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

(11) **EP 4 386 103 A1**

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

(43) Date of publication:
19.06.2024 Bulletin 2024/25

(21) Application number: **22856080.1**

(22) Date of filing: **01.08.2022**

(51) International Patent Classification (IPC):
C22C 38/38 *(2006.01)* ***C22C 38/22*** *(2006.01)*
C22C 38/28 *(2006.01)* ***C22C 38/00*** *(2006.01)*
C22C 38/60 *(2006.01)* ***C22C 38/06*** *(2006.01)*
C22C 38/26 *(2006.01)* ***C22C 38/32*** *(2006.01)*
C21D 8/02 *(2006.01)*

(52) Cooperative Patent Classification (CPC):
B21C 47/02; C21D 8/02; C22C 38/00; C22C 38/06; C22C 38/22; C22C 38/26; C22C 38/28; C22C 38/32; C22C 38/38; C22C 38/60; C23G 1/02

(86) International application number:
PCT/KR2022/011282

(87) International publication number:
WO 2023/018081 (16.02.2023 Gazette 2023/07)

(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: **09.08.2021 KR 20210104554**

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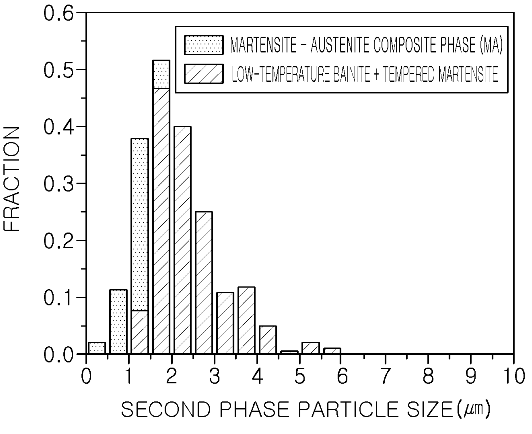
(54) HIGH STRENGTH HOT-ROLLED STEEL SHEET HAVING EXCELLENT FORMABILITY, AND MANUFACTURING METHOD THEREFOR

(57) The present invention relates to a hot-rolled steel sheet, which can be suitably applied to an automobile chassis structure member and the like, and, more specifically, to a high-strength hot-rolled steel sheet having excellent formability, and a manufacturing method therefor.

FIG.1



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Processed by Luminess, 75001 PARIS (FR)

Description

Technical Field

[0001] The present disclosure relates to a hot-rolled steel sheet, which can be suitably applied to an automobile chassis structural member and the like, and, more specifically, to a high-strength hot-rolled steel sheet having excellent formability, and a manufacturing method therefor.

Background Art

[0002] Recently, in order to reduce global warming, there has been a rapid transition from internal combustion engine vehicles, which are a mainstream in the automobile market, to eco-friendly vehicles such as electric vehicles, or the like.

[0003] As internal combustion engine vehicles are converted to electric vehicles, the types of parts consisting the vehicle are also changing, and a weight of the vehicle is also changing. For example, when comparing a weight of the internal combustion engine vehicle and an electric vehicle released in the same model, it is known that the weight of the electric vehicle increases by approximately the weight of the battery, as compared to the internal combustion engine vehicle.

[0004] Meanwhile, chassis parts of a vehicle serve to support a vehicle body and are an important part in ensuring ride comfort and driving stability by absorbing vibrations and shocks from a road surface while driving. As the weight of a vehicle increases, a fatigue load applied to the chassis parts increases, so steel materials applied to chassis parts such as electric vehicles, or the like, are required to have excellent fatigue strength.

[0005] Since the fatigue strength of steel material is proportional to a tensile strength and yield strength, the steel material used for chassis parts such as electric vehicles, or the like is required to improve the tensile strength and yield strength.

[0006] In addition, since chassis parts are manufactured by press molding, it is required to secure formability such as elongation and hole expandability, suitable for press molding, in addition to improving tensile strength and yield strength to improve fatigue strength.

[0007] Until now, various technologies have been proposed to improve the strength and formability of hot-rolled steel sheet.

[0008] As an example, Patent Document 1 discloses a method of forming a microstructure of steel comprising 90% or more of bainitic ferrite, and a fraction of martensite and bainite is controlled to 5% or less, respectively, to improve hole expandability. According to Patent Document 1, the tensile strength of a hot- rolled steel sheet may be secured at 980 MPa or more and the hole expandability can be secured at 70% or more, but it does not disclose improvement in elongation required for press forming.

[0009] Therefore, in order to ensure the driving stability of chassis parts for eco-friendly vehicles such as electric vehicles, or the like, there is a need to develop a steel material, not only having high tensile strength and yield strength and excellent fatigue life, but also having excellent formability, such as elongation and hole expandability, to facilitate press forming.

[Prior art Document]

[Patent Document]

[0010] (Patent Document 1) Japanese Patent Publication No. 2008-255484

Summary of Invention

Technical Problem

[0011] An aspect of the present disclosure is to provide a hot-rolled steel sheet with high strength and excellent fatigue performance as well as excellent formability and a method for manufacturing the same.

[0012] The object of the present disclosure is not limited to the above. A person skilled in the art will have no difficulty in understanding the further subject matter of the present disclosure from the general content of this specification.

Solution to Problem

[0013] According to an aspect of the present disclosure, provided is a high-strength hot-rolled steel sheet having excellent formability, the hot-rolled steel sheet including, by weight: carbon (C): 0.05 to 0.17%, silicon (Si): 0.01 to 1.5%, manganese (Mn): 1.5 to 3.0%, aluminum (Al): 0.01 to 0.1%, chromium (Cr): 2.0% or less (including 0%), molybdenum (Mo): 2.0% or less (including 0%), titanium (Ti): 0.01 to 0.15%, phosphorus (P): 0.001 to 0.05%, sulfur (S): 0.0001 to 0.05%, nitrogen (N): 0.0001 to 0.02%, with a balance of Fe and other unavoidable impurities, wherein the following Relational Expression 1 is satisfied,

wherein a microstructure comprises 70 to 90% of acicular ferrite or bainitic ferrite by area faction, as a matrix structure, and one or more second phases among low-temperature bainite, tempered martensite, and MA phase.

[Relational Expression 1]

$$(1.5\times[Si] + 1.2\times[Cr] + 0.7\times[Mo] + 8.0\times[Ti])/[C] > 20$$

[0014] In Relational Expression 1, each element refers to a weight content.

[0015] According to another aspect of the present disclosure, provided is a manufacturing method for a high-strength hot-rolled steel sheet having excellent formability, the manufacturing method including operations of: reheating a steel slab satisfying the above-described alloy composition and Relational Expression 1, to a temperature within a range of 1100 to 1350°C; hot rolling the reheated steel slab to manufacture a hot-rolled steel sheet; primarily cooling the hot-rolled steel sheet to a temperature of Bs or lower at a cooling rate of 70°C/s or more; secondarily cooling the steel sheet to a temperature of (Bs+Ms)/2 or higher at a cooling rate of 20 °C/s or less after the primary cooling; tertiarily cooling the steel sheet to a temperature range of Ms-20°C to 500°C at a cooling rate of 30°C/s or more after the secondary cooling; and winding the steel sheet in the tertiarily-cooled temperature range,

wherein finish hot rolling is performed to satisfy the following Relational Expression 2 within a temperature range of 750 to 1150°C during the hot rolling, and a total reduction amount of two final passes is 10 to 40%.

[Relational Expression 2]

$$800 \le Du \le 1106$$

[0016] In Relational Expression 2, Du is an indicator indicating an effective grain size of austenite immediately before primary cooling after hot rolling, expressed as $Du = FDT + (7.35\times[C]) - (24.7\times[Si]) - (4.7\times[Mn]) - (3.9\times[Cr]) - (5.2\times[Mo]) - (560\times[Ti]) - (1110\times[Nb])$, where FDT refers to a rolling end temperature (°C), and each element refers to a weight content.

Advantageous Effects of Invention

[0017] As set forth above, according to the present disclosure, a steel sheet having high strength and excellent formability may be provided. The steel material of the present disclosure is suitable for automobile chassis structural members, or the like.

Brief description of drawings

[0018]

FIG. 1 is a graph illustrating classification of a phase type of a second phase according to a size of the second phase of a hot-rolled steel sheet according to an embodiment of the present disclosure.

FIG. 2 is a photograph in which microstructures of Inventive Example (a) and Comparative Example (b) are observed according to an embodiment of the present disclosure.

Best Mode for Invention

[0019] A high-strength hot-rolled steel sheet applied to conventional chassis parts, or the like, has been widely used as precipitation-strengthened steel, having excellent yield strength, elongation, and hole expandability, at the same time, by adding a large amount of carbonitrides forming elements such as Ti, Nb, and V, and winding the same in a high-temperature region around 600°C and inducing precipitation of fine carbonitrides within a ferrite matrix structure. However, when grain boundary carbides are formed at ferrite grain boundaries during high-temperature winding, not only resistance to crack propagation poor, but in a process in which generation of local micro defects is inevitable, such as shear forming during forming into parts, there is a problem in that micro-cracks easily propagate inside steel, resulting in forming cracks.

[0020] Since a shear stress which causes crack propagation increases in proportion to strength of a material, a risk of cracks that may occur during shear forming is expected to increase rapidly in a material having a tensile strength of 980 MPa.

[0021] Accordingly, the inventors of the present disclosure conducted in-depth research to develop a composite hot-rolled steel sheet having high strength even during the winding process at a low- temperature by utilizing a low-temperature transformation structure.

[0022] In general, a low-temperature transformation structure is a term referring to a microstructure created by displacive phase transformation, and as a representative phase, the low-temperature transformation structure includes bainite and martensite.

[0023] Bainite can be defined as a composite structure composed of bainitic ferrite formed through displacive phase transformation without diffusion, and a secondary product created by subsequent diffusion of an interstitial alloying element such as carbon. In order to accommodate an amount of shear strain generated during the displacive phase transformation of bainitic ferrite, a large amount of dislocations are generated within the structure, and density of the dislocations generated inside the bainitic ferrite gradually decrease by a recovery phenomenon while maintained at a temperature range above room temperature in which bainitic transformation occurs. In this case, since a speed of the recovery phenomenon is greatly affected by a temperature, the dislocation density existing inside bainitic ferrite varies depending on a formation and maintenance temperature of bainitic ferrite. Therefore, when a phase fraction and internal dislocation density are adjusted by including bainitic ferrite as a matrix structure, and adjusting the generation temperature and transformation time of bainite, elongation and yield strength of a hot-rolled steel material can be controlled.

[0024] Meanwhile, after bainitic ferrite is formed, carbon diffuses from bainitic ferrite with low solid solution to untransformed austenite, and the untransformed austenite undergoes additional bainitic transformation or is converted into a secondary product. The secondary product may be present in a form of carbides, pearlite, martensite-austenite composite phase (MA phase), etc., depending on the temperature of bainite formation and the type of alloying element. Depending on the type of these secondary products, the strength and formability of the steel may vary. In the case of finely dispersed iron carbide, it is known that it increases the strength of steel without deteriorating hole expandability, but pearlite deteriorates both the strength and hole expandability of steel. In addition, while the MA phase has an excellent effect of improving the strength of steel, if it is excessively present in the steel, it deteriorates hole expandability.

[0025] Accordingly, as the present inventors found that it is important to control the formation temperature and transformation time of bainite to maintain the type and fraction of microstructure at an appropriate level, the present inventors have confirmed that a steel sheet having not only high strength but also excellent formability may be provided by optimizing an alloy composition range of the steel sheet and process conditions such as hot rolling and cooling, to control a matrix structure of a microstructure and a type and fraction of a secondary phase, thereby completing the present disclosure.

[0026] Hereinafter, the present disclosure will be described in detail.

[0027] According to an aspect of the present disclosure, a high-strength hot-rolled steel sheet having excellent formability may include by weight: carbon (C): 0.05 to 0.17%, silicon (Si): 0.01 to 1.5%, manganese (Mn): 1.5 to 3.0%, aluminum (Al): 0.01 to 0.1%, chromium (Cr): 2.0% or less (including 0%), molybdenum (Mo): 2.0% or less (including 0%), titanium (Ti): 0.01 to 0.15%, phosphorus (P): 0.001 to 0.05%, sulfur (S): 0.0001 to 0.05%, and nitrogen (N): 0.0001 to 0.02%.

[0028] Hereinafter, the reason for limiting an alloy composition of the hot-rolled steel sheet provided by the present disclosure as described above will be explained in detail. Meanwhile, in the present disclosure, unless specifically stated, a content of each element is based on weight, and a ratio of structure is based on area.

Carbon (C): 0.05 to 0.17%

[0029] Carbon (C) is the most economical and effective element in strengthening steel, and as a C content increases, formation of ferrite is suppressed during cooling.

[0030] In the present disclosure, C diffuses into austenite during bainite transformation and stabilize austenite, thereby being transformed into a second phase, which is low-temperature bainite, tempered martensite, and martensite-austenite composite phase (MA phase) during a subsequent cooling process, so it is effective in improving tensile strength and yield strength of steel.

[0031] If the C content is less than 0.05%, a fraction of the above-described second phase decreases, making it difficult to secure a high strength. On the other hand, if the C content exceeds 0.17%, formation of pearlite is promoted, making it impossible to secure strength, and there is a problem in that formability and weldability are poor.

[0032] Therefore, in the present disclosure, C may be included in an amount of 0.05 to 0.17%, and more advantageously, C may be included in an amount of 0.06% or more and 0.15% or less.

Silicon (Si): 0.01 to 1.5%

[0033] Silicon (Si) is an element improving hardenability of steel, and serves to improve strength through a solid solution strengthening effect. In addition, by delaying formation of carbides and preventing pearlite formation, a second phase is formed into low-temperature bainite, tempered martensite, and a MA phase, the strength is improved.

[0034] If the Si content is less than 0.01%, carbides are formed and a fraction of the MA phase is relatively low, making it difficult to secure tensile strength. On the other hand, if the Si content exceeds 1.5%, a Fe-Si composite oxide is formed on a surface of a slab when the slab is reheated, which not only deteriorates surface quality of the steel sheet, but also reduces weldability.

[0035] Therefore, in the present disclosure, Si may be included in an amount of 0.01 to 1.5%, more advantageously 0.1% or more, and even more advantageously 0.3% or more. In addition, it would be effective to include Si in an amount of 1.3% or less.

Manganese (Mn): 1.5 to 3.0%

[0036] Manganese (Mn) is an element improving hardenability of steel, and prevents formation of ferrite during cooling after finish rolling, thereby facilitating formation of a low-temperature transformation structure.

[0037] If a Mn content is less than 1.5%, hardenability is insufficient so that there is a problem in that a fraction of ferrite increases excessively. On the other hand, if the Mn content exceeds 3.0%, the hardenability increases significantly and bainite transformation does not occur smoothly in a cooling zone, so a holing time to sufficiently form acicular ferrite or bainitic ferrite, to be obtained as a matrix structure in the present disclosure, is excessively increased, and an elongation rate is reduced.

[0038] Therefore, in the present disclosure, Mn may be included in an amount of 1.5 to 3.0%, and more advantageously, may be included in an amount of 1.8% or more and 2.4% or less.

Aluminum (Al): 0.01 to 0.1%

[0039] Aluminum (Al) is an element added to deoxidize molten steel, and a portion thereof is present in steel after deoxidation. When an Al content exceeds 0.1%, oxide and nitride-based inclusions increase in the steel, deteriorating formability of the steel sheet. On the other hand, when the Al content is excessively reduced to less than 0.01%, it may be economically disadvantageous as it causes an unnecessary increase in refining costs.

[0040] Therefore, in the present disclosure, Al may be included in 0.01 to 0.1%.

Chromium (Cr): 2.0% or less (including 0%)

[0041] Chromium (Cr) is an element of improving hardenability of steel, and suppresses formation of ferrite during cooling after finish rolling. In addition, chromium has excellent affinity with carbon, so Cr slows down a diffusion rate of carbon and prevents over-concentration of carbon into untransformed austenite after winding, thereby suppressing formation of pearlite and inducing a second phase to become a low-temperature transformation phase, and contributes to improve yield strength and tensile strength.

[0042] If a Cr content exceeds 2.0%, hardenability increases significantly, and bainite transformation does not occur smoothly in a cooling zone, so that a holding time to secure a fraction of acicular ferrite or bainitic ferrite, which is a matrix structure, increases excessively, thereby deteriorating elongation.

[0043] Therefore, in the present disclosure, Cr may be included in an amount of 2.0% or less, and more advantageously, may be included in an amount of 1.5% or less.

[0044] Meanwhile, in the present disclosure, even if Cr is not contained, there is no significant difficulty in securing intended physical properties. However, when adding Cr, it should be noted that it is effective to be added in amount of at least 0.01%.

Molybdenum (Mo): 2.0% or less (including 0%)

[0045] Molybdenum (Mo) is an element improving hardenability of steel, and serves to improve strength through a solid solution strengthening effect, and suppresses formation of ferrite during cooling after finish rolling. In addition, Mo suppresses formation of pearlite by slowing down a diffusion rate of carbon and preventing carbon overconcentration into untransformed austenite after winding, and improves yield strength and tensile strength by allowing a second phase to become a low-temperature transformation phase.

[0046] If a Mo content exceeds 2.0%, hardenability increases significantly, and bainite transformation does not occur smoothly in a cooling zone. As a result, a holding time for securing a fraction of acicular ferrite or bainitic ferrite, which is a matrix structure, is excessively increased, resulting in a decrease in elongation.

[0047] Therefore, in the present disclosure, Mo may be included in an amount of 2.0% or less, more advantageously in an amount of 1.0% or less, and even more advantageously in an amount of 0.5% or less.

[0048] Meanwhile, in the present disclosure, even if Mo is not contained, there is no significant difficulty in securing intended physical properties. However, when adding Mo, it should be noted that it is effective to be added in amount of at least 0.01%.

Titanium (Ti): 0.01 to 0.15%

[0049] Titanium (Ti) is an element of forming carbonitrides in steel, and is widely used to secure strength of steel by inducing formation of precipitates as described above. However, in the present disclosure, Ti is added to obtain an effect of preventing formation of pearlite by slowing a diffusion rate of carbon.

[0050] In order to sufficiently obtain the desired effect in the present disclosure, it is preferable to contain Ti in an amount of 0.01% or more. However, if a Ti content exceeds 0.15%, a fraction of a MA phase constituting a second phase becomes excessive, resulting in poor hole expandability.

[0051] Therefore, in the present disclosure, Ti may be included in an amount of 0.01 to 0.15%, and more advantageously, Ti may be included in an amount of 0.05% or more and 0.10% or less.

Phosphorus (P): 0.001 to 0.05%

[0052] Phosphorus (P) is an impurity inevitably contained in steel, and is an element which is a main cause of impairing workability of steel due to segregation. Therefore, it is desirable to control a P content as low as possible.

[0053] In theory, it is advantageous to limit the P content to 0%, but since excessive manufacturing costs are required to control the P content to less than 0.001%, a lower limit thereof can be set to 0.001%. However, if the P content exceeds 0.05%, there may be a risk that workability may deteriorate, so an upper limit of the P content may be limited to 0.05%.

Sulfur (S): 0.0001 to 0.05%

[0054] Sulfur (S) is an impurity inevitably contained in steel, and forms non-metallic inclusions by combining with Mn, or the like, which causes a problem in that workability of steel is deteriorated. Therefore, it is desirable to control a S content to as low as possible.

[0055] In theory, it is advantageous to limit a S content to 0%, but since excessive manufacturing costs are required to control the S content to less than 0.0001%, a lower limit thereof can be set to 0.0001%. However, if the S content exceeds 0.05%, there may be a risk that workability may deteriorate, so an upper limit of the S content may be limited to 0.05%.

Nitrogen (N): 0.0001 to 0.02%

[0056] Nitrogen (N) is an impurity inevitably contained in steel, and has a problem of inhibiting workability of steel by combining with Al to form nitrides. Therefore, it is desirable to control a N content as low as possible.

[0057] In theory, it is advantageous to limit the N content to 0%, but since excessive manufacturing costs are required to control the N content to less than 0.0001%, a lower limit thereof can be set to 0.0001%. However, if the N content exceeds 0.02%, there may be a risk that workability may deteriorate, so an upper limit of the N content may be limited to 0.02%.

[0058] In addition to the alloy composition described above, the hot-rolled steel sheet of the present disclosure may further include one or more of niobium (Nb) and boron (B).

Niobium (Nb): 0.01 to 0.1%

[0059] Niobium (Nb), similar to Ti, has an effect of preventing formation of pearlite by slowing a diffusion rate of carbon. However, since recrystallization is delayed during hot rolling compared to Ti, an effect of refining austenite grains is significant, and if a Ni content exceeds 0.1%, a fraction of a MA phase, which is a second phase, is excessively formed, resulting in inferior hole expandability.

[0060] Meanwhile, in order to obtain the effect when adding Nb, it is advantageous that Nb is added in an amount of 0.01% or more.

Boron (B): 0.0005 to 0.005%

[0061] Boron (B) is an element of significantly improving hardenability of steel by delaying formation of nucleation of ferrite by being segregated at austenite grain boundaries. The addition of B has an excellent effect of the formation of ferrite during cooling after hot rolling.

[0062] The present inventors found that, in addition to the already well-known effect of adding B, a transformation rate of bainite was also delayed upon addition of B. That is, the addition of B affects a fraction of acicular ferrite or bainitic ferrite generated during cooling after hot rolling (preferably during secondary cooling), so in the present disclosure, by adding B, secondary cooling conditions may be easily adjusted.

[0063] When adding B, if a B content is less than 0.0005%, an effect of delaying phase transformation of not only ferrite but also bainite cannot be sufficiently obtained. On the other hand, if the B content exceeds 0.005%, the above-mentioned effect is saturated, so it is advantageous that B is included in an amount of 0.005% or less.

[0064] The steel of the present disclosure may include remaining iron (Fe) and unavoidable impurities in addition to the above-described composition. Since unavoidable impurities may be unintentionally incorporated in a common manufacturing process, the component may not be excluded. Since these impurities are known to any person skilled in the common manufacturing process, the entire contents thereof are not particularly mentioned in the present specification.

[0065] In the hot-rolled steel sheet of the present disclosure having the above-described alloy composition, it is preferable that a content relationship of specific elements satisfies the following Relational Expression 1.

[Relational Expression 1]

$$(1.5\times[Si] + 1.2\times[Cr] + 0.7\times[Mo] + 8.0\times[Ti])/[C] > 20$$

[0066] In Relational Expression 1, each element means a weight content.

[0067] In order to secure strength targeted in the present disclosure, a low-temperature transformation phase should be formed as intended, which should suppress formation of pearlite after winding the hot-rolled steel sheet. Since driving force for the formation of pearlite increases as the content of carbon concentrated in untransformed austenite increases, it is necessary to prevent overconcentration of carbon by adding an element that slows a diffusion rate of carbon.

[0068] In the present disclosure, Cr or Mo is added along with Ti as an element that slows the diffusion rate of carbon, and it was confirmed that the formation of pearlite was delayed by preventing overconcentration of untransformed austenite therefrom.

[0069] Meanwhile, Si has a low solubility in iron carbides constituting pearlite, so Si serves to prevent formation of carbides, and as a result, prevents formation of pearlite, similarly to the elements described above.

[0070] In particular, in the present disclosure, a ratio of a sum of Si, Cr, Mo, and Ti, which prevents the formation of pearlite, and a ratio of C, which promotes the formation of pearlite, is controlled as shown in Relational Expression 1 above, by preventing the formation of pearlite and securing a second phase as a low-temperature transformation phase, yield strength and tensile strength can be improved.

[0071] The hot-rolled steel sheet of the present disclosure satisfying the above-described alloy composition and component relational expression (Relational expression 1) may have a microstructure containing acicular ferrite or bainitic ferrite as a matrix structure, and it is preferable that the matrix structure include, by area fraction, 70 to 90% of acicular ferrite or bainitic ferrite.

[0072] Steel of the present disclosure is cooled to a temperature below Bs (bainite transformation start temperature) to avoid ferrite phase transformation during primary cooling after hot rolling, and then bainite transformation proceeds by slow cooling during subsequent secondary cooling. In this case, bainitic transformation occurs in a high-temperature bainitic transformation zone, so bainitic ferrite is formed and carbon diffuses into untransformed austenite, and carbides are not formed inside bainitic ferrite. Meanwhile, although a large amount of dislocations exist inside the bainitic ferrite formed by displacive phase transformation, the dislocation density is reduced to an appropriate level by a recovery phenomenon after secondary cooling and subsequent winding, which has the effect of improving the elongation of the steel sheet.

[0073] Since the bainitic ferrite generated below the Bs temperature is similar in shape and properties to the acicular ferrite generated when supercooled in ultra-low carbon steel, in the present disclosure, it should be noted that the bainitic ferrite is managed by a total fraction of the bainitic ferrite and acicular ferrite.

[0074] If the total fraction of the matrix structure, acicular ferrite or bainitic ferrite, is less than 70%, there is a problem in that it is difficult to secure elongation. On the other hand, if the total fraction of the matrix structure exceeds 90%, there is a problem in that it is difficult to secure a low-temperature transformation structure, serving to improve the strength.

[0075] The hot-rolled steel sheet of the present disclosure a second phase, in addition to the matrix structure described above, wherein the second phase may include a low-temperature transformation structure, preferably at least one of low-temperature bainite, tempered martensite, and MA phase, and has an area fraction of 10 to 30%.

[0076] In the present disclosure, carbon diffusion to untransformed austenite proceeds with the formation of bainitic ferrite during secondary cooling during cooling after hot rolling, and the untransformed austenite is transformed into low-temperature bainite, tempered martensite, and MA phase, which is a second phase during an additional cooling process after the secondary cooling (e.g., cooling process after winding).

[0077] As described above, when bainitic ferrite is formed during secondary cooling, a size of the untransformed austenite distributed within the structure varies depending on the location, and the type of the second phase also changes. The relatively large-sized untransformed austenite has a low carbon content and can be transformed into low-temperature bainite during subsequent cooling to a winding temperature, and austenite having the smaller size is transformed into martensite at a lower temperature. Since the martensite is transformed in a relatively high temperature zone, a tempering phenomenon occurs after the martensite transformation, and a final structure becomes tempered martensite.

[0078] In the present disclosure, it should be noted that the low-temperature bainite and tempered martensite commonly contain iron carbides at grain boundaries and within a lath structure, and are therefore managed by the total fraction.

[0079] Meanwhile, since the small-sized austenite has the highest carbon concentration during secondary cooling, transformation of the austenite into low-temperature bainite or martensite immediately after winding does not proceed, and when the austenite is transformed into martensite during a final cooling step, or is not transformed into martensite, it may remain as austenite. In this case, martensite with a high carbon content has a characteristic of having a plate-type martensite rather than a lath shape, and an internal twin structure is not clearly observed during Nital etching, so it can be clearly distinguished from low-temperature bainite and tempered martensite.

[0080] This MA phase is effective in improving yield strength and tensile strength, but a difference in hardness between bainitic ferrite (or acicular ferrite), which is a matrix structure and a phase, is high, making hole expandability inferior.

[0081] Therefore, in the present disclosure, it is preferable to include 10% or more of the second phase, by area fraction in terms of securing yield strength and tensile strength, and it is preferable to limit the second phase to 30% or less to secure elongation at the same time.

[0082] Furthermore, in the present disclosure, in order to improve hole expandability of the steel sheet, it is preferable that a ratio of the MA phase in the second phase is controlled, and the MA phase is included in the ratio of less than 30% of the total area fraction of the second phase.

[0083] The hot-rolled steel sheet of the present disclosure may include one or more of ferrite and carbide as other structures in addition to the matrix structure and second phase described above, but these are preferably controlled to be less than 5%, by area fraction. Here, ferrite means granular ferrite.

[0084] Ferrite formed during cooling after hot rolling is typically formed through diffusion transformation and therefore has low strength. In the present disclosure, when the ferrite is formed in an amount of less than 5%, the previous formed ferrite undergoes shear strain to accommodate particle strain generated when the remaining austenite is transformed into bainite and martensite after ferrite is formed, so that it was confirmed that the dislocation density inside the ferrite is maintained at a high level, so strength of steel does not decrease significantly. However, if the fraction is 5% or more, it is not desirable because the strength of the steel decreases.

[0085] Meanwhile, during bainite transformation, iron carbides may be generated, along with carbon diffusion into austenite. In the present disclosure, since it is intended to improve the strength by utilizing a low-temperature transformation structure as a second phase, the formation of iron carbides may cause a decrease in the fraction of the second phase. In other words, excessive generation of iron carbides inhibits a strengthening effect targeted by the present disclosure. However, when Ti and Nb are present in steel, alloy carbonitrides may be formed, and in this case, an additional strengthening effect may be expected by grain refinement. However, coarse carbides inhibit toughness of the steel, so in the present disclosure, it is preferable that the carbides present in the hot-rolled steel sheet of the present disclosure is less than 5%.

[0086] The hot-rolled steel sheet of the present disclosure having the above-described alloy composition and microstructure has high strength with a yield strength of 750 MPa or more and a tensile strength of 980 MPa or more, and have excellent formability with an elongation of 9% or more and a hole expansion rate of 30% or more.

[0087] Hereinafter, a method for manufacturing a high-strength hot-rolled steel sheet with excellent formability according to another aspect of the present disclosure will be described in detail.

[0088] The hot-rolled steel sheet according to the present disclosure may be manufactured by performing a series of process of [reheating, hot rolling, cooling, and winding] on a steel slab satisfying the alloy composition, and Relational Expression 1 proposed in the present disclosure.

[0089] Hereinafter, each of the process conditions described above will be described in detail.

[Reheating a steel slab]

[0090] In the present disclosure, it is preferable to perform a process of reheating and homogenizing a steel slab prior to performing a rolling process, and in this case, the reheating may be performed in a temperature range of 1100 to 1350°C.

[0091] If the temperature during reheating of the steel slab is less than 1100°C, there may be a problem in that homogenization of alloy elements is not sufficient. On the other hand, if the temperature exceeds 1350°C, there may be a risk that the surface quality of the steel sheet may deteriorate due to excessive formation of oxides on a surface of the slab.

[Hot rolling]

[0092] The reheated steel slab may be hot rolled to manufacture a hot-rolled steel sheet. In this case, the hot rolling is performed in a temperature range of 750 to 1150°C, and it is preferable that a total reduction amount of two final passes is controlled to 10 to 40%.

[0093] First, if the hot rolling is started at a temperature of exceeding 1150°C, excessive oxides are formed on a surface of the steel sheet after rolling and cannot be effectively controlled even if a pickling process is performed, resulting in poor surface quality. On the other hand, if hot rolling is performed at a temperature of lower than 750°C, a rolling load increases excessively, which reduces workability, and ferrite is generated during rolling, resulting in poor anisotropy.

[0094] Typically, hot rolling is performed in multiple stages to reduce the rolling load and precisely control a thickness thereof. When performing hot rolling by such multistage rolling, if a total reduction ratio of two final passes (two subsequent passes) exceeds 40%, there may be a problem that the rolling load of the two final passes becomes excessive and workability is deteriorated. On the other hand, if the total reduction ratio of the two final passes is less than 10%, the temperature of the steel sheet reduces rapidly, causing shape defects.

[0095] Meanwhile, a grain size of austenite after hot rolling is affected by an alloy composition, rolling end temperature, and reduction amount, which affects ferrite and bainite formation behavior and final microstructure in a subsequent cooling process. In addition, in the present disclosure, a fraction of a MA phase in a second phase, which is a main constituent phase, is greatly affected by the austenite grains after hot rolling.

[0096] As illustrated in FIG. 1, it can be seen that the smaller the size (particle size) of the second phase, the more the phase type of the second phase tends to be the MA phase. The size (grain size) of this second phase is affected by nucleation generation behavior in bainite transformation, but due to a nature of displacive phase transformation, the size of the second phase cannot be larger than the size of austenite before transformation, so in order to control the size of the second phase, it is advantageous to control the grain size of austenite after hot rolling.

[0097] Accordingly, in the present disclosure, an effective grain size of austenite after hot rolling is derived as a relationship between a rolling end temperature FDT and a specific alloy composition, and is specifically defined by the following relational expression 2. If a value of Du according to the following relational expression 2 is 800 or more, a MA phase may be appropriately formed and a hole expansion rate can be secured at 30% or more. On the other hand, if the value exceeds 1106, an austenite grain size becomes excessively coarse, to delay bainite transformation, causing a problem that elongation is inferior.

[Relational Expression 2]

$$800 \leq \mathrm{Du} \leq 1106$$

[0098] In Relational Expression 2, Du is an indicator of an effective grain size of austenite immediately before primary cooling after hot rolling, expressed as $\mathrm{Du} = \mathrm{FDT} + (7.35\times[\mathrm{C}]) - (24.7\times[\mathrm{Si}]) - (4.7\times[\mathrm{Mn}]) - (3.9\times[\mathrm{Cr}]) - (5.2\times[\mathrm{Mo}]) - (560\times[\mathrm{Ti}]) - (1110\times[\mathrm{Nb}])$, where FDT is a rolling end temperature (°C), and each element refers to a weight content.

[Cooling and winding]

[0099] A hot-rolled steel sheet manufactured as described above is cooled, and it is preferable to perform the processes in stages according to a cooling temperature.

[0100] Specifically, it is preferable that the hot-rolled steel sheet is primarily cooled at a cooling rate of 70°C/s or more to a temperature of Bs or lower, and then secondarily cooled at a cooling rate of 20°C/s or less to a temperature of (Bs+Ms)/2 or higher, and then tertiarily cooled at a cooling rate of 30°C/s or more to a temperature range of Ms-20°C to 500°C.

[0101] The hot-rolled steel sheet manufactured according to the above may be rapidly cooled below a temperature at which bainite starts to form (Bs) to suppress formation of ferrite (granular ferrite), and then may be gradually cooled to an intermediate temperature between the bainite start temperature (Bs) and the martensite start temperature (Ms), or a temperature thereabove, so that acicular ferrite or bainitic ferrite may be secured as a matrix structure.

[0102] When primary cooling is performed at a temperature of Bs or lower after the hot rolling is completed, if a cooling rate is less than 70°C/s, there may be a problem in that a ferrite phase is excessively formed during cooling. In this case, an upper limit of a primary cooling rate is not particularly limited, but when a steel sheet is cooled too excessively, there may be a risk that a shape of the sheet may be distorted, so the upper limit thereof may be limited to 200°C/s or less.

[0103] A lower limit of a cooling end temperature during the primary cooling is not particularly limited, but when it is excessively lowered, there may be a risk that a cooling time during subsequent secondary cooling may not be sufficient, so it has been revealed that the lower limit thereof may be limited to Bs-100°C.

[0104] When the temperature of the hot-rolled steel sheet is Bs or lower by the primary cooling, quenching may be terminated, and secondary cooling may be performed at a cooling rate of 20°C/s or less and a temperature of (Bs+Ms)/2 or higher.

[0105] While the primarily-cooled hot-rolled steel sheet is cooled from a temperature, which is primarily cooled, to a target temperature for secondary cooling, growth of bainitic ferrite and diffusion of carbon into untransformed austenite occur. In particular, in order to obtain a matrix structure and a secondary phase having a target fraction in the present disclosure, it is preferable that the secondary cooling is maintained for a time (ts, seconds) satisfying the following Relational Expression 3.

[0106] In Relational Expressions, k(T) is an indicator of a growth rate of bainitic ferrite, and is affected not only by an alloy composition of steel but also by a phase transformation temperature and grain size after hot rolling. Accordingly, if a value of Relational Expression 3, that is, a relationship between k(T) and holding time ($\exp(-k(T) \times (ts)^2)$) is less than 0.1, a fraction of the matrix structure becomes excessive, elongation may be excellent, but a target strength may not be secured. On the other hand, if the value exceeds 0.3, there is a problem of deterioration of elongation.

[Relational Expression 3]

$$0.1 \leq \exp(-k(T) \times (ts)^2) \leq 0.3$$

[0107] The k(T) is expressed by the following Relational Expression, and each element is a weight content. In the Relational Expression below, T1 represents a primary cooling end temperature (°C), and T2 represents a secondary cooling end temperature (°C).

$$k(T)=\frac{20}{0.049Du-34.2}\exp\left\{-\left(\frac{(T1+T2)/2-557+320[C]+35[Si]+90[Mn]+70[Cr]+120[Mo]+7800[B]}{112}\right)^{1.92}\right\}$$

[0108] While secondary cooling is performed according to the above-described conditions, a temperature of the steel sheet may increase due to transformation heat generation due to bainite phase transformation. In this case, since the dislocation density may be excessively reduced due to excessive heat generation, a cooling rate during secondary cooling may be controlled to 20°C/s or less to minimize an increase in the temperature of the steel sheet due to transformation heat generation. If the cooling rate exceeds 20°C/s, there may be a risk that a shape of the sheet may be distorted. It should be noted that the secondary cooling also includes an air cooling process.

[0109] It is preferable that the hot-rolled steel sheet that has secondarily cooled as described above is tertiary cooled at a cooling rate of 30°C/s or more to a temperature range of Ms-20°C to 500°C, and then wound at the temperature.

[0110] While the secondary cooling is performed, austenite is stabilized and a Ms temperature further decreases. Therefore, the tertiary cooling end temperature, that is, a winding temperature may be applied to be lower than Ms, and is a fraction of bainitic ferrite is formed at 70% or more, cooling may be performed up to a temperature of Ms-20°C.

[0111] During the tertiary cooling, transformation of low-temperature bainite is performed, and depending on a carbon content within austenite, a portion thereof may be transformed into martensite even after winding. Therefore, by setting a cooling rate during the tertiary cooling at 30°C/s or more during the tertiary cooling, formation of additional high-temperature bainite during cooling can be avoided. An upper limit of the cooling rate is not particularly limited, but cooling may be performed at a cooling rate of 100°C or less to prevent distortion of the plate shape.

[0112] Meanwhile, when the colling end temperature, that is, a winding temperature exceeds 500°C, pearlite is easily formed and the dislocation density within acicular ferrite or bainitic ferrite constituting a matrix structure decreases excessively, so that there may be a risk that yield strength is inferior.

[0113] In the present disclosure, Bs and Ms may be may be derived by the formula below, and each element means a weight content.

$$Bs(°C) = 830 - (320 \times [C]) - (90 \times [Mn]) - (35 \times [Si]) - (70 \times [Cr]) - (120 \times [Mo])$$

$$Ms(°C) = 550 - (330 \times [C]) - (41 \times [Mn]) - (20 \times [Si]) - (20 \times [Cr]) - (10 \times [Mo]) + (30 \times [Al])$$

[Final cooling]

[0114] After completing the cooling and winding processes according to the above, a target hot-rolled steel sheet can be obtained by final cooling. In this case, final cooling may be completed by air cooling to room temperature.

[0115] Meanwhile, the hot-rolled steel sheet of the present disclosure obtained by completing final cooling as described above may be additionally pickled and oiled.

[0116] In addition, a hot-dip galvanizing process may be performed by heating the pickled and oiled hot-rolled steel sheet to a temperature range of 450 to 740°C.

[0117] The hot-dip galvanizing process may be performed using a zinc-based plating bath, and there is no particular limitation on an alloy composition in the zinc-based plating bath.

[0118] Hereinafter, the present disclosure will be specifically described through the following Examples. However, it should be noted that the following examples are only for describing the present disclosure by illustration, and not intended to limit the right scope of the present disclosure. The reason is that the right scope of the present disclosure is determined by the matters described in the claims and reasonably inferred therefrom.

Mode for Invention

(Example)

[0119] A steel slab having the composition shown in Table 1 below was prepared, and in this case, remaining components of each steel slab were Fe and unavoidable impurities.

[0120] Each prepared steel slab was reheated at 1200°C, and then hot-rolled, cooled, wound, and finally cooled (air-cooled) under the conditions disclosed in Table 2 below to prepare a hot-rolled steel sheet having a thickness of 2.5 mm. During the hot rolling, a total reduction rate of two final passes was equally applied at 25%, and a cooling rate during tertiary cooling was uniformly applied at 35°C/s.

[0121] Mechanical properties of each hot-rolled steel sheet were measured and a microstructure thereof was observed, and the results were shown in Table 3 below.

[0122] Among the mechanical properties, yield strength, tensile strength, and elongation were measured at room temperature using a universal tensile tester after collecting JIS-5 standard test specimens in a direction, perpendicular to a rolling direction. In this case, the yield strength, tensile strength, and elongation were expressed as 0.2% off-set yield strength, maximum tensile strength, and elongation at break, respectively.

[0123] In addition, hole expandability was measured according to the ISO TS16630 standard method on the same specimen as during a tensile test.

[0124] In addition, a microstructure of each hot-rolled steel sheet was observed at 10,000x magnification using a scanning electron microscope and an image analyzer after etching the same specimen as during the tensile test using a Nital etching method, and the microstructure was observed at 10,000 magnification using a scanning electron microscope and image analyzer, and a fraction of each phase was calculated. In this case, the microstructure was observed on a cross-section of the specimen, that is, a cross-section perpendicular to the rolling direction.

[Table 1]

Steeltype	Alloy composition (weight %)												Relational Expression 1
	C	Si	Mn	P	S	Sol. Al	Cr	Mo	Ti	Nb	B	N	
1	0.0 7	1 . 2	2 . 4	0.0 12	0.0 03	0.03	0. 2	0	0.1 2	0	0	0.0 04	42.9
2	0.0 7	0. 7	2. 1	0.0 11	0.0 03	0.02	0. 5	0	0.0 8	0	0	0.0 04	32.7
3	0.0 7	1. 1	2. 1	0.0 12	0.0 03	0.03	0. 9	0	0.0 6	0.0 3	0	0.0 04	45.9
4	0.0 9	1 . 3	2. 1	0.0 13	0.0 03	0.03	0	0	0.0 3	0	0	0.0 04	24.3
5	0.09	1.3	2.2	0.012	0.003	0.03	0.2	0	0.12	0	0	0.005	35.0
6	0.0 8	0. 9	2. 5	0.0 12	0.0 03	0.03	0	0.1	0.0 9	0	0	0.0 04	26.8
7	0.1 2	1 . 2	1. 8	0.0 12	0.0 03	0.03	0. 2	0.1	0.0 8	0	0	0.0 04	22.9
8	0.1 2	0. 3	1. 8	0.0 12	0.0 03	0.03	0. 9	0.2	0.1 2	0	0	0.0 04	21.9
9	0.0 7	0. 6	2. 1	0.0 11	0.0 03	0.03	0. 5	0	0.0 8	0	0.00 15	0.0 04	30.6

(continued)

Steel type	Alloy composition (weight %)												Relational Expression 1
	C	Si	Mn	P	S	Sol. Al	Cr	Mo	Ti	Nb	B	N	
10	0.17	1.1	2.1	0.012	0.003	0.03	0.9	0	0.06	0.03	0	0.004	18.9
11	0.09	0.8	2.1	0.012	0.003	0.03	0.2	0.15	0	0	0	0.004	17.2
12	0.09	0	2.5	0.012	0.003	0.03	0.3	0	0.09	0.015	0	0.004	12.0
13	0.07	0.7	2.1	0.013	0.003	0.03	0.5	0	0.08	0.015	0	0.005	32.7
14	0.09	1.3	2.2	0.012	0.003	0.03	0.2	0	0.12	0	0	0.004	35.0
15	0.09	1.3	2.1	0.011	0.003	0.02	0	0	0.03	0	0	0.004	24.3
16	0.07	0.7	2.1	0.012	0.003	0.03	0.5	0	0.08	0.015	0	0.004	32.7
17	0.06	1.3	1.8	0.012	0.003	0.03	0.9	0	0	0.015	0	0.004	50.5

[Table 2]

Steel type	Hot rolling		Primary cooling			Secondary cooling					Tertiary cooling (winding)	Reference
	FDT (°C)	Relational Expression 2	End temperature (°C)	Cooling rate (°C/s)	Bs (°C)	End temperature (°C)	Cooling rate (°C/s)	(Bs+Ms)/2 (°C)	Relational Expression 3	Time (s)	End temperature (°C)	
1	925	816.6	500	75	536	480	4.0	468.5	0.17	5.0	420	Inventive Example 1
2	925	851.6	520	75	559	500	4.0	488	0.22	5.0	450	Inventive Example 2
2	920	846.6	520	75	559	520	공냉	488	0.20	6.0	430	Inventive Example 3
2	925	851.6	520	75	559	500	4.0	488	0.22	5.0	450	Inventive Example 4
2	925	851.6	520	75	559	500	3.6	488	0.16	5.5	420	Inventive Example 5
3	925	818.1	500	75	517	480	2.9	459.5	0.17	7.0	450	Inventive Example 6
4	920	861.9	520	75	567	500	4.0	488	0.16	5.0	420	Inventive Example 7
4	920	861.9	520	75	567	500	4.0	488	0.16	5.0	430	Inventive Example 8
5	920	810.2	510	75	544	480	6.0	472.5	0.13	5.0	420	Inventive Example 9
6	925	840.7	500	75	536	480	4.0	469.5	0.23	5.0	430	Inventive Example 10

(continued)

Steel type	Hot rolling		Primary cooling			Secondary cooling					Tertiary cooling (winding)	Reference
	FDT (°C)	Relational Expression 2	End temperature (°C)	Cooling rate (°C/s)	Bs (°C)	End temperature (°C)	Cooling rate (°C/s)	(Bs+Ms)/2 (°C)	Relational Expression 3	Time (s)	End temperature (°C)	
7	920	836.7	520	75	562	500	4.0	485.5	0.16	5.0	430	Inventive Example 11
8	910	823.3	500	75	532	480	4.0	472	0.23	5.0	450	Inventive Example 12
9	925	854.1	520	75	563	500	3.6	491.5	0.25	5.5	450	Inventive Example 13
10	925	818.8	500	75	483	480	2.9	427	0.58	7.0	450	Comparative Example 1
11	925	894.5	520	75	552	500	4.0	483	0.39	5.0	450	Comparative Example 2
12	925	845.7	520	75	555	500	3.3	484	0.14	6.0	450	Comparative Example 3
13	870	780.0	520	75	559	500	5.0	488.5	0.16	4.0	430	Comparative Example 4
14	890	780.2	510	75	544	480	7.5	472.5	0.17	4.0	420	Comparative Example 5
15	920	861.9	520	75	567	500	6.7	488	0.52	3.0	420	Comparative Example 6
16	920	830.0	520	75	559	500	6.7	488.5	0.53	3.0	450	Comparative Example 7
17	890	829.7	520	75	540	520	공냉	465.5	0.28	7.0	510	Comparative Example 8

[Table 3]

Reference	Microstructure					Mechanical properties			
	AF/BF (Area %)	LB+TM+MA (Area %)	MA (Area %)	P (Area %)	MA ratio (%)	YS (MPa)	TS (MPa)	El (%)	HER (%)
Inventive Example 1	83	17	4	0	24	893	1140	12	37
Inventive Example 2	78	22	4	0	18	814	1087	12	45
Inventive Example 3	80	20	4	0	20	805	1069	13	45
Inventive Example 4	78	22	4	0	18	814	1087	12	45
Inventive Example 5	84	16	3	0	19	821	1065	13	46
Inventive Example 6	83	17	4	0	24	836	1083	13	39
Inventive Example 7	84	16	3	0	19	823	1068	13	46

(continued)

Reference	Microstructure					Mechanical properties			
	AF/BF (Area %)	LB+TM+ MA (Area %)	MA (Area %)	P (Area %)	MA ratio (%)	YS (MPa)	TS (MPa)	El (%)	HER (%)
Inventive Example 8	84	16	3	0	18	806	1052	14	47
Inventive Example 9	87	13	3	0	23	847	1076	14	37
Inventive Example 10	77	23	5	0	22	913	1182	11	40
Inventive Example 11	84	16	4	0	25	808	1055	14	44
Inventive Example 12	77	23	6	0	26	882	1156	11	38
Inventive Example 13	75	25	5	0	20	837	1123	11	43
Comparative Example 1	42	49	12	9	24	726	924	15	42
Comparative Example 2	61	32	5	7	16	715	896	18	43
Comparative Example 3	86	8	2	6	25	654	875	17	39
Comparative Example 4	84	16	6	0	38	824	1077	13	26
Comparative Example 5	83	17	6	0	35	887	1136	13	10
Comparative Example 6	48	52	10	0	19	1076	1465	6	34
Comparative Example 7	47	53	12	0	23	1047	1449	7	32
Comparative Example 8	72	25	6	3	24	709	983	14	36
AF: Acicular Ferrite BF: Bainitic Ferrite LB: Low-temperature Bainite TM: Tempered Martensite									
MA: Martensite-Austenite composite phase P: Pearlite YS: Yield Strength TS: Tensile Strength El: Elongation HER: Hole Expansion Rate									

[0125] As shown in Tables 1 to 3, Inventive Examples 1 to 13, which satisfy all of the alloy composition and manufacturing conditions proposed in the present disclosure, as acicular ferrite or bainitic ferrite was sufficiently formed as a matrix structure, and a low-temperature transformation structure (LB+TM+MA) was properly formed as a second phase, target strength and formability could be secured.

[0126] On the other hand, in Comparative Examples 1 to 3, which did not satisfy the alloy composition system (Relational Expression 1) proposed in the present disclosure, the target strength could not be secured due to formation of a large amount of pearlite in the microstructure, and due to this low strength, the elongation tended to be relatively high.

[0127] Meanwhile, in Comparative Examples 4 to 8, in which the alloy composition satisfies the present disclosure, but the manufacturing conditions deviate from the present disclosure.

[0128] Thereamong, in Comparative Examples 4 and 5, in which a rolling end temperature during hot rolling was too low, deviating from Relational Expression 2, resulting in excessive formation of a MA phase, and as a result thereof, the hole expandability was poor.

[0129] In Comparative Examples 6 and 7, a matrix structure was not sufficiently formed due to an insufficient cooling time during the secondary cooling during the cooling process after hot rolling, and as a result thereof, the elongation was inferior.

[0130] In Comparative Example 8, in which a winding temperature was significantly high, as dislocation density in the matrix structure was lost, the yield strength was inferior.

[0131] FIG. 1 is a graph illustrating classification of a phase type of a second phase according to a size of the second phase of each hot-rolled steel sheet.

[0132] As illustrated in FIG. 1, it can be seen that the smaller the size of the second phase, the higher the tendency of the second phase to remain as a MA phase due to carbon overconcentration.

[0133] FIG. 2 is a photograph of the microstructure of Inventive Example 4 and Comparative Example 3 observed with a scanning microscope.

[0134] As illustrated in FIG. 2, it can be seen that in Inventive Example 4(a), as the microstructure, the matrix structure and the second phase to be implemented in the present disclosure are appropriately formed, while in Comparative Example 3(b), pearlite, which is not intended for the present disclosure, was excessively generated.

[0135] While example embodiments have been illustrated 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 hot-rolled steel sheet having excellent formability comprising, by weight:

   carbon (C): 0.05 to 0.17%, silicon (Si): 0.01 to 1.5%, manganese (Mn): 1.5 to 3.0%, aluminum (Al): 0.01 to 0.1%, chromium (Cr): 2.0% or less (including 0%), molybdenum (Mo): 2.0% or less (including 0%), titanium (Ti): 0.01 to 0.15%, phosphorus (P): 0.001 to 0.05%, sulfur (S): 0.0001 to 0.05%, nitrogen (N): 0.0001 to 0.02%, with a balance of Fe and other unavoidable impurities, wherein the following Relational Expression 1 is satisfied, wherein a microstructure comprises 70 to 90% of acicular ferrite or bainitic ferrite by area fraction, as a matrix structure, and one or more second phases among low-temperature bainite, tempered martensite, and MA phase,

   [Relational Expression 1]

   $$(1.5\times[Si] + 1.2\times[Cr] + 0.7\times[Mo] + 8.0\times[Ti])/[C] > 20$$

   In Relational Expression 1, each element refers to a weight content.

2. The high-strength hot-rolled steel sheet having excellent formability of claim 1, wherein the hot-rolled steel sheet further comprises, one or more of niobium (Nb): 0.01 to 0.1%, and boron (B): 0.0005 to 0.005%.

3. The high-strength hot-rolled steel sheet having excellent formability of claim 1, wherein a ratio of a MA phase among a total fraction of the second phase is less than 30%.

4. The high-strength hot-rolled steel sheet having excellent formability of claim 1, wherein the hot-rolled steel sheet comprises at least one of ferrite and carbide of less than 5% by area fraction.

5. The high-strength hot-rolled steel sheet having excellent formability of claim 1, wherein the hot-rolled steel sheet has a yield strength of 750MPa or more, a tensile strength of 980MPa or more, and elongation of 9% or more.

6. The high-strength hot-rolled steel sheet having excellent formability of claim 1, wherein the hot-rolled steel sheet has a hole expansion rate of 30% or more.

7. A manufacturing method for a high-strength hot-rolled steel sheet having excellent formability, comprising operations

of:

reheating a steel slab including, by weight: carbon (C): 0.05 to 0.17%, silicon (Si): 0.01 to 1.5%, manganese (Mn): 1.5 to 3.0%, aluminum (Al): 0.01 to 0.1%, chromium (Cr): 2.0% or less (including 0%), molybdenum (Mo): 2.0% or less (including 0%), titanium (Ti): 0.01 to 0.15%, phosphorus (P): 0.001 to 0.05%, sulfur (S): 0.0001 to 0.05%, nitrogen (N): 0.0001 to 0.02%, with a balance of Fe and other unavoidable impurities, the steel slab satisfying the following Relational Expression 1 to a temperature within a range of 1100 to 1350°C;
hot rolling the reheated steel slab to manufacture a hot-rolled steel sheet;
primarily cooling the hot-rolled steel sheet to a temperature of Bs or lower at a cooling rate of 70°C/s or more;
secondarily cooling the steel sheet to a temperature of (Bs+Ms)/2 or higher at a cooling rate of 20 °C/s or less after the primary cooling;
tertiarily cooling the steel sheet to a temperature range of Ms-20°C to 500°C at a cooling rate of 30°C/s or more after the secondary cooling; and
winding the steel sheet in the tertiarily-cooled temperature range,
wherein finish hot rolling is performed to satisfy the following Relational Expression 2 within a temperature range of 750 to 1150°C during the hot rolling, and a total reduction amount of two final passes is 10 to 40%,

[Relational Expression 1]

$$(1.5\times[Si] + 1.2\times[Cr] + 0.7\times[Mo] + 8.0\times[Ti])/[C] > 20$$

In Relational Expression 1, each element refers to a weight content,

[Relational Expression 2]

$$800 \le Du \le 1106$$

In Relational Expression 2, Du is an indicator indicating an effective grain size of austenite immediately before primary cooling after hot rolling, expressed as Du = FDT + (7.35×[C]) - (24.7×[Si]) - (4.7×[Mn]) - (3.9×[Cr]) - (5.2×[Mo]) - (560×[Ti]) - (1110×[Nb]), where FDT refers to a rolling end temperature (°C), and each element refers to a weight content.

8. The manufacturing method for a high-strength hot-rolled steel sheet having excellent formability of claim 7, wherein the steel slab further comprises, one or more of niobium (Nb): 0.01 to 0.1% and boron (B): 0.0005 to 0.005%.

9. The manufacturing method for a high-strength hot-rolled steel sheet having excellent formability of claim 7, wherein the secondary cooling step is performed for a time (ts) satisfying the following Relational Expression 3,

[Relational Expression 3]

$$0.1 \le \exp(-k(T) \times (ts)^2) \le 0.3$$

where k(T) is expressed by the following formula, and each element is a weight content, in the formula below, T1 is a primary cooling end temperature (°C), and T2 is a secondary cooling end temperature (°C).

$$k(T)=\frac{20}{0.049Du-34.2}\exp\left\{-\left(\frac{(T1+T2)/2-557+320[C]+35[Si]+90[Mn]+70[Cr]+120[Mo]+7800[B]}{112}\right)^{1.92}\right\}$$

10. The manufacturing method for a high-strength hot-rolled steel sheet having excellent formability of claim 7, further comprising:
a step of finally cooling to room temperature after the winding.

11. The manufacturing method for a high-strength hot-rolled steel sheet having excellent formability of claim 10, further comprising:

a step of pickling and oiling after the final cooling.

12. The manufacturing method for a high-strength hot-rolled steel sheet having excellent formability of claim 11, further comprising:
a step of hot-dip galvanizing after the pickling and oiling.

FIG.1

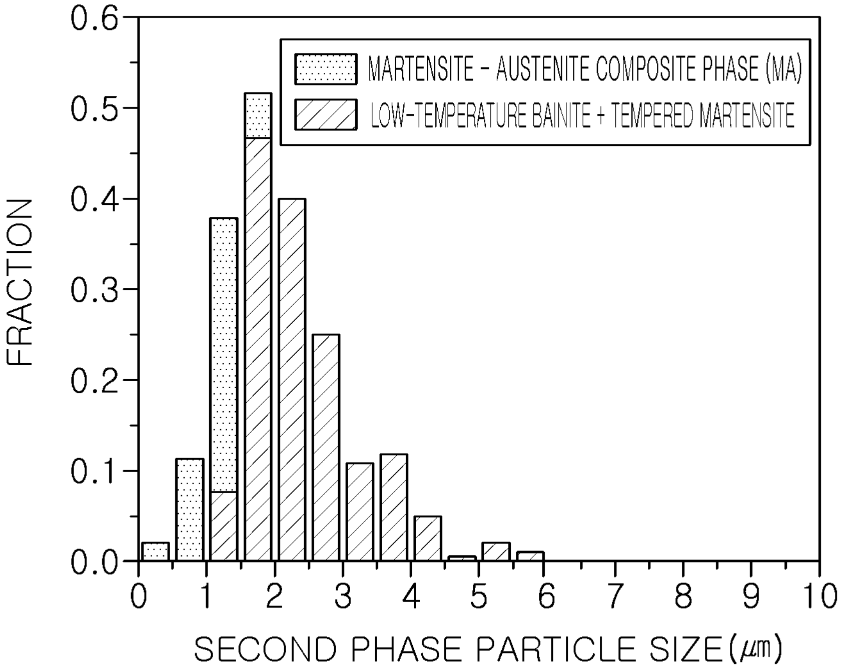
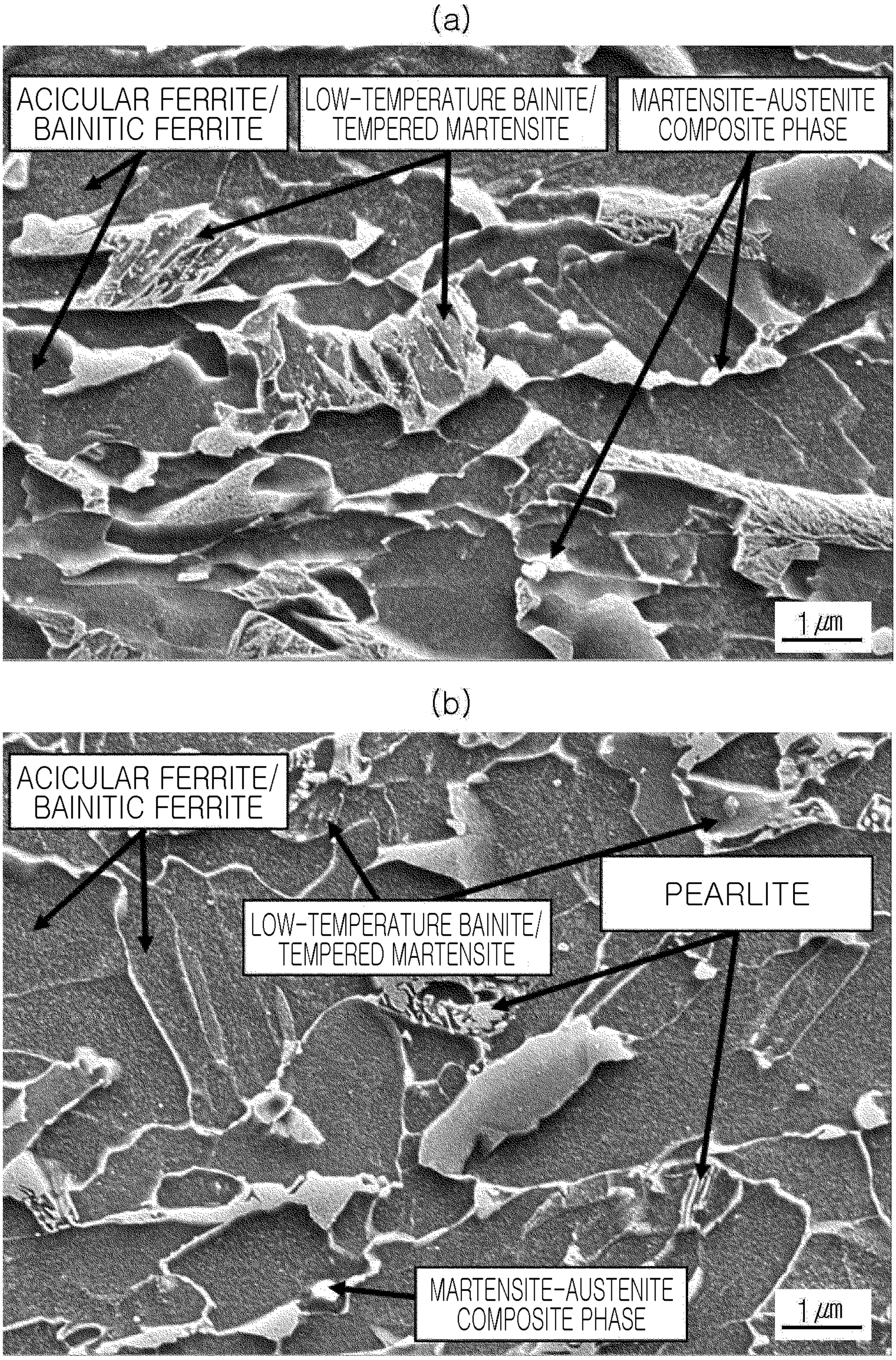

0.6
0.5
0.4
0.3
0.2
0.1
0.0
FRACTION
0
1
2
3
4
5
6
7
8
9
10
SECOND PHASE PARTICLE SIZE(㎛)
MARTENSITE - AUSTENITE COMPOSITE PHASE (MA)
LOW-TEMPERATURE BAINITE + TEMPERED MARTENSITE

FIG.2


(a)
ACICULAR FERRITE/ BAINITIC FERRITE
LOW-TEMPERATURE BAINITE/ TEMPERED MARTENSITE
MARTENSITE-AUSTENITE COMPOSITE PHASE
1 µm
(b)
ACICULAR FERRITE/ BAINITIC FERRITE
LOW-TEMPERATURE BAINITE/ TEMPERED MARTENSITE
PEARLITE
MARTENSITE-AUSTENITE COMPOSITE PHASE
1 µm

INTERNATIONAL SEARCH REPORT

International application No.
PCT/KR2022/011282

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/38(2006.01)i; **C22C 38/22**(2006.01)i; **C22C 38/28**(2006.01)i; **C22C 38/00**(2006.01)i; **C22C 38/60**(2006.01)i; **C22C 38/06**(2006.01)i; **C22C 38/26**(2006.01)i; **C22C 38/32**(2006.01)i; **C21D 8/02**(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(2006.01); B32B 15/01(2006.01); C21D 8/00(2006.01); C21D 8/04(2006.01); C21D 9/52(2006.01); C22C 38/00(2006.01); C22C 38/02(2006.01); C22C 38/04(2006.01); C22C 38/14(2006.01); C22C 38/22(2006.01)

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: 베이니틱 페라이트(bainitic ferrite), 마르텐사이트(martensite), 열연강판(hot-rolled steel plate), 베이나이트(bainite), MA상(MA)

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
DX DA	JP 2008-255484 A (KOBE STEEL LTD.) 23 October 2008 (2008-10-23) See paragraphs [0025], [0060]-[0066] and [0075]-[0078], claims 1-2 and 4 and tables 1-3.	1-6 7-12
A	WO 2020-079096 A1 (TATA STEEL NEDERLAND TECHNOLOGY B.V.) 23 April 2020 (2020-04-23) See page 1, lines 18-29 and page 16, lines 28-29, claims 1 and 11 and figures 1-2.	1-12
A	KR 10-2015-0040814 A (TATA STEEL IJMUIDEN B.V.) 15 April 2015 (2015-04-15) See claims 1 and 10 and figure 1.	1-12
A	KR 10-2015-0112490 A (HYUNDAI STEEL COMPANY) 07 October 2015 (2015-10-07) See claims 1-6.	1-12
A	US 10876184 B2 (TATA STEEL LIMITED) 29 December 2020 (2020-12-29) See claims 1 and 9.	1-12

☐ 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
"D" document cited by the applicant in the international application
"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 **27 October 2022**	Date of mailing of the international search report **28 October 2022**
Name and mailing address of the ISA/KR **Korean Intellectual Property Office** **Government Complex-Daejeon Building 4, 189 Cheongsa-ro, Seo-gu, Daejeon 35208** Facsimile No. **+82-42-481-8578**	Authorized officer Telephone No.

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INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/KR2022/011282

Patent document cited in search report	Publication date (day/month/year)	Patent family member(s)	Publication date (day/month/year)
JP 2008-255484 A	23 October 2008	CN 101265553 A	17 September 2008
		CN 101265553 B	19 January 2011
		GB 2448114 A	08 October 2008
		GB 2448114 B	12 May 2010
		JP 5172403 B2	27 March 2013
		KR 10-0987877 B1	13 October 2010
		KR 10-2008-0084667 A	19 September 2008
		US 2008-0223491 A1	18 September 2008
		US 8052808 B2	08 November 2011
WO 2020-079096 A1	23 April 2020	EP 3867417 A1	25 August 2021
		US 2021-0310093 A1	07 October 2021
KR 10-2015-0040814 A	15 April 2015	BR 112015000178 A2	27 June 2017
		BR 112015000178 B1	17 March 2020
		CA 2880063 A1	06 February 2014
		CA 2880063 C	14 March 2017
		CN 104520449 A	15 April 2015
		CN 104520449 B	14 December 2016
		EP 2880189 A1	10 June 2015
		EP 2880189 B1	21 September 2016
		ES 2608036 T3	05 April 2017
		JP 2015-528858 A	01 October 2015
		JP 6359534 B2	18 July 2018
		MX 2014015984 A	20 March 2015
		MX 356495 B	31 May 2018
		US 10053757 B2	21 August 2018
		US 2015-0191807 A1	09 July 2015
		WO 2014-019844 A1	06 February 2014
		ZA 201409114 B	28 September 2016
KR 10-2015-0112490 A	07 October 2015	None	
US 10876184 B2	29 December 2020	EP 3436613 A1	06 February 2019
		EP 3436613 B1	27 May 2020
		JP 2019-508572 A	28 March 2019
		JP 6692429 B2	13 May 2020
		KR 10-2019-0008062 A	23 January 2019
		US 2018-0187283 A1	05 July 2018
		WO 2017-168436 A1	05 October 2017

REFERENCES CITED IN THE DESCRIPTION

Patent documents cited in the description

- JP 2008255484 A **[0010]**