling as described above may be air-cooled to a temperature within a range of room temperature to 200°C. In this case, the air-cooling of the coil may refer to cooling in at atmosphere having a cooling rate of 0.001 to 10°C/hour. In this case, when the cooling rate exceeds 10°C/hour, some of non-transformed phases of steel may be easy to be transformed into a MA phase, to deteriorate shear formability, punch formability and durability of the steel. In order to control the cooling rate to less than 0.001°C/hour, separate heating and heat retention facilities or the like are required, which may be economically disadvantageous.

[0085] As described above, the air-cooled steel material may be pickled and oiled, and then heated to a temperature within a range of 450 to 740°C to perform a hot-dip galvanizing process.

[0086] The hot-dip galvanizing process may use a zinc-based plating bath, and an alloy composition in the zinc-based plating bath is not particularly limited, but, for example, may be a plating bath containing magnesium (Mg) : 0.01 to 30wt%, aluminum (Al) : 0.01 to 50wt%, a balance Zn, and inevitable impurities.

[0087] In the description below, an example embodiment of the present disclosure will be described in greater detail. It should be noted that the example embodiments are provided to describe the present disclosure in greater detail, and to not limit the scope of rights of the present disclosure. The scope of rights of the present disclosure may be determined on the basis of the subject matters recited in the claims and the matters reasonably inferred from the subject matters.

[Mode for Invention]

(Example)

[0088] A steel slab having an alloy composition of Table 1 below was prepared. In this case, amounts of the alloy composition are % by weight, and the remainder may include Fe and inevitable impurities. Steel materials were manufactured according to manufacturing conditions of Table 2 below for the prepared steel slabs. In this case, when cooling after hot-rolling, first cooling was completed at 600°C, and second cooling was completed at a coiling temperature.

[0089] In Table 2 below, FDT denotes a temperature at an finishing hot-rolling (a hot-rolling end temperature), and CT denotes a coiling temperature, and a cooling rate in air-cooling after completion of coiling was uniformly applied at 1°C/hour.

[Table 1]

Steel	Alloy Composition (wt%)									
	C	Si	Mn	Cr	Al	P	S	N	Ti	Nb
CS1	0.06	0.3	1.8	0.2	0.03	0.01	0.004	0.004	0.005	0.02
CS2	0.06	0.3	1.8	0.2	0.03	0.008	0.004	0.004	0.005	0.02
CS3	0.07	0.04	1.7	0.6	0.03	0.01	0.005	0.004	0.05	0.005
CS4	0.06	0.5	2.0	0.007	0.03	0.01	0.004	0.005	0.04	0.03
CS5	0.06	0.5	2.0	0.007	0.03	0.005	0.004	0.005	0.04	0.03
CS6	0.07	0.5	1.6	0.008	0.03	0.01	0.003	0.004	0.08	0.03
CS7	0.07	0.5	1.6	0.008	0.03	0.01	0.003	0.004	0.08	0.03
CS8	0.07	0.4	2.2	0.012	0.03	0.007	0.004	0.004	0.1	0.02
CS9	0.07	0.4	2.2	0.012	0.03	0.01	0.004	0.004	0.1	0.02
CS10	0.08	0.4	1.6	0.8	0.05	0.01	0.003	0.006	0.04	0.045
CS11	0.08	0.4	1.6	0.8	0.05	0.01	0.003	0.006	0.04	0.045

(continued)

Steel	Alloy Composition (wt%)									
	C	Si	Mn	Cr	Al	P	S	N	Ti	Nb
CS12	0.04	0.5	1.8	0.3	0.03	0.01	0.002	0.004	0.065	0.03
CS13	0.16	0.55	1.6	0.2	0.03	0.01	0.003	0.004	0.07	0.03
CS14	0.08	1.2	2.0	0.3	0.03	0.01	0.003	0.004	0.06	0.025
CS15	0.08	0.5	0.8	0.8	0.03	0.01	0.003	0.004	0.05	0.035
CS16	0.07	0.5	2.5	0.01	0.03	0.01	0.003	0.004	0.07	0.03
CS17	0.08	0.5	1.7	1.1	0.03	0.01	0.004	0.004	0.05	0.03
IS1	0.06	0.05	1.5	0.05	0.03	0.005	0.003	0.005	0.095	0.03
IS2	0.06	0.3	1.2	0.9	0.03	0.01	0.003	0.005	0.04	0.04
IS3	0.08	0.5	1.7	0.5	0.03	0.01	0.003	0.005	0.06	0.05
IS4	0.07	0.3	1.6	0.8	0.03	0.008	0.003	0.005	0.07	0.06
IS5	0.09	0.3	1.6	0.9	0.03	0.01	0.002	0.004	0.07	0.04
IS6	0.09	0.1	1.85	0.8	0.03	0.01	0.003	0.004	0.05	0.04
IS7	0.11	0.5	1.95	0.7	0.03	0.01	0.003	0.004	0.06	0.045
IS: Inventive Steel, CS: Comparative Steel										

[0090] (Comparative steels 1 to 11 in Table 1 may have an alloy composition satisfying the scope of the present disclosure, but may be indicated as comparative steels as manufacturing conditions in Table 2 below deviate from the present disclosure.)

[Table 2]

Steel	Thick ness (mm)	FDT (°C)	CT (°C)	CRi (°C/s)	CR_2 (°C/s)	Relationship 1		Relationship 2		Relationship 3		
						Tn	Satisfied	CRi	Satisfied	CR_{Max}	CR_{Min}	Satisfied
CS1	2.9	890	455	78	30	865	×	56.2	○	38.0	6.9	○
CS2	10	885	460	65	25	855	×	56.2	○	38.0	6.9	○
CS3	7	835	470	72	22	900	×	31.0	○	29.5	3.9	○
CS4	3.2	870	443	54	38	874	○	69.3	×	42.1	9.4	○
CS5	8	858	485	51	25	868	○	69.3	×	42.1	9.4	○
CS6	3.3	906	475	88	35	863	×	102.4	×	48.6	15.1	○
CS7	9	895	430	80	42	856	×	102.4	×	48.6	15.1	○
CS8	3.8	850	450	35	30	915	×	53.9	×	39.2	7.9	○
CS9	8	845	448	33	35	909	×	53.9	×	39.2	7.9	○
CS10	9	880	450	78	75	893	○	57.1	○	32.2	7.8	×
CS11	9	875	465	85	3	893	○	57.1	○	32.2	7.8	×
CS12	8	840	450	120	35	875	○	76.8	○	41.6	11.4	×
CS13	8	848	450	95	22	866	○	62.3	○	27.3	9.4	○
CS14	8	820	455	105	12	832	○	48.4	○	12.3	10.5	×
CS15	8	816	450	125	28	828	○	108.3	○	37.1	16.7	○
CS16	8	910	445	80	20	916	○	35.8	○	35.2	4.7	○
CS17	8	904	462	75	9	902	○	21.5	○	13.9	4.2	○
IS1	8	855	442	120	42	893	○	110.9	○	62.6	14.6	○
IS2	7	850	450	125	31	876	○	81.1	○	38.5	11.4	○
IS3	9	870	443	98	18	890	○	76.2	○	39.9	10.7	○
IS4	8	880	455	85	20	924	○	78.0	○	44.8	10.2	○
IS5	9	872	466	105	28	916	○	49.0	○	30.0	7.0	○
IS6	10	890	452	89	25	935	○	32.4	○	33.0	3.0	○

(continued)

Steel	Thick ness (mm)	FDT (°C)	CT (°C)	CRi (°C/s)	CR_2 (°C/s)	Relationship 1		Relationship 2		Relationship 3		
						Tn	Satisfied	CRi	Satisfied	CR_{Max}	CR_{Min}	Satisfied
IS7	11	880	440	102	18	914	○	33.1	○	23.9	4.7	○
IS: Inventive Steel, CS: Comparative Steel												

[0091] For each of the steel sheets manufactured according to the above, mechanical properties of tensile strength (TS), yield strength (YS), and elongation (T-El), and durability were evaluated, a microstructure thereof was also observed, and results therefrom may be illustrated in Table 3 below.

[0092] Specifically, yield strength and elongation mean 0.2% off-set yield strength and elongation at break, respectively, and measurement of tensile strength was performed by taking a JIS No. 5 standard test piece in a direction perpendicular to a rolling direction.

[0093] The evaluation of durability was performed by a high-cycle fatigue test (bending fatigue test) on a test piece having a punched-formed portion, and results therefrom were shown. In this case, a piece for the fatigue test was produced by punching a hole having a diameter of 10 mm with a clearance of 12% in a central portion of a piece for the bending fatigue test having a length of 40 mm and a width of 20 mm in a gauge, and was tested under conditions of a stress ratio of -1 and a frequency of 15 Hz. Fatigue strength ($S_{Fatigue}$) was expressed as a strength ratio ($S_{Fatigue}/YS$) compared to the yield strength, from which changes in cross-sectional quality and durability of a punched portion may be confirmed.

[0094] A microstructure of each of the steel materials was observed from a central portion (t/2) in a thickness direction. A length of a grain boundary observed in a unit area (1 mm^2), corresponding to an area of the grain boundary, and an aspect ratio (AR) of a crystal grain having a large tilt angle grain boundary of 15° or more, were measured using electron back scattered diffraction (EBSD) . An area fraction on MA was etched by a Lepera etching process and was then analyzed at 1000 magnification using an optical microscope and an image analyzer. In addition, phase fractions of martensite (M), ferrite (F), bainite (B), and pearlite (P) were measured from results of analysis at 3000 and 5000 magnifications using a scanning electron microscope (SEM).

[0095] In Table 3 below, F denotes polygonal ferrite having an equiaxed crystal shape, and B denotes a sum of fractions of ferrite phases observed in a low temperature region such as a bainite phase, an acicular ferrite phase, a bainitic ferrite phase, and the like.

[0096] In addition, in Table 3 below, AR0.3 denotes a ratio (an area fraction) of crystal grains having an aspect ratio of 0.3 or less, and shows results obtained by observing at 1000 magnification.

[Table 3]

Steel	Mechanical Properties			Microstructure							Durability	
	YS (MPa)	TS (MPa)	T-El (%)	Grain Boundary Length (mm)	ARO.3 (%)	F (%)	B (%)	P (%)	M (%)	MA (%)	$S_{Fatigue}$ (MPa)	$S_{Fatigue}$ /YS
CS1	518	632	19	855	56	54	43	0	1	2	169	0.33
CS2	470	565	27	608	35	47	45	2	1	5	113	0.24
CS3	539	651	26	785	53	66	30	1	2	1	128	0.24
CS4	632	775	17	880	58	43	53	0	3	1	198	0.31
CS5	510	623	18	730	45	59	32	4	1	4	121	0.24
CS6	560	686	17	826	48	41	54	1	2	2	172	0.31
CS7	538	652	26	688	29	45	46	4	1	4	129	0.24
CS8	715	872	14	1028	62	61	30	2	5	2	204	0.29
CS9	602	738	22	840	39	67	22	6	1	4	130	0.22
CS10	625	764	22	963	31	40	48	0	8	4	142	0.23
CS11	587	708	24	765	40	56	38	4	1	1	136	0.23
CS12	499	604	26	670	45	70	28	0	1	1	117	0.23
CS13	762	915	16	966	41	29	50	8	9	4	180	0.24
CS14	**568**	**690**	**25**	**799**	**67**	**36**	**56**	**1**	**1**	**6**	**137**	**0.24**
CS15	504	611	27	882	72	59	38	0	1	2	123	0.24
CS16	715	866	19	1036	28	13	72	2	8	5	168	0.23
CS17	721	882	20	1015	35	17	76	0	6	1	172	0.24
IS1	554	675	25	750	25	47	51	0	1	1	169	0.31
IS2	601	724	22	882	30	35	62	0	1	2	187	0.31
IS3	674	816	22	977	28	30	65	0	4	1	193	0.29
IS4	690	821	21	1014	30	32	64	0	3	1	195	0.28
IS5	762	943	19	1020	35	28	67	0	3	2	213	0.28
IS6	771	924	18	1155	38	20	76	1	2	1	208	0.27

(continued)

Steel	Mechanical Properties			Microstructure							Durability	
	YS (MPa)	TS (MPa)	T-El (%)	Grain Boundary Length (mm)	ARO.3 (%)	F (%)	B (%)	P (%)	M (%)	MA (%)	$S_{Fatigue}$ (MPa)	$S_{Fatigue}$ /YS
IS7	780	955	17	1084	42	9	85	1	3	2	220	0.28
IS: Inventive Steel, CS: Comparative Steel												

[0097] As shown in Tables 1 to 3, Inventive steels 1 to 7 satisfying all of the alloy composition and the manufacturing conditions proposed in the present disclosure had a matrix structure formed as a composite structure of ferrite and bainite . In addition, as a fraction of a crystal grain having an aspect ratio of 0.3 or less in a central portion of a steel material in a thickness direction was less than 50% (see FIG. 2), and lengths of all grain boundaries were formed to be 700 mm or more, durability as well as high strength to be intended were excellently secured.

[0098] Comparative steels 1 to 11 satisfying the alloy composition proposed in the present disclosure but deviating from the manufacturing conditions the present disclosure could not secure intended physical properties.

[0099] In comparative steels 1 to 3, a hot-rolled finishing temperature did not satisfy Relationship 1 proposed in the present disclosure. In comparative steel 1 having a final steel thickness of 2.9 mm, a ferrite phase elongated from a central portion was excessively formed, but fatigue characteristics were not significantly deteriorated. This was judged due to the facts that, when hot-rolled to a thickness of 2.9 mm, an amount of reduction in a non-recrystallized temperature region was greatly increased and an elongated microstructure was developed, and cross-sectional quality of the punched portion was good as the microstructure in a thickness direction was uniform.

[0100] In comparative Steel 2 and 3, thick steel materials respectively having a thickness of 10 mm and 7 mm, comparative steel 2 has been shown to grow easily micro-cracks formed on a cross-section when exposed to a fatigue environment, to deteriorate fatigue characteristics, as an MA phase was developed and a length of a grain boundary was formed to be less than 700 mm in a microstructure of central portion. In addition, it was judged that comparative steel 3 had excessively elongated crystal grains formed in a central portion in a thickness direction due to hot-rolling in a low temperature region, to generate fatigue failure along a weak grain boundary. That is, this may be due to development of fine cracks along the elongated ferrite grain boundaries in the central portion in the thickness direction during punch forming.

[0101] In comparative steels 4 and 5 having the same components, and in which a condition of first cooling during cooling after hot-rolling did not satisfy relationship 2, comparative steel 4 had a thickness of 3.2 mm, and comparative steel 5 had a thickness of 8 mm. Among them, the comparative steel 4 having a thickness of less than 5 mm formed a lot of elongated crystal grains similar to comparative steel 1, but did not significantly deteriorate fatigue characteristics, because there was little formation of coarse carbides at a grain boundary even when a cooling rate was slow during the first cooling. Meanwhile, the comparative steel 5 having a greater thickness had a slow cooling rate during the first cooling, to form pearlite in the central portion in the thickness direction, and have somewhat excessive ferrite phase fraction, and, in addition, an MA phase was also observed in the crystal grains, indicating that fatigue characteristics were deteriorated.

[0102] Comparative steels 6 and 7 had the same components, respectively had thicknesses of 3.3 mm and 9 mm, and did not satisfied both Relationships 1 and 2. Comparative steel 6, a thinner steel material, may be judged to secure an effect of delaying recrystallization in a thickness direction even when a hot-rolling temperature is high. In this case, a cooling rate was slow during first cooling, but no pearlite or MA phase did develop in a central portion in the thickness direction, resulting in good fatigue characteristics. Meanwhile, comparative steel 7 having a greater thickness had a large microstructure and a grain boundary length of less than 700 mm due to high rolling temperature and slow cooling rate during first cooling, and also formed an MA phase and a pearlite phase, to deteriorate fatigue characteristics.

[0103] Comparative steels 8 and 9 had a lower finishing temperature during hot-rolling, as compared to the range suggested by the present disclosure, and a slower cooling rate during first cooling. These also had the same composition but different thicknesses. Comparative steel 8, a thinner steel material, had a large number of fine and elongated ferrite phases formed throughout the thickness, but did not deteriorate fatigue characteristics. Meanwhile, comparative steel 9, a thicker steel material, had an MA phase and a pearlite phase in a central portion in a thickness direction to deteriorate fatigue characteristics.

[0104] Comparative steel 10 had a cooling rate during second cooling, deviating from the present disclosure, that is, not satisfying Relationship 3, and was judged that a martensite phase was excessively formed in a central portion in a thickness direction due to an excessively fast cooling rate during second cooling, to proceed a fracture easily in a region with a large difference in hardness with a surrounding phase when exposed to fatigue environment.

[0105] Comparative steel 11 also did not satisfy Relationship 3, and had a greatly slower cooling rate during second cooling to excessively form a pearlite phase, to deteriorate fatigue characteristics.

[0106] Comparative steels 12 to 17 were manufactured to have an alloy composition deviated from the present disclosure, satisfy all Relationships 1 to 3 at the time of manufacture, and have the same thickness (8 mm), to deteriorate fatigue characteristics.

[0107] Specifically, comparative Steel 12 had an insufficient amount of C, excessively formed a ferrite phase in a central portion in a thickness direction, and did not sufficiently formed a bainite phase. Due to the above, a microstructure became coarse and a fatigue strength was low.

[0108] Comparative steel 13 had an excessively addition amount of C, and excessively formed pearlite and martensite phases during a phase transformation process due to a high concentration of C in a non-transformed phase, to have a lower fatigue strength than a yield strength.

[0109] Comparative steel 14 had a greatly high amount of Si, and formed an MA phase together with a bainite phase. In this case, a large number of elongated microstructures were observed. Due to the above, fatigue characteristics were deteriorated, which was believed to be due to formation of a large number of cracks around the relatively hard MA phase.

[0110] Comparative steel 15 had an insufficient amount of Mn. In this case, although it was manufactured by satisfying Relationships 1 to 3 to obtain a recrystallization delay effect and a uniform microstructure, a ferrite phase was excessively formed in a central portion in a thickness direction, to have low strength and fatigue strength.

[0111] Comparative steel 16 had an excessively addition amount of Mn. In this case, a martensite phase was excessively developed along an Mn segregation zone developed in a central portion in a thickness direction, to deteriorate cross-sectional quality and fatigue characteristics.

[0112] In addition, comparative steel 17 had an excessively addition amount of Cr. In this case, similarly to comparative steel 16 above, a lot of martensite phases was observed to be formed locally in a central portion in a thickness direction, to deteriorate fatigue characteristics.

[0113] Meanwhile, after observing a microstructure of inventive steel 4, results obtained by measuring an aspect ratio of a crystal grain were shown in Table 4 below.

[Table 4]

Min	Max	Total Fraction
0	0.1	0.000
0.1	0.2	0.088
0.2	0.3	0.212
0.3	0.5	0.523
0.5	1	0.176

[0114] In Table 4, Min and Max refer to minimum and maximum values of an aspect ratio (short side length of crystal grains/long side length of crystal grains), respectively, and Total Fraction refers to an area fraction of a crystal grain corresponding to a range from more than the minimum value (Min) to the maximum value (Max) or less.

[0115] As shown in Table 4, in inventive steel 4, it can be seen that a fraction of a crystal grain having an aspect ratio (a ratio of short side/long side) of 0.3 or less was less than 50% (total fraction less than 0.5).

[0116] 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 durability, comprising, by weight, carbon (C): 0.05 to 0.15%, silicon (Si): 0.01 to 1.0%, manganese (Mn): 1.0 to 2.3%, aluminum (Al) : 0.01 to 0.1%, chromium (Cr): 0.005 to 1.0%, phosphorus (P): 0.001 to 0.05%, sulfur (S): 0.001 to 0.01%, nitrogen (N): 0.001 to 0.01%, niobium (Nb): 0.005 to 0.07%, titanium (Ti): 0.005 to 0.11%, a balance of Fe, and other inevitable impurities,

   wherein a sum of a fraction of a ferrite phase and a fraction of a bainite phase in a microstructure is 90% or more, and
   a fraction of a crystal grain, in which an aspect ratio (a ratio of short side/long side) of the crystal grain in a central portion (a portion ranging a t/4 point to a t/2 point in a thickness direction) is 0.3 or less, is less than 50%, and a length of a grain boundary observed in a unit area (1 mm^2) in the central portion is 700 mm or more.

2. The high-strength steel material of claim 1, wherein a fraction of an MA phase (a martensite-austenite mixed structure) is less than 3%.

3. The high-strength steel material of claim 1, wherein a combined area fraction of an MA phase (a martensite-austenite mixed structure) and a martensite phase is 1 to 10%.

4. The high-strength steel material of claim 1, wherein a tensile strength is 650 MPa or more, and a ratio of fatigue limit and yield strength (fatigue limit/yield strength) is 0.25 or more.

5. A method of manufacturing a high-strength steel material having excellent durability, comprising:

reheating a steel slab including, by weight, carbon (C) : 0.05 to 0.15%, silicon (Si): 0.01 to 1.0%, manganese (Mn): 1.0 to 2.3%, aluminum (Al) : 0.01 to 0.1%, chromium (Cr): 0.005 to 1.0%, phosphorus (P): 0.001to 0.05%, sulfur (S): 0.001 to 0.01%, nitrogen (N): 0.001 to 0.01%, niobium (Nb): 0.005 to 0.07%, titanium (Ti): 0.005 to 0.11%, a balance of Fe, and other inevitable impurities, in a temperature within a range of 1200 to 1350°C;
hot-rolling the heated steel slab to prepare a hot-rolled steel sheet;
cooling the hot-rolled steel sheet to a temperature within a range of 400 to 500°C and then coiling (CT); and
air-cooling to a temperature within a range of room temperature to 200°C after coiling,
wherein the hot-rolling includes finish hot-rolling performed at a temperature (FDT (°C)) satisfying the following Relationship 1, and
the cooling includes first cooling and second cooling, wherein the first cooling is performed at a cooling rate (CR_1) satisfying the following Relationship 2, and the second cooling is performed at a cooling rate (CR_2) satisfying the following Relationship 3:

[Relationship 1]

$$Tn-50 \leq FDT \text{ (hot-rolling end temperature (°C))} \leq Tn$$

where Tn = 730 + 92×[C] + 70×[Mn] + 45×[Cr] + 650× [Nb] + 410× [Ti] - 80×[Si] - 1.4× (t-5) (where, an element refers to wt% of the element, and t refers to a thickness (mm) of the final hot-rolled steel sheet)

[Relationship 2]

$$CR_1 \geq 196 - 300\times[C] + 4.5\times[Si] - 71.8\times[Mn] - 59.6\times[Cr] + 187\times[Ti] + 852\times[Nb]$$

where an element refers to wt% of the element

[Relationship 3]

$$CR_{Min} \leq CR_2 \leq CR_{Max}$$

where CR_{Max}= 76.6 - 157× [C] - 25.2×[Si] - 14.1×[Mn] - 27.3× [Cr] + 61×[Ti] + 448× [Nb], CR_{Min}= 27.4 - 45.3× [C] + 5.28× [Si] - 11×[Mn] - 7.33×[Cr] + 42.3×[Ti] + 82×[Nb] (where, an element refers to wt% of the element).

6. The method of claim 5, wherein the first cooling ends at 600°C.

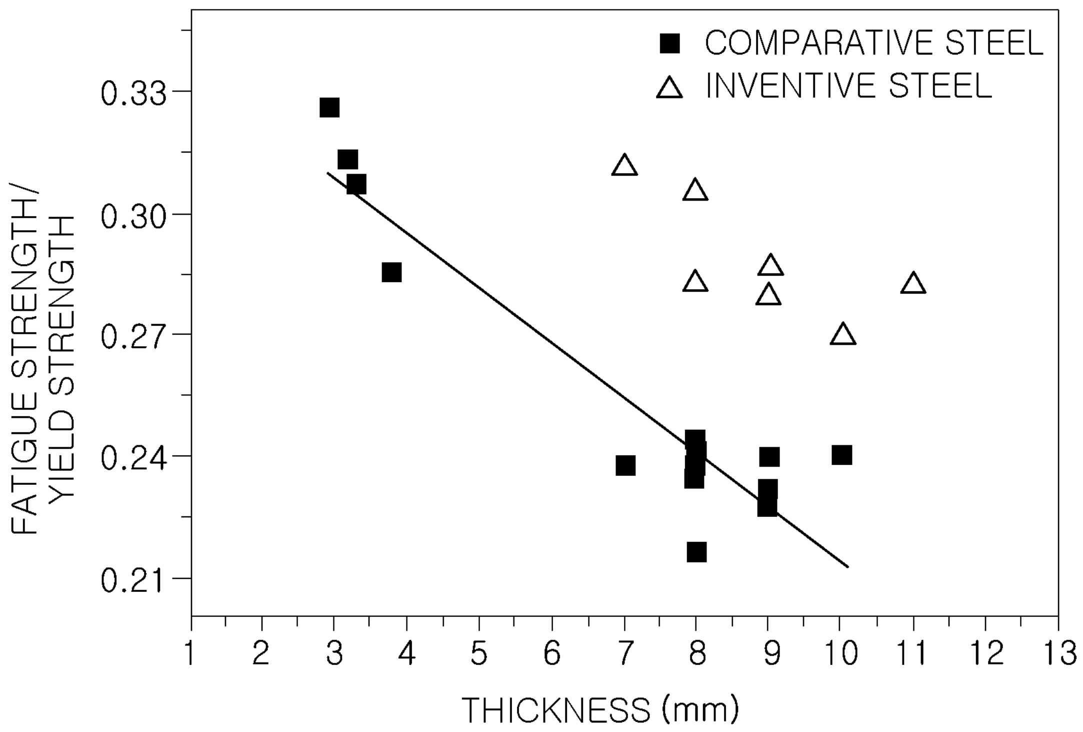
7. The method of claim 5, wherein the second cooling ends at a coiling temperature (CT (°C)).

8. The method of claim 5, further comprising pickling and oiling the cooled steel sheet, after the cooling.

9. The method of claim 8, further comprising heating the pickled and oiled steel sheet to a temperature within a range of 450 to 740°C, after the pickling and oiling, and then hot-dip galvanizing.

10. The method of claim 9, wherein the hot-dip galvanizing is performed using a plating bath containing, by weight, magnesium (Mg): 0.01 to 30%, aluminum (Al) : 0.01 to 50%, a balance Zn, and inevitable impurities.

【FIG. 1】


COMPARATIVE STEEL
INVENTIVE STEEL
FATIGUE STRENGTH/
YIELD STRENGTH
0.33
0.30
0.27
0.24
0.21
1
2
3
4
5
6
7
8
9
10
11
12
13
THICKNESS (mm)

INTERNATIONAL SEARCH REPORT

International application No.
PCT/KR2019/016299

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/38(2006.01)i, C22C 38/26(2006.01)i, C22C 38/28(2006.01)i, C22C 38/04(2006.01)i, C22C 38/02(2006.01)i, C22C 38/06(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 38/38; B21B 3/00; C21D 9/46; C22C 38/00; C22C 38/26; C22C 38/28; C22C 38/04; C22C 38/02; C22C 38/06; C21D 8/02

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: hot rolled steel, niobium(Nb), titanium(Ti), ferrite, bainite

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X	JP 2010-174343 A (JFE STEEL CORP.) 12 August 2010 See paragraphs [0027], [0030], [0031], [0033]-[0037], claims 1, 3 and table 2, table 3-1, table 3-2.	1-7
Y		8-10
Y	KR 10-2015-0074943 A (POSCO) 02 July 2015 See claims 8, 9.	8-10
A	KR 10-2004-0027981 A (NIPPON STEEL CORPORATION) 01 April 2004 See claims 14, 20.	1-10
A	JP 2000-297350 A (KAWASAKI STEEL CORP.) 24 October 2000 See claims 1, 2, 6.	1-10
A	JP 2001-226744 A (KAWASAKI STEEL CORP.) 21 August 2001 See claims 1, 2, 6.	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

"E" earlier application or patent but published on or after the international filing date

"L" document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified)

"O" document referring to an oral disclosure, use, exhibition or other means

"P" document published prior to the international filing date but later than the priority date claimed

"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention

"X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone

"Y" document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art

"&" document member of the same patent family

Date of the actual completion of the international search	Date of mailing of the international search report
04 MARCH 2020 (04.03.2020)	**04 MARCH 2020 (04.03.2020)**

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.
PCT/KR2019/016299

Patent document cited in search report	Publication date	Patent family member	Publication date
JP 2010-174343 A	12/08/2010	JP 5195469 B2	08/05/2013
KR 10-2015-0074943 A	02/07/2015	KR 10-1568519 B1	11/11/2015
KR 10-2004-0027981 A	01/04/2004	CN 1547620 A	17/11/2004
		EP 1431407 A1	23/06/2004
		EP 1431407 B1	29/10/2014
		EP 2415893 A2	08/02/2012
		EP 2415893 A3	17/10/2012
		EP 2415893 B1	05/11/2014
		EP 2415894 A2	08/02/2012
		EP 2415894 A3	17/10/2012
		EP 2415894 B1	19/12/2018
		JP 2003-064443 A	05/03/2003
		JP 2003-064444 A	05/03/2003
		JP 2004-052103 A	19/02/2004
		JP 4041295 B2	30/01/2008
		JP 4041296 B2	30/01/2008
		JP 4171296 B2	22/10/2008
		KR 10-0548864 B1	02/02/2006
		TW I290177 B	21/11/2007
		US 2004-0238081 A1	02/12/2004
		US 2008-0166257 A1	10/07/2008
		US 2008-0295924 A1	04/12/2008
		US 2008-0308200 A1	18/12/2008
		US 7534312 B2	19/05/2009
		US 7749343 B2	06/07/2010
		US 7776161 B2	17/08/2010
		US 8052807 B2	08/11/2011
		WO 03-018857 A1	06/03/2003
JP 2000-297350 A	24/10/2000	AT 272723 T	15/08/2004
		BR 0000325 A	23/01/2001
		BR 0000325 B1	05/05/2009
		CA 2297291 A1	09/08/2000
		CA 2297291 C	05/08/2008
		CN 1263168 A	16/08/2000
		DE 60012595 T2	16/12/2004
		EP 1028167 A2	16/08/2000
		EP 1028167 A3	17/04/2002
		EP 1028167 B1	04/08/2004
		ES 2224922 T3	16/03/2005
		JP 3858551 B2	13/12/2006
		KR 10-0511727 B1	31/08/2005
		KR 10-2000-0057842 A	25/09/2000
		TW 466276 B	01/12/2001
		US 6425963 B1	30/07/2002
JP 2001-226744 A	21/08/2001	JP 4306078 B2	29/07/2009

Form PCT/ISA/210 (patent family annex) (January 2015)

REFERENCES CITED IN THE DESCRIPTION

This list of references cited by the applicant is for the reader's convenience only. It does not form part of the European patent document. Even though great care has been taken in compiling the references, errors or omissions cannot be excluded and the EPO disclaims all liability in this regard.

Patent documents cited in the description

- JP 2002322541 A **[0008]**
- JP 101528084 B **[0008]**
- JP 9143570 A **[0008]**