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

(11) **EP 4 428 262 A1**

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

(43) Date of publication:
11.09.2024 Bulletin 2024/37

(21) Application number: **22890363.9**

(22) Date of filing: **02.11.2022**

(51) International Patent Classification (IPC):
C22C 38/38 *(2006.01)* ***C22C 38/22*** *(2006.01)*
C22C 38/32 *(2006.01)* ***C22C 38/28*** *(2006.01)*
C21D 8/02 *(2006.01)* ***C21D 9/46*** *(2006.01)*
C23C 2/06 *(2006.01)*

(52) Cooperative Patent Classification (CPC):
C21D 8/02; C21D 9/46; C22C 38/22; C22C 38/28; C22C 38/32; C22C 38/38; C23C 2/06

(86) International application number:
PCT/KR2022/017031

(87) International publication number:
WO 2023/080632 (11.05.2023 Gazette 2023/19)

(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 ME MK MT NL NO PL PT RO RS SE SI SK SM TR
Designated Extension States:
BA
Designated Validation States:
KH MA MD TN

(30) Priority: **05.11.2021 KR 20210151262**

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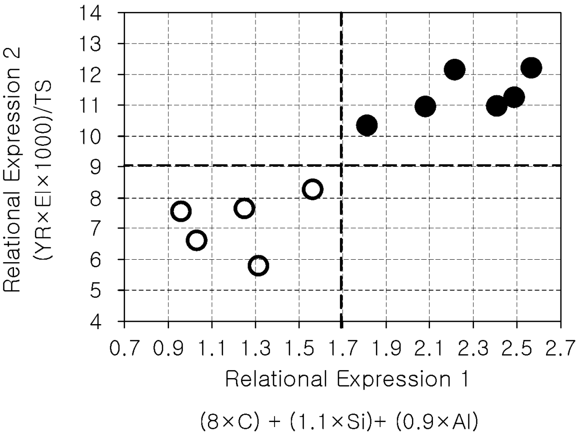
(54) HIGH-STRENGTH STEEL SHEET HAVING EXCELLENT CRASHWORTHINESS AND FORMABILITY, AND METHOD FOR MANUFACTURING SAME

(57) The present invention relates to a high-strength steel sheet used for a vehicle structure member and, more specifically, to a high-strength steel sheet having excellent crashworthiness and formability and a method for manufacturing same.

[FIG. 1]



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Description

Technical Field

[0001] The present disclosure relates to a high-strength steel sheet used for a vehicle structure member, and more specifically, to a high-strength steel sheet having excellent crashworthiness and formability, and a method for manufacturing the same.

Background Art

[0002] Recently, as environmental and safety regulations have become increasingly harsh in the automobile industry, carbon dioxide emission regulations and fuel efficiency regulations have also been increasingly strengthened. The Insurance Institute for Highway Safety in the United States has been increasingly tightening collision stability regulations to protect passengers, and since 2013, severe collision performance such as 25% small overlap has been required.

[0003] The only solution to address these environmental and safety issues is automotive weight reduction. High strength of steel is required to achieve weight reduction, and high formability is required to apply high-strength steel. In addition, in order to simultaneously increase the collision performance of the vehicle body, a yield ratio (YS/TS) is required to be high.

[0004] In general, high-strength vehicle materials may be classified into precipitation reinforced steel, bake hardened steel, solid solution strengthened steel, and transformation reinforced steel.

[0005] The transformation reinforced steel includes Dual Phase Steel (DP steel), Complex Phase Steel (CP steel), or Transformation Induced Plasticity Steel (TRIP steel). The transformation reinforced steel is also referred to as Advanced High Strength Steel (ASS).

[0006] Thereamong, DP steel is a steel that secures high strength by finely and homogeneously dispersing a hard martensite in a soft ferrite, and CP steel is a steel that includes two or three phases of a ferrite, a martensite, and a bainite, and adds precipitation hardening elements such as Ti and Nb to improve strength. TRIP steel is a steel that includes fine and homogeneously dispersed retained austenite, and may ensure high strength and high ductility by causing a retained austenite phase to be transformed into a martensite during room temperature processing.

[0007] On the other hand, recent steel sheets for vehicles require steel sheets with higher strength to improve fuel efficiency and durability, and in terms of collision safety and passenger protection, ultra-high-strength steel sheets with a tensile strength of 980 MPa or higher are increasingly being applied as body structures or reinforcement materials.

[0008] Specifically, high-strength steel having excellent yield strength is adopted in structural members such as members, seat rails, and pillars, in order to improve crashworthiness of a vehicle body. As yield strength (YS), that is, a yield ratio (YR=YS/TS), compared to tensile strength (TS), becomes higher and higher, a structural member is more advantageous for crashworthiness characteristics.

[0009] However, in general, ductility decreases as the strength of the steel sheet increases, which may cause a problem of deteriorating molding processability, and thus, the development of a material capable of supplementing the problem is required. In other words, it may be essential to develop a material with high yield strength and excellent ductility to simultaneously secure collision stability and component formability.

[0010] Furthermore, most of the processed components have a shear surface therein or at an edge thereof, and accordingly, when the Hall Expansion Station (HER) is excellent, the components may be molded without defects such as cracks in the shear surface, and even in the event of collisions, collision performance may be improved through impact absorption.

[0011] Accordingly, when it is possible to secure excellent ductility while increasing a low yield ratio and an inferior hole expansion ratio, which are shortcomings of the most widely used DP steel among the transformation reinforced high-strength steel, in order to improve the impact characteristics and simultaneously of high-strength steel, an application of high-strength steel could be further expanded.

[0012] On the other hand, in order to increase the yield ratio of the steel, it is necessary to increase yield strength compared to tensile strength, and a representative way to achieve the increased yield strength is to use water cooling during continuous annealing. As an example, a steel sheet in which a microstructure has a tempered martensite phase may be manufactured through a tempering process after forming a martensite by performing cracking in an annealing process and then depositing in water.

[0013] However, in the method, since a shape quality is deteriorated due to temperature variations in a width direction and a length direction during water cooling, there may be very serious disadvantages such as deterioration of workability such as cracks during molding and material deviation by location.

[0014] Patent Document 1, which is a prior art related to the aforementioned technology, discloses martensite steel with a volume ratio of 80 to 97% of martensite, obtained by continuously annealing a steel material containing 0.18% or more of carbon (C) and then cooling the steel material to room temperature, and then treating over-aging on the steel material at a temperature of 120 to 300°C for 1 to 15 minutes. As described above, when ultra-high strength steel is manufactured by a water cooling and tempering manner, the yield ratio is very high, but the shape quality of a coil may be deteriorated due to temperature variations in a width direction and a length direction, resulting in problems such as an occurrence of cracks and workability reduction occur during molding processing.

[0015] Patent Document 2 relates to a steel sheet formed of a complex tissue mainly composed of martensite, and discloses a method of manufacturing a high-tensile steel sheet in which fine precipitated copper particles having a particle diameter of 1to 100 nm are dispersed inside the tissue to improve processability. However, by adding Cu in excess of 2 to 5% to precipitate fine copper particles, there may be a concern that hot shortness due to Cu may occur, which may excessively increase manufacturing costs.

[0016] Patent Document 3 relates to a precipitation-reinforced steel sheet containing 2 to 10% of pearlite with ferrite as a base structure, which adds carbon/nitride forming elements such as Nb, Ti, V, and the like, thereby improving strength by precipitation reinforcement and grain refinement. The steel sheet has a good hole expansion ratio, but has a limitation to increasing tensile strength, and also has high yield strength and low ductility, resulting in an occurrence of cracks during press molding.

[0017] Patent Document 4 discloses a method of manufacturing a cold-rolled steel sheet with excellent plate shape after continuous annealing and securing high strength and high ductility at the same time by utilizing a tempered martensite, but since a carbon content is as high as 0.2% or more, the weldability may be degraded, and dent defects in a furnace may occur due to a large amount of Si.

[Prior Art Document]

[0018]

(Patent Document 1) Japanese Unexamined Patent Application Publication No. 1992-289120
(Patent Document 2) Japanese Unexamined Patent Application Publication No. 2005-264176
(Patent Document 3) Korean Unexamined Patent Application Publication No. 2015-0073844
(Patent Document 4) Japanese Unexamined Patent Application Publication No. 2010-090432

Summary of Invention

Technical Problem

[0019] An aspect of the present disclosure is to provide a steel sheet used for a vehicle structure member, that is, a steel sheet having excellent strength as well as ductility and having improve crashworthiness and formability, and a method for manufacturing the same.

[0020] The aspect of the present disclosure is not limited to the above. The aspect of the present disclosure may be understood from overall contents of the present specification, and those of ordinary skill in the technical field to which the present disclosure pertains will have no difficulty in understanding an additional aspect of the present disclosure.

Solution to Problem

[0021] According to an aspect of the present disclosure, a high-strength steel sheet having excellent crashworthiness and formability may include: by wt%, carbon (C): 0.06 to 0.2%, silicon (Si): 0.4 to 1.4%, manganese (Mn): 1.8 to 3.0%, aluminum acid (Sol.Al): 1.0% or less, molybdenum (Mo): 0.4% or less, chromium (Cr): 1.0% or less, antimony (Sb): 0.06% or less, boron (B): 0.01% or less, phosphorus (P): 0.1% or less, sulfur (S): 0.01% or less, and a balance of Fe and unavoidable impurity elements, wherein the C, Si and Al satisfy the following Relational Expression 1,

[0022] a microstructure includes 40 to 80% of a sum of tempered martensite and bainite phases as an area fraction, 3 to 15% of a retained austenite phase, and a balance of ferrite and fresh martensite, and in the retained austenite phase, a share (A_{TM+B}/A_T) of retained austenite (A_{TM+B}) adjacent to tempered martensite and bainite among a total retained austenite fraction (A_T) is 90% or more,

[Relational Expression 1]

$$(8\times C) + (1.1\times Si) + (0.8\times Al) \geq 1.7$$

(In the relational expression 1, each element denotes a weight content).

[0023] According to another aspect of the present disclosure, a method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability may include: heating a steel slab in a temperature range of 1050 to 1250°C, the steel slab including, by wt%, carbon (C): 0.06 to 0.2%, silicon (Si): 0.4 to 1.4%, manganese (Mn): 1.8 to 3.0%, aluminum acid (Sol.Al): 1.0% or less, molybdenum (Mo): 0.4% or less, chromium (Cr): 1.0% or less, antimony (Sb): 0.06% or less, boron (B): 0.01% or less, phosphorus (P): 0.1% or less, sulfur (S): 0.01% or less, and a balance of Fe, and unavoidable impurity elements, the C, Si and Al satisfying the following Relational Expression 1; manufacturing a hot-rolled steel sheet by finish hot rolling the reheated steel slab at a temperature range of a finish hot rolling outlet-side temperature Ar3 to Ar3+50°C; coiling the hot-rolled steel sheet in a temperature range of 400 to 700°C; after the coiling, cooling the hot-rolled steel sheet to room temperature at cooling rate of 0.1°C/s; after the cooling, manufacturing a cold-rolled steel sheet by cold rolling the hot-rolled steel sheet with a total reduction ratio of 30 to 80%; continuously annealing the cold-rolled steel sheet; primarily cooling the continuously annealed cold-rolled steel sheet at cooling rate of 10°C/s or less by 450 to 700°C; after the primary cooling, performing secondary cooling at cooling rate of 3°C/s or higher by 250 to 500°C; and reheating the secondary cooled cold-rolled steel sheet by a temperature of 490°C or less and maintaining for 30 seconds or more,

wherein the cold rolling is performed at a cumulative reduction ratio of 20% or more for the first 1 to 3 stands.

Advantageous Effects of Invention

[0024] According to an aspect of the present disclosure, a steel sheet having high strength and excellent ductility may be provided. Specifically, since the steel sheet of the present disclosure has a higher yield ratio than that of conventional DP steel, the steel sheet has an excellent hole expansion ratio, thereby having excellent impact resistance and formability.

[0025] The steel sheet of the present disclosure may be suitably applied as a material for a vehicle structure member requiring processing into a complex shape.

Brief Description of Drawings

[0026]

FIG. 1 is a graph showing a change in mechanical properties (Relational Expression 2) according to Relational Expression 1 according to an example embodiment of the present disclosure.

FIG. 2 is a microstructure measurement image of a steel of the present disclosure according to an example embodiment of the present disclosure.

Best Mode for Invention

[0027] The inventors of the present disclosure studied deeply to provide a high-strength steel sheet having improved crashworthiness and formability by increasing a yield ratio (YR) and a hole expansion ratio, as compared to conventional DP steel, while satisfying high ductility, which is a characteristic of the conventional DP steel.

[0028] As a result, the inventors confirmed that by optimizing an alloy component system and manufacturing conditions of the steel, a steel sheet may have an advantageous structure for securing target physical properties, from which it is possible to provide a steel sheet suitable for a vehicle structure member that require processing into a complex shape, and have completed the present disclosure.

[0029] Hereinafter, the present disclosure will be described in detail.

[0030] A high-strength steel sheet having excellent crashworthiness and formability according to an aspect of the present disclosure may include: by wt%, carbon (C): 0.06 to 0.2%, silicon (Si): 0.4 to 1.4%, manganese (Mn): 1.8 to 3.0%, aluminum acid (Sol.Al): 1.0% or less, molybdenum (Mo): 0.4% or less, chromium (Cr): 1.0% or less, antimony (Sb): 0.06% or less, boron (B): 0.01% or less, phosphorus (P): 0.1% or less, sulfur (S): 0.01% or less.

[0031] Hereinafter, a reason for limiting an alloy composition of a steel sheet provided by the present disclosure as above will be described in detail.

[0032] Meanwhile, unless specifically stated in the present disclosure, a content of each element is based on weight, and a ratio of a structure is based on area.

Carbon (C): 0.06 to 0.2%

[0033] Carbon (C) is a greatly important element added to strengthen a transformation structure. The C promotes the high-strength of steel and promotes a formation of martensite in complex structure steel. When the content of C increases, the amount of martensite increases.

[0034] When the content of C exceeds 0.2%, the strength due to the formation of martensite is increased, but a strength difference from ferrite having a low carbon concentration is increased. Such a difference in strength may result in ductility and a process curing ratio because destruction at an interphase interface is easily caused during plastic deformation. Furthermore, welding defects occur during component processing due to poor weldability, and liquid metal embrittlement (LME) cracks occur during welding, leading to undermining component performance. On the other hand, when the content of C is less than 0.06%, it is difficult to secure a target level of strength, and it becomes difficult to secure a certain percentage of retained austenite phase required for ductility.

[0035] Accordingly, in the present disclosure, C may be included in an amount of 0.06 to 0.20%, and more preferably 0.08% or more and 0.18% or less.

Silicon (Si): 0.4 to 1.4%

[0036] Silicon (Si) is a ferrite stabilizing element that promotes a transformation of ferrite and promotes the concentration of carbon (C) into untransformed austenite, thus contributing to the formation of martensite. In addition, the Si is effective in reducing a hardness difference between phases by increasing the strength of ferrite due to an excellent solid solution strengthening ability thereof. Furthermore, the Si effectively inhibits the precipitation of carbides in bainite when maintained in a bainite area, thereby encouraging C concentration into untransformed austenite, and the Si delays the transformation of martensite during cold rapid cooling, thereby forming retained austenite required for ductility, so that the Si is a useful element for improving the ductility of a steel sheet. That is, the Si is a useful element capable of securing strength without lowering the ductility of the steel sheet.

[0037] When the content of Si exceeds 1.4%, it causes surface scale defects to adversely affect plating surface quality, reduces phosphatability, and causes welding defects during component processing due to poor weldability. Specifically, LME cracks occur during welding, which reduces component performance. On the other hand, when the content of Si is less than 0.4%, it may be difficult to secure a certain percentage of retained austenite phase required for ductility, and the strength of ferrite may decrease due to poor solid solution hardenability, so that there is a limit to reducing a difference in hardness between phases.

[0038] Accordingly, in the present disclosure, the Si may be included in an amount of 0.4 to 1.4%, and more preferably may be 0.5% or more and 1.2% or less.

Manganese (Mn): 1.8 to 3.0%

[0039] Manganese (Mn) is an effective element for micronizing particles without decreasing ductility, and strengthening steel as well as preventing hot brittleness caused by FeS generation by completely precipitating sulfur (S) in steel as MnS. In addition, the Mn reduces critical cooling rate at which a martensite phase is obtained in composite phase steel, thereby facilitating the formation of martensite.

[0040] When the content of Mn is less than 1.8%, it may be difficult to ensure target strength in the present disclosure. On the other hand, when the content of Mn exceeds 3.0%, problems such as weldability and hot rolling may be likely to occur, and since martensite is excessively formed, the material is unstable and a manganese oxide band (Mn-Band) is formed in a structure, which may increase the risk of causing defects such as processing cracks and plate breakage. In addition, there may be a problem that Mn oxide may be eluted to a surface during an annealing process to greatly reduce the surface quality.

[0041] Accordingly, in the present disclosure, the Mn may be included in an amount of 1.8 to 3.0%, and more preferably 2.0% or more and 2.9% or less.

Acid-Soluble Aluminum (Sol.Al): 1.0% or less

[0042] Acid-soluble aluminum (Sol.Al) is an element added for refining a particle size of the steel and deoxidation, and is a ferrite stabilizing element similar to Si. The Al is an element useful for improving hardenability of martensite by distributing carbon in ferrite to austenite. Furthermore, the Al may effectively suppress the precipitation of carbides in bainite when maintained in a bainite area during annealing to promote C concentration with untransformed austenite and delay martensite transformation during cold rapid cooling, and may improve the ductility of a steel sheet by generating a retained austenite phase.

[0043] When continuous casting is performed in steelmaking by allowing the content of Al to exceed 1.0%, excessive formation of inclusions may increase the likelihood of surface defects on a steel sheet surface, and may also cause an increase in manufacturing costs. In addition, due to poor weldability, there may be a concern that welding defects may occur during component processing.

[0044] Thus, the Al may be included in an amount of 1.0% or less, excluding 0%. More preferably, the Al may be included in an amount of 0.01% or more.

Molybdenum (Mo): 0.4% or less

[0045] Molybdenum (Mo) is an element delaying the transformation of austenite into pearlite and improving the miniaturization and strength of ferrite. The Mo has an advantage of improving the hardenability of the steel and controlling a yield ratio by finely forming martensite on a grain boundary. However, the Mo is an expensive element, and as a content thereof increase, the manufacturing costs thereof increases, which is economically disadvantageous.

[0046] In order to sufficiently obtain the above-described effects, the Mo may be added in a maximum amount of 0.4%. When the content of Mo exceeds 0.4%, alloy costs are rapidly increased, leading to a decrease in economic feasibility, and also, an effect refining crystal grains and an effect of solid solution strengthening, leading to a decrease in the ductility of the steel.

[0047] Accordingly, in the present disclosure, the content of the Mo may 0.4% or less. According to the present disclosure, even if the Mo is not added, there is no difficulty in securing intended microstructure and physical properties, and thus, the content of the Mo may be 0%. However, when the Mo is added, the Mo may be more advantageously included in the amount of 0.01% or more.

Chromium (Cr): 1.0% or less

[0048] Chromium (Cr) is an element added to improve the hardenability of steel and secure high strength, and plays an important role in the formation of martensite. Furthermore, the Cr is advantageous for manufacturing highly flexible complex phase steel by minimizing the decrease in an elongation ratio as compared to an increase in strength.

[0049] When the content of Cr exceeds 1.0%, there may be a problem in that not only the above-described effects are saturated, but also hot rolling strength is excessively increased to deteriorate cold rolling. Furthermore, since Cr-based carbides are excessively formed and coarsened, there may be a problem in that a size of martensite becomes coarse after annealing, resulting in a decrease in an elongation ratio.

[0050] Accordingly, in the present disclosure, the content of the Cr may be 1.0% or less. In the present disclosure, even if the Cr is not added, there is no difficulty in ensuring intended microstructure and physical properties, and thus, the content of Cr may be 0%. However, when the Cr is added, the Cr may be more advantageously included in the amount of 0.1% or more.

Antimony (Sb): 0.06% or less

[0051] Antimony (Sb) is distributed in a grain boundary to delay a diffusion of oxidizing elements such as Mn, Si, and Al through the grain boundary, thus suppressing a surface concentration of an oxide. Furthermore, the Sb has an excellent effect on suppressing the coarsening of surface agglomerates due to temperature increase and hot-rolled process change. When the content of Sb exceeds 0.06%, the above-described effects may be saturated, manufacturing costs may be increased, and processability may be degraded.

[0052] Accordingly, the content of the Sb may be 0.06% or less. According to the present disclosure, even if the Sb is not added, there is no difficulty in ensuring intended microstructure, physical properties, and the like, and thus, the content of Sb may be 0%. However, when the Sb is added, the Sb may be more advantageously included in an amount of 0.01% or more.

Boron (B): 0.01% or less

[0053] Boron (B) is an element delaying the transformation of austenite into pearlite in a cooling process during annealing, and is a hardenability element inhibiting ferrite formation and promoting martensite formation. When the content of B exceeds 0.01%, the B may be excessively enriched on a surface of steel, which may degrade plating adhesiveness.

[0054] Accordingly, the content of B may be 0.01% or less. According to the present disclosure, even if the B is not added, there is no difficulty in ensuring intended microstructure and physical properties, and thus, the content of B may be 0%. However, when the B is added, the B may be more advantageously included in an amount of 0.0005% or more.

Phosphorus (P): 0.1% or less

[0055] Phosphorus (P) is a substituted element having a large solid solution strengthening effect and is the most advantageous element for improving in-plane anisotropy and securing strength without significantly impairing formability. However, when the content of P is excessively added, the possibility of brittle destruction increases significantly, which may hinder the possibility of plate fracture of a slab during hot rolling and plating surface properties.

[0056] Accordingly, the P may be included in an amount of 0.1% or less, and the amount of 0% may be excluded in consideration of a level inevitably added during a steel manufacturing process.

Yellow (S): 0.01% or less

[0057] Sulfur (S) is an impurity inevitably added to steel and is an element that inhibits ductility and weldability, and thus, it may be advantageous to manage the content of S to be as low as possible. Specifically, since there is a high possibility of generating hot brittleness, it may be preferable to control the content thereof to 0.01% or less. However, 0% may be excluded in consideration of the level inevitably added during a steel manufacturing process.
[0058] In addition to the above-described alloy composition, the steel sheet of the present disclosure may further include at least one of Ti and Nb for the purpose of further improving the mechanical properties of the steel sheet.

Titanium (Ti): 0.05% or less, and Niobium (Nb): 0.05% or less

[0059] Titanium (Ti) and niobium (Nb) are effective elements for increasing the strength of steel and refining crystal grains by forming nano-precipitation hardening. When these elements are added, they may be coupled to carbon to form very fine nano-precipitation hardening, and the nano-precipitation hardening may serve to strengthen a matrix structure and reduce a hardness difference between phases.
[0060] In a case in which the Ti and the Nb are added, when the content thereof exceeds 0.05%, respectively, manufacturing costs may increase, and the precipitation hardening may be excessively formed, resulting in significantly degraded ductility.
[0061] Accordingly, when one or more of Ti and Nb are added, each of them may be included in an amount of 0.05% or less.
[0062] The remaining component of the present disclosure is iron (Fe). However, since impurities that are not intended from raw materials or the surrounding environments may inevitably be mixed in the normal manufacturing process, the impurities may not be excluded. Since these impurities are known to those skilled in the steel manufacturing field, not all of these impurities are specifically mentioned in this specification.
[0063] In a steel sheet of the present disclosure satisfying the alloy composition described above, preferably, a relationship of the content of C, Si and Al may satisfy Relational Expression 1 below.

[Relational Expression 1]

$$(8\times C) + (1.1\times Si) + (0.8\times Al) \geq 1.7$$

(In the relational expression 1, each element denotes a weight content.)
[0064] Si and Al in steel are ferrite stabilizing elements promoting ferrite transformation, and contributing to the formation of retained austenite and martensite by encouraging C concentration to untransformed austenite. C is also an element contributing to the formation and fraction adjustment of martensite by promoting the C concentration to untransformed austenite.
[0065] Specifically, when a value of R relational Expression 1 is controlled to be 1.7 or more, the fraction of retained austenite that can contribute to ductility may be secured, from which it is possible to improve the ductility and formability of the steel sheet. On the other hand, when Relational Expression 1 above is not satisfied, the retained austenite fraction is insufficient, and a distribution of the generated retained austenite is not uniform, which may make it difficult to secure ductility and formability.
[0066] As will be described in detail below, the present disclosure may distribute retained austenite produced by optimizing the steel sheet manufacturing process together with the above-described alloy component system finely around a hard phase, thus mitigating local stress concentration, from which it may be possible to improve ductility thereof and to secure excellent formability.
[0067] The steel sheet of the present disclosure may include 40 to 80% of tempered martensite and bainite phases as a total area fraction, 3 to 15% of a retained austenite phase, and a balance of ferrite and fresh martensite , as a microstructure.
[0068] In the present disclosure, the tempered martensite and bainite structures assist in forming retained austenite in addition to an effect contributing to strength. Specificall y, when Si and Al are added to steel, by delaying the precipitation of carbides during bainite transformation, carbon (C) is accumulated into untransformed austenite around bainite, thereby lowering a martensite transformation temperature below room temperature. In this case, retained austenite may be secured at room temperature. Furthermore, when tempering the martensite, carbon (C) introduced in the martensite moves to surrounding untransformed austenite and is then accumulated, thereby lowering the martensite transformation temperature below room temperature, as well as securing retained austenite at room temperature.

[0069] From this, the steel sheet of the present disclosure may include 3 to 15% of a retained austenite phase. By securing 3% or more of the retained austenite phase, it is advantageous to secure ductility of the steel sheet by causing transformation induced plasticity during molding. However, if a fraction thereof is excessive, the steel sheet tends to be vulnerable to liquid metal brittleness (LME) during point welding to assemble plated steel sheets into vehicle components. Accordingly, preferably, the retained austenite phase may be included in an amount of 15% or less.

[0070] On the other hand, the retained austenite phase is characterized in that a share (A_{TM+B}/A_T) of retained austenite (A_{TM+B}) adjacent to tempered martensite and bainite among a total retained austenite fraction (A_T) is more than 90%.

[0071] Here, being adjacent to the tempered martensite and the bainite may refer to the periphery of their phases, more specifically an interface area of these phases. As an example, as illustrated in FIG. 2, this denotes that retained austenite phases are mainly distributed around a grain boundary on the tempered martensite and the bainite.

[0072] As described above, theretained austenite phases may be finely and evenly distributed around the tempered martensite and the bainite, thereby improving ductility through an effect of relieving local stress concentration, and accordingly, excellent formability without cracks when forming components may be secured.

[0073] When a total fraction of the tempered martensite and the bainite is less than 40%, retained austenite cannot be evenly distributed around the tempered martensite and the bainite, and the amount thereof may also be reduced, which may make it impossible to improve formability. On the other hand, when the fraction thereof exceeds 80%, the fraction of ferrite contributing to ductility may be significantly lowered, and the fraction of the retained austenite may also be lowered, which may make it impossible to improve ductility and formability.

[0074] More advantageously, the tempered martensite phase of the total fraction may be included in an amount of 25 to 65%.

[0075] In the present disclosure, ferrite and fresh martensite phases may be included in addition to the above-described tempered martensite, bainite, and retained austenite phases. In this case, a ferrite phase may be included in an amount of 40% or less, and a fresh martensite phase may be included in an amount of 20% or less. Here, the amount of 0% of the ferrite phase and the fresh martensite phase are excluded.

[0076] When the fraction of the ferrite phase exceeds 40%, not only cannot a target level of strength be secured, but it may be difficult to improve a yield ratio. Furthermore, when the fraction of the fresh martensite exceeds 20%, the fraction of the retained austenite phase may decrease to reduce the ductility, and the formability thereof may not be secured.

[0077] The steel sheet of the present disclosure, which has the above-described alloy component system and microstructure, not only has high strength having a tensile strength of 980 MPa or more, but also has a yield ratio of 0.6 to 0.9, an elongation ratio of 10% or more, and a hole expansion ratio of 20% or more. When the yield ratio is less than 0.6, the hole expansion ratio is degraded, but when the yield ratio exceeds 0.9, the ductility decreases.

[0078] In addition, the steel sheet of the present disclosure may provide a steel sheet having both high yield ratio and high ductility as a relationship between the yield ratio, the elongation ratio, and the tensile strength satisfies the following Relational Expression 2.

[0079] The high yield ratio of the steel sheet may contribute to the improvement of stability during a vehicle collision due to excellent crashworthiness of the material, and the high ductility may prevent processing defects such as cracks and wrinkles that occur during press processing into components, thereby securing excellent formability.

[0080] In the present disclosure, when a value of the following Relational Expression 2 is 9 or more, the crashworthiness and formability may be ensured at the same time, but when the value thereof is less than 9, the crashworthiness and the formability may not be ensured at the same time.

[Relational Expression 2]

$$(YR \times El \times 1000)/TS \geq 9$$

(In the relational expression 2,units of each physical property are excluded.)

[0081] Since the steel sheet of the present disclosure having the mechanical properties as described above may prevent processing defects such as cracks and wrinkles when processing into components, the steel sheet may be used in various ways for a vehicle structure member. In addition, the steel sheep may contribute to improving the safety of structural components and vehicles by delaying an occurrence of crashworthiness and cracks during a collision.

[0082] On the other hand, the steel sheet of the present disclosure may be a cold-rolled steel sheet, and may be a hot-dip galvanized steel sheet including a zinc-based plating layer on at least one surface of the cold-rolled steel sheet, and an alloyed hot-dip galvanized steel sheet alloyed with the hot-dip galvanized steel sheet.

[0083] Although not particularly limited, the zinc-based plating layer may be a zinc plating layer mainly containing zinc and a zinc alloy plating layer containing aluminum and/or magnesium other than zinc.

[0084] Hereinafter, a method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability provided by the present disclosure, another aspect of the present invention, will be described in detail.

[0085] In short, the present disclosure may manufacture a desired steel sheet through a process of [steel slab reheating - hot rolling - coiling - cooling - cold rolling - continuous annealing - cooling - reheating and maintaining], and then further processes such as [hot dip galvanizing - alloying heat treatment] may be performed.

[0086] The conditions for each operation will be described in detail below.

[Heating Steel Slab]

[0087] First, after preparing a steel slab that satisfy all of the above-described alloy component systems, the steel slab may be heated. This process is performed to smoothly perform a subsequent hot rolling process and to sufficiently obtain properties of a desired steel sheet.

[0088] The heating process may be performed in a temperature range of 1050 to 1250°C. When the heating temperature is less than 1050°C, friction between the steel sheet and a rolling mill increases, which may rapidly increase a load applied to a roller during hot rolling. On the other hand, when the temperature exceeds 1250°C, not only does energy costs required for temperature rise increase, but also the amount of surface scale increases, leading to material loss.

[0089] Accordingly, the heating process may be performed in a temperature range of 1050 to 1250°C.

[Hot Rolling]

[0090] A hot-rolled steel sheet may be manufactured by finishing hot rolling the steel slab heated according to the aforementioned process at a Ar3 transformation point or higher, and in this case, a temperature on an outlet side may satisfy Ar3 to Ar3+50°C.

[0091] When the temperature at the outlet side is less than Ar3 during the finishing hot rolling, ferrite and austenite two-phase rolling may be performed, resulting in material non-uniformity. On the other hand, when the temperature thereof exceeds Ar3+50°C, there may be a concern that material non-uniformity may occur due to the formation of abnormally coarsened grains by high-temperature rolling, and as a result, there may be a problem in that coil distortion occurs during subsequent cooling.

[0092] More specifically, the finishing hot rolling may be performed in a temperature range of 800 to 1000°C.

[Coiling]

[0093] The hot-rolled steel sheet produced according to the aforementioned process may be coiled, and in this case, the coiling may be performed in a temperature range of 400 to 700°C.

[0094] When the coiling temperature is less than 400°C, the strength of the hot-rolled steel sheet may be excessively increased, which may result in a rolling load during subsequent cold rolling. Furthermore, it takes excessive costs and time to cool the hot-rolled steel sheet to the coiling temperature, causing an increase in process costs. On the other hand, when the temperature exceeds 700°C, excessive scale may occur on a surface of the hot-rolled steel sheet, which is likely to cause surface defects, and this may cause the deterioration of plating.

[0095] Accordingly, the coiling process may be performed in a temperature range of 400 to 700°C.

[Cooling]

[0096] The coiled hot-rolled steel sheet may be cool to room temperature at average cooling rate of 0.1°C/s or less (excluding 0°C/s). The cooling may be performed at the average cooling rate of more advantageously 0.05°C/s or less, and more advantageously 0.015°C/s or less. Here, the cooling denotes average cooling rate.

[0097] In this way, by cooling the coiled hot-rolled steel sheet at a constant velocity, it may be possible to obtain a hot-rolled steel sheet in which carbides as a nucleation site of austenite are finely dispersed. In other words, fine carbides are evenly dispersed in the steel during a hot rolling process, and during subsequent annealing, the carbides dissolve to allow an austenite phase in the steel to be finely dispersed and formed, and as a result, after the annealing is completed, a uniformly dispersed fine martensite phase may be obtained.

[Cold rolling]

[0098] The hot-rolled steel sheet coiled according to the aforementioned process may be manufactured as a cold-rolled steel sheet by cold rolling, and in this case, the cold rolling may be performed at a cold reduction ratio (total reduction ratio) of 30 to 80%.

[0099] Specifically, the present disclosure may increase energy stored in the steel by controlling a cumulative reduction ratio of an initial stand, preferably stands 1 to 3, by 20% or more, during the cold rolling, thereby having an effect of acting as a driving force for promoting recrystallization of ferrite in a subsequent annealing process. For this reason, the present disclosure may impart an effect of lowering the fraction of microcrystalline ferrite in the steel.

[0100] When the microcrystalline ferrite is present in the steel, deformation and stress are concentrated locally, resulting in poor steel ductility, while recrystallized ferrite contributes to ductility improvement by mitigating deformation and stress concentration.

[0101] When the cumulative reduction ratio of the initial stands 1 to 3 is less than 20% during the cold rolling, or the cold reduction ratio (total reduction ratio) to a final stand is less than 30%, it may be difficult to secure a desired thickness and to correct a shape of the steel sheet. In addition, there may be a problem in that the fraction of microcrystalline ferrite increases to reduce the ductility. On the other hand, when the cold reduction ratio to the final stand exceeds 80% during the cold rolling, there may be a problem in that the strength thereof increases, resulting in a roll load during cold rolling, and the possibility of cracks occurring at an edge portion of the steel sheet may increase.

[0102] In the present disclosure, the cold rolling may be performed using a rolling mill comprised of five or six stands, but the present disclosure is not limited thereto.

[Continuous Annealing]

[0103] The cold-rolled steel sheet produced according to the aforementioned process may be continuously annealed. The continuous annealing treatment may be performed, for example, in a continuous alloying molten plating furnace.

[0104] The continuous annealing operation is a process of forming ferrite and austenite phases simultaneously with recrystallization and decomposing carbons.

[0105] The continuous annealing treatment may be preferably performed in a temperature range of Ac1+30°C to Ac3+30°C, and more preferably, in a temperature range of 800 to 870°C.

[0106] When a temperature during the continuous annealing is below Ac1+30°C, insufficient recrystallization may not be achieved, and also, it may be difficult to form sufficient austenite, and thus, the fraction of a desired level of martensite phase and bainite phase may not be secured after the annealing. On the other hand, when the temperature exceeds Ac3+30°C, because a size of austenite crystal grains is coarsened, fine retained austenite phase cannot be evenly formed around a hard phase. Furthermore, productivity may be reduced, and due to high-temperature annealing, the formation of surface enriched materials is intensified by elements degrading the wettability of hot-dip galvanizing of Si, Mn, B and the like, and thus plating surface quality may not be ensured.

[Staged Cooling]

[0107] As described above, the continuously annealed cold-rolled steel sheet may be cooled step by step.

[0108] Specifically, the cooling may be performed at average cooling rate of 10°C/s or less (excluding 0°C/s) by 450 to 700°C (the cooling at this time is referred to as primary cooling) and then performed at average cooling rate of 3°C/s or more by 250 to 500°C (the cooling at this time is referred to as secondary cooling).

<u>Primary Cooling</u>

[0109] According to the present disclosure, in formation of tempered martensite and bainite phases with a total area fraction of 40-80% as a final structure, an end temperature in a subsequent secondary cooling process may be controlled, thereby controlling a fraction of martensite and bainite generated in this case.

[0110] Specifically, when the subsequent secondary cooling is terminated below a martensite transformation starting temperature (Ms), a relatively large amount of the martensite phase may be formed, and for this purpose, it is advantageous to control the end temperature of the primary cooling to be as low as possible. Furthermore, if the subsequent secondary cooling is terminated in a bainite temperature range, a bainite phase may be relatively advantageously formed, and for this purpose, it is advantageous to control the end temperature of the primary cooling to be higher.

[0111] Preferably, when the subsequent secondary cooling is terminated below Ms, the primary cooling may be performed up to a temperature range of 450 to 600°C, and when the subsequent secondary cooling is terminated in the bainite temperature range, the primary cooling may be performed by a temperature range of 550 to 700°C.

[0112] When the end temperature during the primary cooling is less than 450°C, a larger load may be applied to facilities configured to cool an atmospheric gas in an annealing furnace, and the cooling rate is accelerated, and thus, a ferrite phase generated during cooling may not be sufficiently ensured. On the other hand, when the end temperature exceeds 700°C, an excessively high cooling rate may be required during subsequent cooling (secondary cooling).

[0113] Furthermore, when the average cooling rate exceeds 10°C/s during the primary cooling, carbon diffusion may not be sufficiently generated. Meanwhile, in consideration of productivity, the primary cooling may be performed at the average cooling rate of 1°C/s or more.

Secondary Cooling

[0114] After completing the first cooling under the above-described conditions, the second cooling may be performed, and in this case, the formation of a desired microstructure may be induced by controlling a cooling end temperature and cooling rate.

[0115] When cooling is performed below Ms during the secondary cooling, quenching martensite may be formed, and as the temperature decreases, the fraction of the quenching martensite increases, which may lead to an improvement in the strength of the steel sheet. Furthermore, in a subsequent heat treatment (i.e., a reheating process of the present disclosure), the martensite may be tempered into tempered martensite, and supersaturated carbon in the martensite may be distributed to surrounding untransformed austenite, thereby increasing the stability of retained austenite and improving ductility.

[0116] When the cooling is performed at a temperature exceeding Ms during the secondary cooling, a bainite fraction may be increased. In this case, during a bainite transformation process, the precipitation of carbides may be delayed due to the effects of Si and Al, and as carbons is distributed from the bainite to the surrounding untransformed austenite, the stability of the retained austenite may be increased and ductility may be improved.

[0117] When the end temperature of the secondary cooling is less than 250°C, there may be a problem in that the fraction of the quenching martensite may increase excessively to decrease the fraction of a retained austenite phase, and the shape of the steel sheet may become inferior. On the other hand, when the temperature exceeds 500°C, bainite may be insufficiently formed to decrease the fraction of the retained austenite phase, and in a subsequent process, there may be a problem in that the fraction on fresh martensite may increase significantly to excessively increase the strength thereof. More advantageously, the secondary cooling may be terminated at 400°C or less.

[0118] Furthermore, when the average cooling rate is less than 3°C/s during the secondary cooling, a pearlite phase may be formed, so that there may be a concern that a bainite phase may not be formed at a target level. On the other hand, an upper limit of the average cooling rate is not particularly limited, and may be selected appropriately by a person skilled in the art in consideration of the specifications of the cooling facility. For example, the cooling may be performed at 100°C/s or less.

[0119] Furthermore, the secondary cooling may use a hydrogen cooling facility using hydrogen gas (H_2 gas). In this manner, the cooling may be performed using a hydrogen cooling facility to obtain the effect of suppressing surface oxidation that may occur during the secondary cooling. In this case, the hydrogen cooling facility may be controlled by 60 to 70% hydrogen (H_2) and residual nitrogen (N_2).

[0120] On the other hand, in performing cooling in stages as described above, it may be possible to perform a faster cooling rate during the secondary cooling than the cooling speed during the primary cooling.

Maintenance

[0121] After completing the secondary cooling according to the aforementioned process, a process of maintaining in the cooling temperature range for 30 seconds or more may be further performed.

[0122] Through the maintenance process, it may be possible to obtain an effect of tempering quenching martensite or further increasing the amount of bainite transformation. When the maintenance time is less than 30 seconds, it may be difficult to expect the above-described effect.

[Reheating and Maintaining]

[0123] According to the aforementioned process, a microstructure intended in the present disclosure may be formed through a process of reheating and maintaining a cold-rolled steel sheet that has been cooled step by step. Specifically, it may be desirable to undergo a process of reheating the secondary cooled cold-rolled steel sheet at a temperature of 490°C or less and maintaining for more than 30 seconds.

[0124] By performing reheating and maintaining at the above-described temperature, the quenching martensite produced during the previous cooling process may be transformed into tempered martensite, and may also be accompanied by bainite transformation.

[0125] During the tempering process, carbons supersaturated in martensite is redistributed into surrounding untransformed austenite. Furthermore, when the secondary cooling is terminated in excess of Ms, a bainite fraction may increase significantly during the reheating and maintenance process, and in this process, carbons discharged from the bainite is redistributed into untransformed austenite, and thus, the stability of the retained austenite may be improved to obtain an effect of increasing ductility.

[0126] In addition, during the reheating and maintenance process, potential may be fixed to the tempered martensite and bainite, which may increase the yield strength, and as a result, it may be possible to obtain a steel sheet having a yield ratio of 0.6 to 0.9.

[0127] However, when the temperature is excessively high during the reheating, carbides in the tempered martensite and bainite may be coarsened to decrease the strength thereof, and a carbon redistribution effect to untransformed austenite due to the formation of coarsened carbides may be reduced, resulting in a decrease in a retained austenite fraction, and thus, it is difficult to expect an improvement in ductility.

[0128] Accordingly, the reheating temperature may be limited to 490°C or less, and more advantageously, the reheating temperature may be 350°C or higher.

[0129] As described above, the cold-rolled steel sheet reheated below 490°C may be maintained for more than 30 seconds at that temperature so that the above-described effects are sufficiently realized.

[0130] When the time exceeds 5 minutes during the maintenance, there may be a problem that a tempering effect of martensite is excessive to decrease the strength.

[0131] According to the present disclosure, by optimizing the alloy component system and manufacturing conditions described above, the tempered martensite and bainite may be formed as a matrix structure in a microstructure, and a certain fraction of retained austenite may be finely and uniformly around the tempered martensite and bainite, thereby increasing a yield ratio and ductility compared to the conventional DP steel, and improving the formability for processing components of a steel sheet and the crashworthiness during a vehicle collision. The steel sheet of the present disclosure, which is precisely controlled, may secure ductility while maintaining a higher yield ratio than the conventional DP steel. As a result, it may be possible to provide a high-strength steel sheet having excellent ductility, hole expansion ratio, formability, and crashworthiness.

[0132] Meanwhile, the present disclosure may provide a plated steel sheet by plating the cold-rolled steel sheet produced according to the aforementioned process.

[Hot-Dip Galvanizing]

[0133] According to the aforementioned process, a hot-dip galvanized steel sheet may be manufactured by immersing a steel sheet in a hot-dip galvanized steel sheet bath after reheating and maintaining processes.

[0134] In this case, the hot-dip galvanizing may be performed under normal conditions, but for example, the hot-dip galvanizing may be performed in a temperature range of 430 to 490°C. In addition, a composition of the hot-dip galvanized bath during the hot-dip galvanizing is not particularly limited, and may be a pure zinc plating bath or a zinc-based alloy plating bath containing Si, Al, Mg, and the like.

[Alloying Heat Treatment]

[0135] Furthermore, if necessary, an alloyed hot-dip galvanized steel sheet may be obtained by performing an alloying heat treatment on the hot-dip galvanized steel sheet. In the present disclosure, the alloying heat treatment process condition is not particularly limited, and may be provided as a normal condition. As an example, an alloying heat treatment process may be performed in a temperature range of 480 to 600°C.

[0136] Furthermore, final cooling and coarse rolling process may be performed after the hot-dip galvanizing or alloying heat treatment.

[Final Cooling]

[0137] The hot-dip galvanized or alloyed heat-treated steel sheet may be finally cooled to further introduce fresh martensite. In this case, the final cooling may be performed at cooling rate of 3°C/s or more at room temperature.

[0138] When the cooling rate is less than 3°C/s during the cooling, a fresh martensite phase may not be ensured at an intended level during the cooling process. On the other hand, an upper limit of the cooling rate is not particularly limited, but the cooling may be performed at the cooling rate of 50°C/s or less to form a certain fraction of press martensite.

[Temper Rolling]

[0139] Furthermore, if necessary, by performing the temper rolling on a finally cooled hot-dip galvanized steel sheet or alloyed hot-dip galvanized steel sheet, a large amount of potential in the steel may be formed to further improve bake hardenability. In this case, a reduction ratio thereof may be less than 2% (excluding 0%). When the reduction ratio is 2% or more, it may be advantageous in terms of potential formation, but side effects such as plate breakage may occur due to a limitation in equipment capability.

[0140] Hereinafter, the present disclosure will be described in more detail through embodiments. However, it should be noted that the following embodiments are only for illustrating and explaining the present disclosure in more detail, and are not intended to limit the scope of the present disclosure. This is because the scope of the present disclosure is determined by the matters described in the scope of the patent claim and the matters reasonably inferred therefrom.

Mode for Invention

[0141] After heating a steel slab having an alloy composition illustrated in Table 1 below at a temperature of 1050 to 1250°C, each heated slab was subjected to finished hot rolling at Ar3 to Ar3+50°C to manufacture a hot-rolled steel sheet. Thereafter, each hot-rolled steel sheet was coiled at 400 to 700°C and then cooled to room temperature at cooling rate of 0.1°C/s or less. Thereafter, the cooled hot-rolled steel sheet was subjected to cold rolling at a cold reduction ratio of 45 to 75%, thereby manufacturing a cold-rolled steel sheet.

[0142] The cold rolling was performed in a rolling mill comprised of six stands, and a cumulative reduction ratio of stands 1 to 3 was performed under the conditions illustrated in Table 2 below.

[0143] Then, each cold-rolled steel sheet was continuously annealed under the conditions illustrated in Table 2 below, followed by step-by-step cooling (primary-secondary cooling and maintenance). After the secondary cooling and maintenance process is completed, the steel sheet was reheated at a temperature of 490°C or less, and then, a process of maintaining at that temperature was performed. The maintaining process was performed for 30 seconds after the secondary cooling.

[0144] Then, the steel sheet was galvanized in a hot-dip galvanizing bath at 430 to 490°C and cooled to room temperature at cooling rate of 5°C/s, and was then subjected to temper rolling to less than 2%, thereby manufacturing a hot-dip galvanized steel sheet. In this case, some steels were subjected to the galvanizing treatment, followed by alloying heat treatment.

[Table 1]

Division	Alloy Composition (weight %)												Expression 1
	C	Si	Mn	P	S	Sol.A	Nb	Ti	B	Cr	Mo	Sb	
Inventive Steel1	0.14	0.60	2.41	0.011	0.002	0.035	0.021	0.040	0	0.41	0.20	0	1.81
Inventive Steel2	0.16	0.70	2.39	0.010	0.003	0.035	0.020	0.011	0	0.22	0.10	0.030	2.08
Inventive Steel3	0.16	1.00	2.21	0.008	0.002	0.030	0	0	0.0016	0.51	0	0	2.40
Inventive Steel4	0.15	0.90	2.22	0.008	0.003	0.030	0	0	0.0010	0.71	0	0.034	2.21
Inventive Steel5	0.17	1.00	2.20	0.008	0.002	0.030	0	0	0.0014	0.42	0	0.038	2.48
Inventive Steel6	0.18	1.00	1.99	0.008	0.004	0.030	0	0	0.0015	0.69	0	0.036	2.56
Comparative Steel1	0.06	0.41	2.29	0.012	0.002	0.032	0.031	0.021	0.0027	0.84	0.12	0.018	0.96
Comparative Steel2	0.07	0.40	2.32	0.021	0.004	0.035	0.051	0.020	0.0024	0.85	0.13	0	1.03
Comparative Steel3	0.07	0.60	2.31	0.022	0.002	0.035	0.021	0.021	0.0020	0.85	0.12	0	1.25
Comparative Steel4	0.11	0.60	2.61	0.011	0.001	0.025	0.023	0.015	0	0.50	0.21	0.031	1.56
Comparative Steel5	0.11	0.30	2.63	0.015	0.001	0.125	0.021	0.015	0	0.51	0.21	0.030	1.31

[Table 2]

Division	coiling Temperature (°C)	Cold Rolling		Annealing	Primary Cooling		Secondary Cooling		Reheating	Maintenance
		1 to 3 Stand Reduction Ratio (%)	Total Reduction Ratio (%)	Temperature (°C)	End Temperature (°C)	Cooling Velocity (°C /s)	End Temperature (°C)	Cooling Velocity (°C /s)	Temperature (°C)	Time (sec)
Inventive Steel 1	540	30	60	860	550	5.4	280	12.8	450	55
Inventive Steel 2	620	35	55	840	680	2.9	480	9.5	460	60
Inventive Steel 3	650	25	70	850	600	4.5	450	7.1	480	45
Inventive Steel 4	570	40	50	860	480	6.8	300	8.5	480	40
Inventive Steel 5	520	30	65	870	500	6.6	300	8.5	420	65
Inventive Steel 6	600	25	60	810	550	4.6	300	11.9	460	45
Comparative Steel 1	750	25	45	770	650	3.9	550	4.7	550	45
Comparative Steel 2	580	15	60	830	680	2.7	500	9.5	530	60
Comparative Steel 3	620	30	70	850	750	1.8	300	21.4	300	20
Comparative Steel 4	600	40	55	850	680	3.2	200	22.8	460	25
Comparative Steel 5	300	35	75	810	680	3.1	450	10.9	560	50

[0145] The microstructure of each steel sheet manufactured according to the conditions described above were observed and illustrated in Table 3 below. In this case, in the microstructure of each steel plate, fractions of tempered martensite (TM), bainite (B), ferrite (F), fresh martensite (FM) and retained austenite (A) were measured using FE-SEM, an image analyzer, EBSD, and XRD, after Nital corrosion at 1/4t (t: steel plate thickness (unit mm)) point of a steel sheet. In this case, a share of retained austenite was also measured using the EBSD.

[0146] Furthermore, tensile properties were evaluated in an L direction using the DIN standard for test specimens collected for a tensile test of each steel plate.

[0147] Furthermore, the hole expansion ratio (HER) was measured by pushing up a circular hole punched in a 10mm diameter according to the ISO 16630 procedure with a cone punch until cracks occurred in the specimen, and measuring a ratio of an initial hole diameter to a hole diameter after change, and was calculated using the following expression.

[Expression]

$$\text{Hole Expansion Ratio (HER, \%)} = \{(D - D_0) / D_0\} \times 100$$

(Here, D means a hole diameter (mm) when the crack penetrates through the steel sheet in a thickness direction, and D_0 means an initial hole diameter (mm) .

[Table 3]

Division	Microstructure (Area%)					Mechanical Properties					
	TM+B	F	FM	A	Share (A_{TM+B}/A_T)	YS (MPa)	TS (MPa)	E1 (%)	YR	HER (%)	Expression 2
Inventive Steel 1	55	35	5	5	90	799	1065	14.7	0.75	30	10.4
Inventive Steel 2	48	32	13	7	91	755	1042	15.8	0.72	25	10.9
Inventive Steel 3	57	29	7	7	94	780	1041	15.3	0.75	31	11.0
Inventive Steel 4	60	25	7	8	95	845	1073	16.6	0.79	35	12.2
Inventive Steel 5	64	27	2	7	91	848	1077	15.4	0.79	40	11.3
Inventive Steel 6	58	26	7	9	98	823	1072	17.1	0.77	32	12.3
Comparative Steel 1	31	56	9	4	76	500	891	14.0	0.56	11	8.8
Comparative Steel 2	35	30	32	3	72	752	1081	10.3	0.70	21	6.7
Comparative Steel 3	85	5	8	2	65	986	1076	9.0	0.92	30	7.7
Comparative Steel 4	82	12	3	3	77	947	1084	10.3	0.87	32	8.3
Comparative Steel 5	38	20	40	2	70	802	1211	10.6	0.66	20	5.8
YS: Yield Strength, TS: Tensile Strength, El: Elongation Ratio, YR: Yield Ratio (YS/TS)											

[0148] As illustrated in Tables 1 to 3, in Inventive Examples 1 to 6 satisfying both the alloy component system and the manufacturing conditions proposed in the present disclosure, a tempered martensite phase and a bainite phase were formed in a total of 40 to 80 area% as intended, and retained austenite phases were mainly formed around the tempered martensite phase and the bainite phase. Accordingly, a yield ratio thereof may satisfy 0.6 to 0.9 as well as a high strength of 980 MPa, and an elongation ratio of 10% or more and a hole expansion ratio of 20% or more may be secured.

[0149] In other words, in the steel sheet manufactured by the present disclosure, the strength and ductility may be significantly improved at the same time, and specifically, by satisfying the value of Relational Expression 2, it may be possible to secure the crashworthiness and formability targeted in the present disclosure.

[0150] On the other hand, in Comparative steels 1 to 5 that deviate from Relational Expression 1 proposed in the present disclosure and do not satisfy the manufacturing conditions, at least one physical property was degraded because an intended microstructure was not formed.

[0151] Comparative Steel 1 was unable to secure the target level of strength due to an excessive ferrite phase, and had a poor hole expansion ratio and deviated from Relational Expression 2, which made it impossible to secure crashworthiness and formability.

[0152] Comparative Steels 2 and 5 were excessively formed on a fresh martensite phase to secure high strength, whereas they deviated from Relational Expression 2, which made it impossible to secure crashworthiness and formability.

[0153] Comparative Steel 3 had a small ferrite phase, and the ductility thereof significantly decreased as the retained austenite phase was not formed around a hard phase.

[0154] In Comparative Steel 4, the retained austenite phase was not formed around the hard phase to lead to a relatively low elongation ratio, it was impossible to secure crashworthiness and formability as the Comparative Steel 4 deviated from Relational Expression 2.

[0155] FIG. 1 illustrates a graph of a change in mechanical properties (Relational Expression 2) according to a value of Relational Expression 1.

[0156] As illustrated in FIG. 1, it may be confirmed that when the value of Relational Expression 1 proposed in the present disclosure satisfies 1.7 or more, a value of Relational Expression 2 may be ensured to be 9 or more.

[0157] FIG. 2 illustrates an image of a microstructure of Inventive Steel 4 measured by the EBSD.

[0158] As illustrated in FIG. 2, it may be confirmed that the retained austenite phase is mainly formed around the tempered martensite phase and the bainite phase, and it may be seen that the ferrite phase and the fresh martensite phase are properly formed.

Claims

1. A high-strength steel sheet having excellent crashworthiness and formability, comprising:

   by wt%, carbon (C): 0.06 to 0.2%, silicon (Si): 0.4 to 1.4%, manganese (Mn): 1.8 to 3.0%, aluminum acid (Sol.Al) : 1.0% or less, molybdenum (Mo): 0.4% or less, chromium (Cr): 1.0% or less, antimony (Sb): 0.06% or less, boron (B): 0.01% or less, phosphorus (P): 0.1% or less, sulfur (S): 0.01% or less, and a balance of Fe and unavoidable impurity elements,
   wherein the C, Si and Al satisfy a following Relational Expression 1,
   wherein the steel sheet has a microstructure comprising, 40 to 80% of a sum of an area fractions of tempered martensite and bainite, 3 to 15% of retained austenite, and a balance of ferrite and fresh martensite, and
   in the retained austenite, a share (A_{TM+B}/A_T) of retained austenite (A_{TM+B}) adjacent to tempered martensite and bainite among a total retained austenite fraction (A_T) is 90% or more,

   [Relational Expression 1]

   $$(8\times C) + (1.1\times Si) + (0.8\times Al) \geq 1.7$$

   In the relational expression 1, each element denotes a weight content.

2. The high-strength steel sheet having excellent crashworthiness and formability of claim 1, wherein the steel sheet further includes at least one of titanium (Ti): 0.05% or less and niobium (Nb): 0.05 or less.

3. The high-strength steel sheet having excellent crashworthiness and formability of claim 1, wherein an area fraction of the ferrite is 40% or less.

4. The high-strength steel sheet having excellent crashworthiness and formability of claim 1, wherein an area fraction of the fresh martensite is 20% or less.

5. The high-strength steel sheet having excellent crashworthiness and formability of claim 1, wherein the steel sheet has tensile strength of 980 MPa or more, a yield ratio of 0.6 to 0.9, an elongation ratio of 10% or more, and a hole expansion ratio of 20% or more.

6. The high-strength steel sheet having excellent crashworthiness and formability of claim 1, wherein in the steel sheet, a relationship between a yield ratio, an elongation ratio, and tensile strength satisfies a following Relational Expression 2,

[Relational Expression 2]

$$(YR \times El \times 1000)/TS \geq 9$$

In the relational expression 2, units of each physical property are excluded.

7. A method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability, the method comprising:

heating a steel slab in a temperature range of 1050°C to 1250°C, the steel slab including, by wt%, carbon (C): 0.06 to 0.2%, silicon (Si): 0.4 to 1.4%, manganese (Mn): 1.8 to 3.0%, aluminum acid (Sol.Al): 1.0% or less, molybdenum (Mo): 0.4% or less, chromium (Cr): 1.0% or less, antimony (Sb): 0.06% or less, boron (B): 0.01% or less, phosphorus (P): 0.1% or less, sulfur (S): 0.01% or less, and a balance of Fe and unavoidable impurity elements the C, Si and Al satisfying a following Relational Expression 1;
manufacturing a hot-rolled steel sheet by finish hot rolling the steel slab at a temperature range of a finish hot rolling outlet-side temperature of Ar3°C to Ar3+50°C;
coiling the hot-rolled steel sheet in a temperature range of 400°C to 700°C;
after the coiling, cooling the hot-rolled steel sheet to room temperature at cooling rate of 0.1°C/s;
after the cooling, manufacturing a cold-rolled steel sheet by cold rolling the hot-rolled steel sheet with a total reduction ratio of 30 to 80%;
performing continuous annealing the cold-rolled steel sheet;
primarily cooling the cold-rolled steel sheet at cooling rate of 10°C/s or less to 450 to 700°C;
after the primary cooling, secondarily cooling at cooling rate of 3°C/s or higher to 250 to 500°C; and
reheating the cold-rolled steel sheet to a temperature of 490°C or less and maintaining for 30 seconds or more,
wherein the cold rolling is performed at a cumulative reduction ratio of the first 1 to 3 stands of 20% or more ,

[Relational Expression 1]

$$(8 \times C) + (1.1 \times Si) + (0.8 \times Al) \geq 1.7$$

In the relation expression 1, each element denotes a weight content.

8. The method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability of claim 7, wherein the continuous annealing is performed within a temperature range of Ac1+30°C to Ac3+30°C.

9. The method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability of claim 7, wherein the cooling rate during the secondary cooling is faster than the cooling rate during the primary cooling.

10. The method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability of claim 7, wherein the secondary cooling is performed in a hydrogen rapid cooling facility using hydrogen (H_2) gas.

11. The method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability of claim 7, further comprising:
maintaining for 30 seconds or more after the secondary cooling.

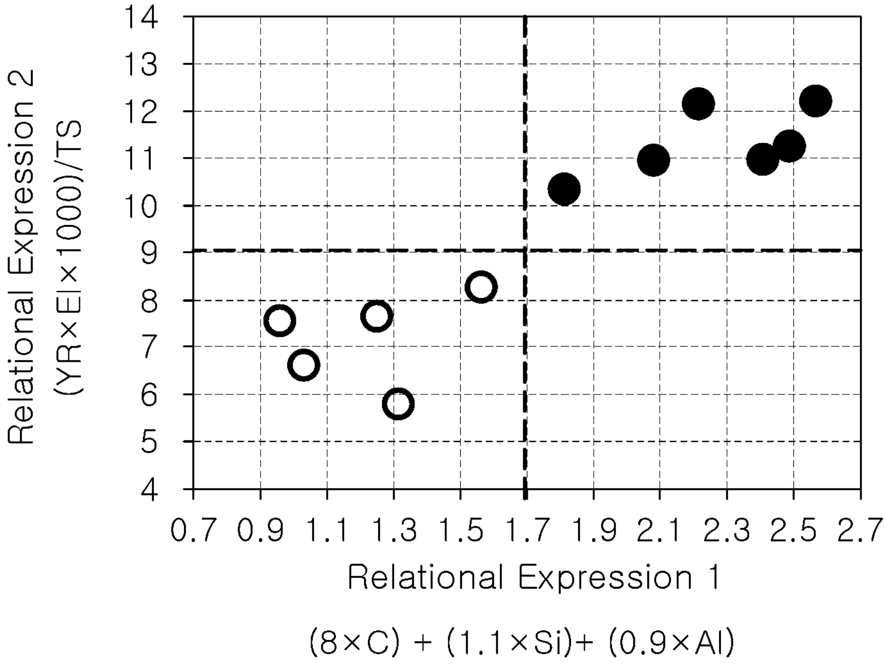
12. The method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability of claim 7, further comprising:
after the reheating and maintaining, hot-dip galvanizing the steel sheet in a bath temperature of 430 to 490°C.

13. The method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability of claim 12, further comprising:
after the hot-dip galvanizing, performing an alloying heat treatment.

14. The method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability of claim 13, further comprising:
after the hot-dip galvanizing or the alloying heat treatment, final cooling the steel sheet to room temperature at average cooling rate of 3°C/s or more.

15. The method for manufacturing a high-strength steel sheet having excellent crashworthiness and formability of claim 14, further comprising:
after the final cooling, temper rolling the steel sheet at a reduction ratio of less than 2%.

[FIG. 1]


Relational Expression 2
(YR×El×1000)/TS
14
13
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11
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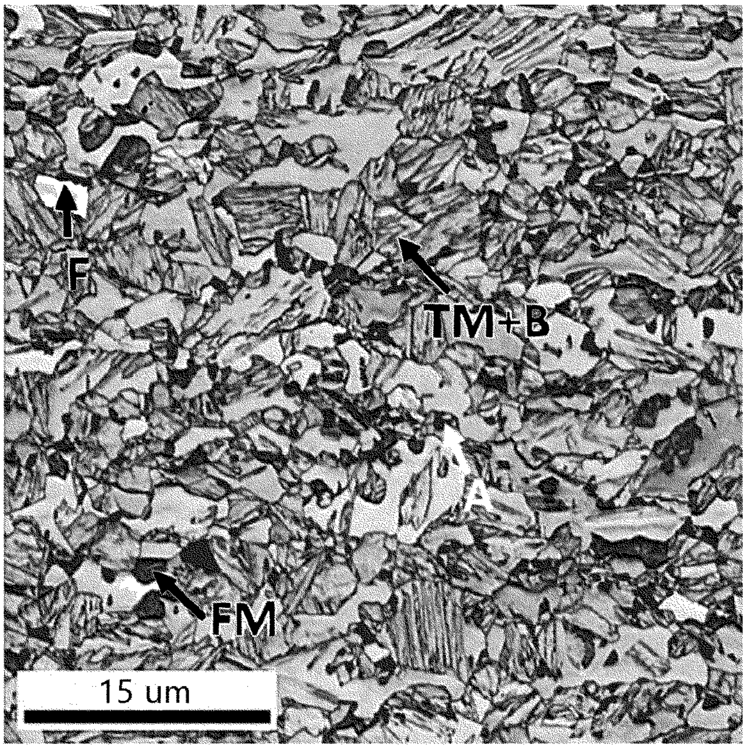
4
0.7 0.9 1.1 1.3 1.5 1.7 1.9 2.1 2.3 2.5 2.7
Relational Expression 1
(8×C) + (1.1×Si)+ (0.9×Al)

[FIG. 2]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INTERNATIONAL SEARCH REPORT

International application No.
PCT/KR2022/017031

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/38(2006.01)i; **C22C 38/22**(2006.01)i; **C22C 38/32**(2006.01)i; **C22C 38/28**(2006.01)i; **C21D 8/02**(2006.01)i; **C21D 9/46**(2006.01)i; **C23C 2/06**(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); C21D 8/02(2006.01); C21D 9/46(2006.01); C22C 38/00(2006.01); C22C 38/14(2006.01); C22C 38/58(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: 템퍼드 마르텐사이트(tempered martensite), 오스테나이트(austenite), 프레시 마르텐사이트(fresh martensite), 베이나이트(bainite), 페라이트(ferrite), 탄소(C), 규소(Si), 망간(Mn), 알루미늄(Al), 소둔(annealing), 냉각(cooling)

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X	WO 2021-200578 A1 (JFE STEEL CORPORATION et al.) 07 October 2021 (2021-10-07) See paragraphs [0065], [0067], [0070], [0074]-[0078], [0082], [0085] and [0092], claims 3-5 and table 3.	1-15
A	JP 2017-002384 A (NIPPON STEEL & SUMITOMO METAL) 05 January 2017 (2017-01-05) See paragraph [0125] and claims 2-4, 6 and 8.	1-15
A	KR 10-2017-0071658 A (POSCO) 26 June 2017 (2017-06-26) See paragraphs [0106]-[0107] and claims 1-2 and 5.	1-15
A	US 2019-0368002 A1 (NIPPON STEEL CORPORATION) 05 December 2019 (2019-12-05) See paragraphs [0162]-[0163] and claims 18, 20 and 27.	1-15
A	JP 2021-123801 A (JFE STEEL CORP.) 30 August 2021 (2021-08-30) See paragraphs [0057]-[0058] and claims 1-2.	1-15

☐ 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	"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
"D" document cited by the applicant in the international application	"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
"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)	"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
"O" document referring to an oral disclosure, use, exhibition or other means	"&" document member of the same patent family
"P" document published prior to the international filing date but later than the priority date claimed	

Date of the actual completion of the international search **08 February 2023**	Date of mailing of the international search report **08 February 2023**
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.

Form PCT/ISA/210 (second sheet) (July 2022)

INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/KR2022/017031

Patent document cited in search report	Publication date (day/month/year)	Patent family member(s)	Publication date (day/month/year)
WO 2021-200578 A1	07 October 2021	JP 2022-020866 A	01 February 2022
		JP 2022-023085 A	07 February 2022
		JP 7001203 B1	03 February 2022
		JP 7044196 B2	30 March 2022
		JP WO2021-200578 A1	07 October 2021
		KR 10-2022-0144405 A	26 October 2022
JP 2017-002384 A	05 January 2017	JP 6524810 B2	05 June 2019
KR 10-2017-0071658 A	26 June 2017	CN 108431273 A	21 August 2018
		CN 108431273 B	12 June 2020
		EP 3392363 A1	24 October 2018
		EP 3392363 B1	25 March 2020
		JP 2019-504196 A	14 February 2019
		JP 6763023 B2	30 September 2020
		KR 10-1758485 B1	17 July 2017
		MX 2018007347 A	24 August 2018
		US 10900097 B2	26 January 2021
		US 2018-0371570 A1	27 December 2018
		WO 2017-105064 A1	22 June 2017
US 2019-0368002 A1	05 December 2019	BR 112019013362 A2	31 December 2019
		CN 110177896 A	27 August 2019
		CN 110177896 B	14 September 2021
		EP 3569727 A1	20 November 2019
		JP 6844627 B2	17 March 2021
		JP WO2018-131722 A1	19 July 2018
		KR 10-2019-0109407 A	25 September 2019
		MX 2019008167 A	06 September 2019
		TW 201833343 A	16 September 2018
		TW I666329 B	21 July 2019
JP 2021-123801 A	30 August 2021	None	

REFERENCES CITED IN THE DESCRIPTION

Patent documents cited in the description

- JP 4289120 A **[0018]**
- JP 2005264176 A **[0018]**
- KR 20150073844 **[0018]**
- JP 2010090432 A **[0018]**