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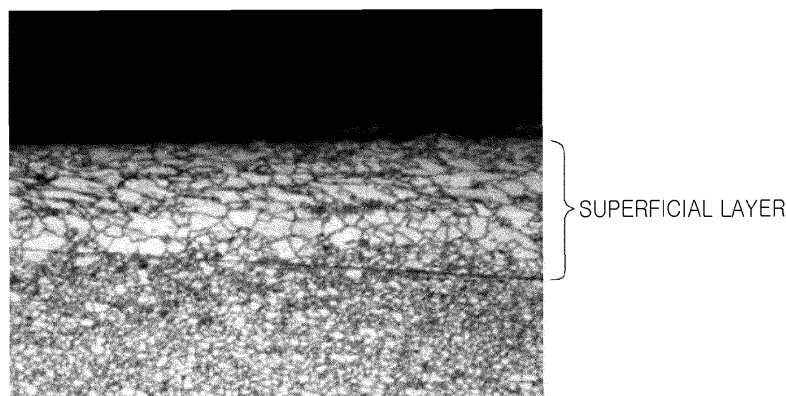
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(54) **HIGH STRENGTH AND HIGH FORMABILITY STEEL SHEET HAVING EXCELLENT SPOT WELDABILITY, AND METHOD FOR MANUFACTURING SAME**

(57) The present invention relates to a steel sheet for use in automobiles, etc., and to a steel sheet that has high strength and high formability and is superb in terms of spot weldability, and a manufacturing method therefor.

FIG. 2



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Description

Technical Field

5 **[0001]** The present disclosure relates to a steel sheet for use in automobiles, etc., and to a steel sheet not only having high strength and high formability, but also having excellent spot weldability, and a method for manufacturing the same.

Background Art

10 **[0002]** Improving fuel efficiency and durability is an important issue that automobile companies should address. To this end, by using thin and high-strength steel, it is possible to simultaneously improve various issues, environmental, fuel efficiency, crash resistance, and durability. For example, the Insurance Institute for Highway Safety in the United States has gradually strengthened crash safety regulations to protect occupants, and has been requiring harsh crash components such as 25% small overlap since 2013. Such a solution is in reducing the weight of automobiles, and in order to reduce the weight thereof, high strength of a steel material is required, and high formability thereof is also required.

15 **[0003]** However, as the strength of steel increases, it has advantageous characteristics in absorbing impact energy, but generally, as the strength increases, elongation decreases so there is a problem that molding processability deteriorates. In addition, yield strength is excessively high, an inflow of a material from a mold is reduced during molding, resulting in poor formability. Accordingly, the automobile industry is requesting that the steel industry develop a steel material with excellent strength and formability, that is, excellent balance between strength and elongation (TS*EI).

20 **[0004]** Steel companies are developing various products to meet such demands. For example, dual phase steel (DP steel), transformation induced plasticity steel (TRIP steel), complex phase steel (CP steel), ferrite-bainite steel (FB steel), and the like are developed, and products are manufactured through iron making, steel making, casting, hot rolling, cold rolling, and annealing processes.

25 **[0005]** In order to secure such strength and formability, a steel material often has an increased amount of alloying elements added thereto, and in this case, defects may occur during spot welding while forming automobile parts. Spot welding is the most common process for bonding automobile parts and is the most widely used method therefor, due to low cost and excellent productivity.

30 **[0006]** In order to ensure corrosion resistance of the steel material, plating is often performed. In particular, galvanized steel sheets have excellent corrosion resistance and formability, but liquid metal embrittlement (LME) sometimes occurs. Liquid metal embrittlement is a phenomenon in which brittleness occurs when a ductile material is in contact with liquid metal, in particular, a phenomenon in which liquid metal rapidly penetrates along grain boundaries of a base material, causing embrittlement, when there is a tensile stress.

35 **[0007]** Welding, such as spot welding, is commonly used as a method of joining automobile parts. However, during welding, a temperature of a heat-affected zone of the material may rise, melting of a plating layer may occur, and tensile stress may occur due to electrode pressure, so cracks due to liquid metal embrittlement may occur. In particular, when a galvanized steel sheet is spot welded, a melting point thereof is low, so molten zinc causes the problem of liquid metal embrittlement cracking in a weld zone.

40 **[0008]** Therefore, there is a need for a technology that can solve defect problems caused by liquid metal embrittlement (LME) by securing excellent strength and formability while improving spot weldability.

Summary of Invention

Technical Problem

45 **[0009]** An aspect of the present disclosure is to provide a steel sheet securing excellent strength and formability, and simultaneously having excellent spot weldability and a method for manufacturing the same.

50 **[0010]** An object of the present disclosure is not limited to the above description. The object of the present disclosure will be understood from the entire content of the present specification, and a person skilled in the art to which the present disclosure pertains will understand an additional object of the present disclosure without difficulty.

Solution to Problem

55 **[0011]** According to an aspect of the present disclosure, provided is a high-strength and high-formability steel sheet having excellent spot weldability, the steel sheet including by weight: 0.05 to 0.10% of carbon (C), 0.3% or less (excluding 0%) of silicon (Si), 2.0 to 2.5% of manganese (Mn), 0.05% or less (excluding 0%) of titanium (Ti), 0.1% or less (excluding 0%) of niobium (Nb), 1.5% or less (excluding 0%) of chromium (Cr), 0.1% or less of phosphorus (P), and 0.01% or less of sulfur (S), with a balance of Fe and inevitable impurities,

[0012] wherein a microstructure of the steel sheet at a point of 1/4 of a thickness (t) of the steel sheet includes, by volume fraction, 65 to 85% of a soft phase and a remainder of a hard phase,

[0013] wherein a surface layer portion of the steel sheet includes a sound layer, and a thickness of the sound layer is 5 to 50 μm .

[0014] According to another aspect of the present disclosure, provided is a method for manufacturing a high-strength and high-formability steel sheet having excellent spot weldability, the method including: heating a steel slab comprising, by weight: 0.05 to 0.10% of carbon (C), 0.3% or less (excluding 0%) of silicon (Si), 2.0 to 2.5% of manganese (Mn), 0.05% or less (excluding 0%) of titanium (Ti), 0.1% or less (excluding 0%) of niobium (Nb), 1.5% or less (excluding 0%) of chromium (Cr), 0.1% or less of phosphorus (P), and 0.01% or less of sulfur (S), with a balance of Fe and inevitable impurities, to a temperature within a range of 1100 to 1300°C;

hot rolling the heated steel slab, and cooling and hot rolling the steel slab so that a surface temperature of a material is Ar3 or lower for a predetermined period of time during finish hot rolling;

coiling and cooling the hot-rolled steel sheet after the hot rolling;

cold rolling the cooled hot-rolled steel sheet at a reduction ratio of 70 to 90%;

heating and maintaining the cold-rolled steel sheet to a temperature range of Ac1 to Ac1+30°C after the cold rolling; and

slowly cooling the cold-rolled steel sheet at an average cooling rate of 1 to 10°C/s to a temperature within a range of 650 to 700°C, and then rapidly cooling the cold-rolled steel sheet at an average cooling rate of 5 to 50°C/s to a temperature within a range of 300 to 580°C.

Advantageous Effects of Invention

[0015] As set forth above, according to the present disclosure, since a steel sheet having excellent strength and formability, particularly, excellent balance between strength and ductility (TS*EI), may be provided, processing defects in the steel sheet, such as cracks or wrinkles occurring during press forming, may be prevented, so the steel sheet may be appropriately applied to structural members requiring processing into complex shapes. In addition, product quality may be improved by improving spot weldability and reducing the occurrence of liquid metal embrittlement (LME).

[0016] The various and beneficial advantages and effects of the present disclosure are not limited to the above-described content, and may be more easily understood through description of specific embodiments of the present disclosure.

Brief Description of the Drawings

[0017]

FIG. 1 is a diagram illustrating cracks occurring during spot welding of an alloyed hot-dip galvanized steel sheet.

FIG. 2 is an image obtained by observing a surface layer portion of a steel sheet in Inventive Example 1, among examples of the present disclosure.

FIG. 3 is an image obtained by observing a surface layer portion of a steel sheet in Comparative Example 1, among examples of the present disclosure.

FIG. 4 is a schematic diagram illustrating an example of a method for measuring an aspect ratio of a hard phase.

FIG. 5 is a graph illustrating a temperature change of the material when water cooling is applied and when water cooling is not applied during finish hot rolling in the present disclosure.

FIG. 6 is a graph illustrating a heat treatment step in a continuous annealing process.

FIG. 7 is an image obtained by observing cracks caused by LME after spot welding in Comparative Example 1, among examples of the present disclosure, is completed.

Best Mode for Invention

[0018] The terms used in this specification are used to describe the present disclosure and are not intended to limit the present disclosure. In addition, as used herein, singular forms include plural forms unless the relevant definition clearly indicates the contrary.

[0019] The meaning of "including" or "comprising" used in the specification specifies a configuration and does not exclude the presence or addition of another configuration.

[0020] Unless otherwise defined, all terms, including technical terms and scientific terms used in this specification, have the same meaning as that which could be commonly understood by those skilled in the art in the technical field to which the present disclosure pertains. Terms defined in the dictionary are interpreted as having meanings consistent

with related technical literature and current disclosure.

[0021] In order to secure strength and formability, high-strength steel used in automobile materials may typically be dual phase steel (DP steel), transformation induced plasticity steel (TRIP steel), complex phase steel (CP steel), ferrite-bainite steel (FB steel), or the like. The high-strength steel generally has high alloys, and the high-alloy galvanized high-strength steel may cause cracks due to liquid metal embrittlement (LME) during spot welding.

[0022] As a cause of the LME, there is a load exceeding a critical level, molten metal, appearance of austenite, or the like. During spot welding, as current is applied, a temperature of steel rises due to resistance heat, and zinc, which has a low melting point, begins to melt first. Thereafter, the steel is transformed into austenite. The lower an austenite formation temperature, the longer a contact time between molten zinc and an austenite structure of the steel in a surface layer portion thereof. In this case, when thermal stress or external stress is applied, an austenite grain boundary in a portion in which the stress is concentrated glides and is transformed. In this case, when interface energy between the steel and the molten zinc is lower than austenite grain boundary energy, the molten zinc penetrates into the austenite grain boundaries and grain boundary cracks occur, ultimately resulting in deterioration in spot weldability. FIG. 1 illustrates cracks which appear when commercially available 1200 MPa grade alloyed galvanized high-strength steel is welded.

[0023] In order to eliminate weld cracks due to the LME, methods to suppress austenite, limit molten metal, and limit external stress are being discussed.

[0024] However, as a welding temperature rises to a melting point, it is not easy to prevent the appearance of austenite in steel. Rather, the more alloying elements such as C and Mn, the lower an A3 temperature. Therefore, in high-strength steel, austenite can appear at a lower temperature, so cracks due to LME may occur more easily.

[0025] Meanwhile, in the case of zinc plating, it is possible to consider a method to increase the melting temperature of zinc, or reduce a plating thickness to reduce an amount of molten zinc, but a decrease in corrosion resistance and steel sheet processability, an increase in plating costs, and the like should be considered.

[0026] Reducing external stress may be difficult, because bonding strength and quality of a weld zone are related.

[0027] According to the above methods, microcracks caused by LME are not sufficiently suppressed. Accordingly, as a result of studying the cracks caused by the LME, fortunately the present inventors have discovered that LME during spot welding is a phenomenon that occurs in a surface layer portion of the steel sheet, and have recognized that spot weldability may be improved by reducing LME by changing a material of the surface layer portion of the steel sheet, and excellent strength and formability may be secured, thereby completing the present disclosure.

[0028] Hereinafter, the present disclosure will be described in detail. First, an embodiment of the steel sheet of the present disclosure will be described in detail.

[0029] An alloy composition of the steel sheet in the present disclosure includes by weight: 0.05 to 0.10% of carbon (C), 0.3% or less (excluding 0%) of silicon (Si), 2.0 to 2.5% of manganese (Mn), 0.05% or less (excluding 0%) of titanium (Ti), 0.1% or less (excluding 0%) of niobium (Nb), 1.5% or less (excluding 0%) of chromium (Cr), 0.1% or less of phosphorus (P), and 0.01% or less of sulfur (S), with a balance of Fe and inevitable impurities. The alloy composition is described in detail as follows. Unless otherwise specified in the present disclosure, a content of each element refers to % by weight.

Carbon (C): 0.05 to 0.10%

[0030] Carbon (C) is an important element added for solid solution strengthening, and C combines with precipitated elements to form fine precipitates, thereby contributing to improving strength of steel. When the C content exceeds 0.10%, hardenability may increase and strength may increase excessively as martensite is formed during cooling when manufacturing steel, but which may cause a decrease in elongation. In addition, weldability may be poor, so there may be a risk of welding defects occurring when processing into parts. When the C content is less than 0.05%, it may be difficult to secure a target level of strength. More advantageously, the C content is preferably in a range of 0.06 to 0.08%.

Silicon (Si): 0.3% or less (excluding 0%)

[0031] Silicon (Si) is a ferrite stabilizing element, and it is advantageous to secure a target level of ferrite fraction by promoting ferrite transformation. In addition, Si is an element, which is effective in increasing strength of ferrite due to excellent solid strengthening ability, so is useful for securing the strength while not decreasing ductility of steel. When the Si content exceeds 0.3%, the solid solution strengthening effect becomes excessive, so that ductility may rather be decreased, and surface scale defects are caused to adversely affect plating surface quality, and phosphatability may be deteriorated. More advantageously, the Si content is preferably 0.1% or less.

Manganese (Mn): 2.0 to 2.5%

[0032] Manganese (Mn) is an element which precipitates sulfur (S) in steel as MnS to prevent hot brittleness by

production of FeS, and is favorable to solid solution strengthening of steel. When the Mn content is less than 2.0%, the effects described above may not be obtained, and it is difficult to secure a target level of strength. On the other hand, when the Mn content exceeds 2.5%, problems in weldability, hot rolling, and the like are likely to occur, and at the same time, there is a risk that ductility may decrease as martensite is more easily formed by an increase in hardenability. In addition, a Mn-band (Mn oxide band) may be excessively formed in the structure to increase the risk of defects such as processing cracks. Further, a Mn oxide is eluted on the surface during annealing to greatly deteriorate plating properties. More advantageously, the Mn content is preferably in a range of 2.2 to 2.4%.

Titanium (Ti): 0.05% or less (excluding 0%)

[0033] Titanium (Ti) is an element forming a microcarbide, and contributes to securing yield strength and tensile strength. In addition, Ti precipitates N in steel as TiN to suppress the formation of AlN by Al which is unavoidably present in steel, and thus, reduces the possibility of cracks during continuous casting. When the Ti content exceeds 0.05%, a coarse carbide is precipitated, and strength and an elongation may be decreased by a reduced carbon amount in steel. In addition, nozzle clogging may occur during continuous casting, and manufacturing costs may be increased. Therefore, the Ti content is preferably 0.05% or less, and preferably exceeds 0%.

Niobium (Nb): 0.1% or less (excluding 0)

[0034] Niobium (Nb) is an element which is segregated at an austenite grain boundary, and suppresses coarsening of austenite crystal grains during an annealing heat treatment, and forms a fine carbide to contribute to strength improvement. When the Nb content exceeds 0.1%, a coarse carbide is precipitated, strength and an elongation may be decreased by a reduced carbon amount in steel, and manufacturing costs may be increased. Therefore, the Nb content is preferably 0.1% or less, and preferably exceeds 0%.

Chromium (Cr): 1.5% or less (excluding 0%)

[0035] Chromium (Cr) is an element which facilitates formation of bainite, suppresses formation of martensite during an annealing heat treatment, and forms a fine carbide to contribute to strength improvement. When the Cr content exceeds 1.5%, bainite may be formed excessively and elongation may decrease. When carbides are formed at grain boundaries, strength and elongation may be inferior, and manufacturing costs may increase. Therefore, it is preferable that Cr is included in an amount of 1.5% or less, and it is more preferable that Cr is included in an amount exceeding 0%.

Phosphorus (P): 0.1% or less

[0036] Phosphorus (P) is a substitutional element having the greatest solid solution strengthening effect, and is an element which improves in-plane anisotropy and is advantageous for securing strength without significantly reducing formability. However, when P is excessively added, a possibility of brittle fracture occurrence is greatly increased, so that a possibility of sheet fracture of a slab during hot rolling is increased and plating surface properties are deteriorated. Therefore, the P content is preferably 0.1% or less, and 0% may be excluded considering an avoidably added level.

Sulfur (S): 0.01% or less

[0037] Sulfur (S) is an element which is unavoidably added as an impurity element in steel, and deteriorates ductility, and thus, it is preferable to control a content of S as low as possible. In particular, S has a problem of increasing a possibility of red brittleness occurrence, it is preferable to control the S content to be 0.01% or less. However, 0% may be excluded considering an avoidably added level.

[0038] The remaining component of the present disclosure is iron (Fe). However, since in the common manufacturing process, unintended impurities may be inevitably incorporated from raw materials or the surrounding environment, 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.

[0039] Typically, a microstructure of the steel sheet is observed at a point of 1/4t in a thickness direction (at a point of 1/4 of the thickness (t) of the steel sheet), which is used to explain physical properties such as strength, formability, and the like. based thereon. In the present disclosure, LME characteristics caused by the microstructure on the surface of the steel sheet, an internal microstructure which determines physical properties such as strength, or the like, are explained, and the microstructure in the surface layer portion and the internal microstructure are explained separately. Here, the internal microstructure refers to a microstructure at 1/4t, and unless otherwise specified, the microstructure refers to the internal microstructure. In a microstructure of steel, a phase immediately after rolling is determined depending on whether

a rolling temperature is an austenite phase or a ferrite phase, and then transformation occurs depending on cooling conditions, and a final microstructure is formed. A hot rolling process is a stage in which dynamic recrystallization occurs during rolling. When the rolling temperature is high, in the microstructure immediately after rolling, a recrystallized austenite single structure may be obtained, when the rolling temperature is low, a mixed phase of recrystallized austenite/ferrite may be obtained, and when the rolling temperature is significantly low, a ferrite single phase may be obtained.

[0040] Therefore, when the rolling temperature is different in the thickness direction of the steel sheet, different microstructures may be obtained in the thickness direction, and when a special treatment such as spraying water are performed during rolling, only the temperature in the surface layer portion may be controlled, and the microstructure in the surface layer portion may be differently set. Accordingly, an aspect of the present disclosure relates to a technology for securing strength and formability by controlling the internal microstructure, and improving LME by controlling the microstructure in the surface layer portion.

[0041] Hereinafter, the steel sheet includes a sound layer in the surface layer portion. It is effective that the sound layer is a ferrite main structure comprised of, by area fraction, 95% or more of ferrite, and the ferrite grain has a size of 6 to 20 μm .

[0042] Meanwhile, it is effective for a thickness of the sound layer to be 5 to 50 μm . When the thickness of the sound layer in the surface layer portion is less than 5 μm , it is difficult to improve LME, and when the thickness of the sound layer in the surface layer portion exceeds 50 μm , it is difficult to sufficiently achieve physical properties such as strength of the steel sheet, or the like. FIGS. 2 and 3 illustrate images obtained by observing surface layer portions of Inventive Example 1 and Comparative Example 1, among examples described later, respectively, and in FIG. 2, a sound layer having coarse grains was confirmed in the surface layer portion, but which may be observed in FIG. 3.

[0043] The microstructure of the steel sheet (at a point of 1/4 of the thickness (t) of the steel sheet) ultimately includes a hard phase and a soft phase, and in particular, by maximizing ferrite recrystallization through an optimized annealing process, the microstructure thereof preferably includes a structure in which bainite and a martensite phase, which is a hard phase, are uniformly distributed in a recrystallized ferrite matrix. In the microstructure, the hard phase is a phase mainly formed of martensite, and includes a small amount of bainite and mixed, and the soft phase is a ferrite phase. In a structure composed of a soft phase and a hard phase, among deformation properties, the soft phase determines the formability, and the hard phase determines the strength.

[0044] The hard phase is preferably included, by area fraction, in an amount of 15 to 35%. When the fraction of the hard phase is too high, strength may be high but elongation may be low, and when the fraction of the soft phase is high, elongation may be high but strength may be low. In order to secure the strength of 780 MPa or more provided by the present disclosure, the hard phase is preferably included, by area fraction, in an amount of 15% or more, and to secure formability, it is preferable that the hard phase is included, by area fraction, in an amount of not exceeding 35%.

[0045] In order to secure appropriate strength and formability at the same time, the soft phase is preferably included, by area fraction, in an amount of 65 to 85%. The soft phase ferrite can be divided into recrystallized ferrite and non-recrystallized ferrite. As shown in FIG. 4, a difference between recrystallized ferrite and non-recrystallized ferrite may be distinguished by an aspect ratio of a grain size thereof to a rolling direction. As shown in FIG. 4(b), non-recrystallized ferrite has a large aspect ratio, and when analyzed in detail, a linear deformation structure is observed within the ferrite grains. On the other hand, since recrystallized ferrite is advantageous in securing formability, it is preferable that recrystallized ferrite in the soft phase is included in an amount of 60% or more. Since non-recrystallized ferrite is a soft phase but the fraction is high, formability is reduced, so it is preferable that non-recrystallized ferrite in the soft phase is included in an amount of 5% or less.

[0046] Meanwhile, an aspect ratio of the hard phase is preferably 1.2 or less. As shown in FIGS. 4(a) and 4(b), the aspect ratio means a ratio (b/a) of a major axis (b) and a minor axis (a) to the grain size with respect to the rolling direction, and the aspect ratio of the hard phase is an aspect ratio of a structure formed by stretching the hard phase in the rolling direction. As the aspect ratio of the hard phase increases, bending properties, which are important for resistance to deformation in the thickness direction, are adversely affected. In addition, as the aspect ratio of the hard phase increases, hole expandability decreases. Therefore, it is important to keep the aspect ratio of the hard phase as low as possible, so it is preferable that the aspect ratio of the hard phase does not exceed 1.2.

[0047] The steel sheet of the present disclosure may have a high tensile strength (TS) of 780 MPa or more and an elongation of 18% or more, so that excellent strength and formability may be secured.

[0048] Meanwhile, the steel sheet of the present disclosure may further include a plating layer to improve corrosion resistance, for example, a zinc-based plating layer. Most steel sheet for automobiles may have a plating layer formed by hot dip plating and an electroplating layer formed on a base steel sheet, and the present disclosure relates to a technology that can include both the plating formed by hot-dip plating and the plating layer formed by electroplating. A thickness of the plating layer may vary depending on need, but for example, the thickness may be 10 μm or less.

[0049] Next, according to an aspect, a method for manufacturing a steel sheet of the present disclosure will be described in detail. The steel sheet may be manufactured by performing processes of preparing a steel slab first and heating the same, performing hot rolling, coiling and cooling, and annealing. Meanwhile, if necessary, a process of forming a plating

layer may be further included. In particular, in the present disclosure, to form a sound layer in a sound layer portion, hot rolling is controlled, and an appropriate structure is formed in an annealing process after deformation in cold rolling. Hereinafter, each step will be described in detail.

5 Steel slab heating

[0050] A steel slab having the alloy-described alloy composition, that is, by weight: 0.05 to 0.10% of carbon (C), 0.3% or less (excluding 0%) of silicon (Si), 2.0 to 2.5% of manganese (Mn), 0.05% or less (excluding 0%) of titanium (Ti), 0.1% or less (excluding 0%) of niobium (Nb), 1.5% or less (excluding 0%) of chromium (Cr), 0.1% or less of phosphorus (P), and 0.01% or less of sulfur (S), with a balance of Fe and inevitable impurities, is prepared and then heated. The present process is performed to smoothly perform a subsequent hot rolling process, and is performed to sufficiently obtain physical properties of the target steel sheet. In the present disclosure, conditions of the heating process are not particularly limited, and any method or condition commonly used in the technical field to which the present disclosure pertains may be used. As an example, the heating process may be performed at a temperature within a range of 1100 to 1300°C.

Hot rolling

[0051] The heated steel slab is hot rolled to manufacture a hot-rolled steel sheet. In the present disclosure, as a method to obtain an appropriate surface layer portion, a method of differentiating a temperature of a surface and a temperature at a point of 1/4 of a thickness of the surface of the steel slab during hot rolling is proposed.

[0052] To this end, finish hot rolling is performed so that the temperature of a central portion (at a point of 1/4 of the thickness thereof), that is, the temperature of a material itself, is Ar3 to 1000°C, and finish hot rolling is preferably performed so that a surface temperature of the material is Ar3 or lower for a predetermined time during finish hot rolling. When an outlet temperature of the material itself is lower than Ar3 in the finish hot rolling, the strength of the material may increase and hot deformation resistance during rolling may be rapidly increased. When the temperature is higher than 1000°C, a rolling load may be relatively decreased, so that it is favorable to productivity, but a thick oxide scale may occur and defects in the surface layer portion may be formed. More preferably, hot rolling may be performed at a temperature within a range of 760 to 940°C.

[0053] Meanwhile, the surface temperature of the material is an important factor for forming a sound layer in the surface layer portion, and when the surface layer temperature is set to Ar3 or lower, so that ferrite may be easily formed by recrystallization during hot rolling, and excessive rolling load is prevented from occurring at the point of 1/4 of the thickness thereof. That is, rolling is performed simultaneously so that the surface temperature is Ar3 or lower for a predetermined period of time, and ferrite may be recrystallized in the surface layer portion to form a sound layer of coarse ferrite. The method for lowering the surface temperature to Ar3 or lower is not particularly limited, but for example, a method of spraying water during or between rolling passes to set the surface temperature to Ar3 or lower for a predetermined period of time may be applied.

[0054] FIG. 5 illustrates a time-temperature graph in hot rolling when cooling is applied to set the surface temperature to Ar3 or lower, as in the present disclosure and when water cooling is not applied. In FIG. 5, when water cooling is not applied, both the central portion and the surface thereof are rolled at a temperature of Ar3 or higher, but when water cooling is applied, it can be confirmed that the surface temperature falls below Ar3 for a certain period of time.

Coiling and cooling

[0055] The hot-rolled steel sheet manufactured by hot rolling can be wound into a coil shape. The coiling may be performed at a temperature within a range of 400 to 700°C. When the coiling temperature is lower than 400°C, an increase in excessive strength of the hot-rolled steel sheet may be caused due to excessive formation of martensite or bainite, and problems such as shape defects due to a load during subsequent cold rolling may be caused. However, when the winding temperature is higher than 700°C, a surface scale may be increased to deteriorate pickling properties.

[0056] Meanwhile, it is preferable that the coiled hot-rolled steel sheet is cooled to room temperature at an average cooling rate of 0.1°C/s or less (excluding 0). The coiled hot-rolled steel sheet may be cooled after performing processes such as transport, stacking, and the like, and the process before cooling is not limited thereto. The coiled hot-rolled steel sheet may be cooled at a constant rate, so that a hot-rolled steel sheet in which carbides, serving as austenite nucleation sites are finely dispersed may be obtained.

[0057] Thereafter, an additional process of removing surface scales by pickling the surface of the hot-rolled steel sheet before performing a subsequent cold rolling process may be performed. The pickling method is not particularly limited, and it is sufficient to use a method commonly used in the technical field to which the present disclosure pertains.

Cold rolling

[0058] The hot-rolled steel sheet as above may be cold rolled at a constant reduction ratio at room temperature to be manufactured into a cold-rolled steel sheet.

[0059] During the cold rolling, cold rolling is preferably performed at a reduction ratio of 70 to 90%. When the reduction ratio of the cold rolling is less than 70%, driving force for recrystallization is reduced, so that ferrite is coarsely formed, and austenite formation is also reduced, so that the temperature in the soaking section in the annealing furnace should be increased to ensure a sufficient fraction of austenite. On the other hand, when the cold rolling reduction ratio exceeds 90%, there is a high possibility of cracks occurring at an edge of the steel sheet, an initial thickness before rolling should be excessively thick, and the number of rolling passes increases, leading to lower productivity.

[0060] A method for performing the cold rolling is not particularly limited in the present disclosure, and any method used in the technical field to which the present disclosure pertains may be applied. For example, the method for performing the cold rolling may include a tandem cold rolling mill (TCM) method, a Sendzimir rolling mill (ZRM) method, and the like. To briefly explain the methods, TCM is a reversible rolling method, which has an advantage of low manufacturing cost and excellent productivity as mass production may be performed, but has a disadvantage of being somewhat limited in applying pressing force. ZRM is a reversible batch-type method, which has a disadvantage of low productivity, and has an advantage of being somewhat easier to apply pressing force.

[0061] Since the reduction ratio in the cold rolling is an important operating factor that improves various physical properties by improving the phase transformation of steel, controlling the reduction ratio is especially important to ensure quality. In the present disclosure, it is preferable to adopt an appropriate method considering product material, size, operating environment, and the like.

Continuous annealing

[0062] It is preferable that the cold-rolled steel sheet manufactured as described above is continuously annealed. The continuous annealing treatment may be performed in a continuous annealing furnace (CAL), as an example. An example of a heat treatment step of the continuous annealing process was shown as a graph in FIG. 6. As shown in FIG. 6, the heat treatment step of the continuous annealing process may be comprised of a heating section (HS), a soaking section (SS), a slow cooling section (SCS), a rapid cooling section (RCS), and an over aging section (OAS). Generally, a temperature of each section is measured by measuring a temperature attached to an end point of each section, so the temperature refers to the temperature at the end point of each section. For example, the temperature in the rapid cooling section (RCS) is a temperature of the section at which the rapid cooling section ends, which is indicated as 4 in FIG. 6.

[0063] In the heating section (HS), the steel sheet is heated at a constant temperature increase rate, and as the temperature of the steel sheet increases, recovery of dislocations, precipitation of cementite, recrystallization of ferrite, and reverse transformation of dual phase region occur. A sheet-passing speed may vary depending on the thickness and width of the steel sheet, and a change in microstructure for each of the temperature sections may vary depending on an initial structure during hot rolling and cold rolling reduction ratio.

[0064] When entering the soaking section (SS), it is maintained at a constant temperature for a certain period of time, and in this case, depending on an annealing temperature, dual phase region austenite or single phase region austenite reverse transformation is observed. The soaking section (SS) is known to be one of the sections which consumes the most energy in an annealing furnace. In the slow cooling section (SCS), cooling is usually performed at a low cooling rate, and after the SCS section, continuous cooling is performed at a high cooling rate in the rapid cooling section (RCS), and bainite may be generated during cooling depending on an RCS setting temperature and a degree of hardenability.

[0065] The temperature of the soaking section (SS) is closely related to phase transformation. Factors affecting phase transformation, and a change in a state of a substance including temperature, pressure, composition, and the like, and when the composition is determined, the composition may be adjusted through temperature and pressure. In particular, the higher the temperature and pressure, the faster the phase transformation during heating in the annealing furnace may proceed. However, as the temperature increases, consumed energy costs increase, and carbon emissions such as carbon dioxide increase after combustion, making it unfriendly. In the steel manufacturing process, a variable, compared to pressure is a cold rolling reduction ratio, and when the cold rolling reduction ratio is increased at the same temperature, the phase transformation may progress rapidly. However, when the cold rolling reduction ratio is increased, the phase transformation can be created even at a low temperature. Using this principle, in the present disclosure, the cold rolling reduction ratio is performed at 70 to 90%, which is higher than the existing method.

[0066] The temperature in the soaking section in a common annealing process is generally in a range of $Ac_1+30^{\circ}C$ to $Ac_3-30^{\circ}C$. However, as described above, in the present disclosure, ferrite may be recrystallized and austenite may be formed even when a heat treatment is performed at a low temperature by increasing the cold rolling reduction ratio, so the annealing process of the present disclosure is preferably performed so that the cold rolled steel sheet is heated and maintained to a temperature range of $Ac_1+30^{\circ}C$ to $Ac_3-30^{\circ}C$. In the present disclosure, hardness may be reduced

and processability may be improved through recrystallization and phase transformation even in the above-described temperature range.

[0067] The cold-rolled steel sheet heat treated in the temperature range may be cooled, thereby forming a target structure, and in this case, it is preferable to perform cooling in a stepwise manner. In the present disclosure, the stepwise cooling may be formed in a slow cooling zone (SCS) and a rapid cooling zone (RCS). For example, it is preferable to perform slow cooling at a temperature within a range of 650 to 700°C at an average cooling rate of 1 to 10°C/s, and then to perform rapid cooling at a temperature within a range of 300 to 580°C at an average cooling rate of 5 to 50°C/s. Cooling may be performed with a slow cooling rate during slow cooling, so that plate shape defects caused by a rapid temperature drop during subsequent rapid cooling may be suppressed.

[0068] When an end temperature in the slow cooling is lower than 650°C, diffusion activity of carbon is low due to the too low temperature, so that a carbon concentration in ferrite is increased, but as the carbon temperature in austenite is decreased, so that a fraction of a hard phase is excessive to increase a yield ratio, resulting in a higher tendency to crack occurrence during processing. In addition, a difference in a temperature thereof between the soaking section and the cooling section may be excessively high, so that there may be a problem in that the shape of the plate is non-uniform. When the end temperature is higher than 700°C, there is a disadvantage in that an excessively high cooling rate is required during subsequent cooling (rapid cooling). In addition, when the average cooling rate during the slow cooling exceeds 10°C/s, carbon diffusion may not sufficiently occur. Meanwhile, considering productivity, the slow cooling may be performed at an average cooling rate of 1°C/s or more.

[0069] After completing the slow cooling, rapid cooling is performed. When the rapid cooling end temperature is lower than 300°C, a cooling deviation may occur in a width direction and a length direction of the steel sheet, so that a plate shape may be deteriorated. On the other hand, when the temperature is higher than 580°C, a hard phase may not be sufficiently secured, so that the strength may be lowered. Meanwhile, when the average cooling rate during the rapid cooling is less than 5°C/s, there is a risk that a fraction of the hard phase may be excessive, but when the average cooling rate exceeds 50°C/s, there is a risk that the hard phase may be rather insufficient.

[0070] Meanwhile, in the annealing process, after cooling is completed, an overaging treatment (OAS) may be performed if necessary. The overaging treatment is a process of maintaining for a certain period of time after the rapid cooling end temperature. The overaging treatment does not require any separate treatment, and may be the same as a type of air-cooling treatment. By performing the overaging treatment, a coil is homogenized in a width direction and a length direction of the coil, thereby improving the shape quality. To this end, the overaging treatment may be performed for 200 to 800 seconds.

Plating

[0071] After the annealing, a plating layer can be formed through a plating process. The plating method includes a hot-dip plating method in which a plating bath is installed during annealing and the steel sheet is dipped into a hot-dip plating solution and a method of electroplating in an electrolyte after annealing is completed. LME that occurs during spot welding can occur when there is molten zinc, which is thus unrelated to the method for manufacturing a plated steel sheet. The plating conditions are not particularly limited as long as they are generally known in the technical field to which the present disclosure pertains.

Mode for Invention

[0072] 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 scope of rights of the present disclosure. The reason is that the scope of rights of the present disclosure is determined by the matters described in the claims and reasonably inferred therefrom.

(Example)

[0073] A steel slab having the alloy composition (with a unit of weight %, the remainder, not shown in Table 1 is Fe and inevitable impurities), shown in Table 1 was prepared and then heated at a temperature of 1200°C for 1 hour. Then, a finish rolling temperature was set to a temperature within a range of 800 to 920°C in a central portion of a material thereof, under the conditions shown in Table 2 below, and a process of spraying water during finish rolling was applied to a surface layer thereof.

[0074] The manufactured hot-rolled steel sheet was cooled at a cooling rate of 0.1°C/s and wound at a temperature of 650°C. Thereafter, the wound hot-rolled steel sheet was cold rolled at a reduction ratio of 40% and 80% to manufacture a cold-rolled steel sheet.

[0075] The manufactured cold-rolled steel sheet was heated to a temperature within a range of 730 to 860°C, and

heat treated under the annealing temperature conditions shown in Table 2. For the annealing heat treatment, a temperature in each stage in the heating section (HS), soaking section (SS), slow cooling section (SCS), rapid cooling section (RCS), and overaging section (OAS) in FIG. 1 was shown in Table 2. Meanwhile, slow cooling (SCS section in Table 2) was performed at an average cooling rate of 3°C/s, and rapid cooling (RCS section in Table 2) was performed at an average cooling rate of 20°C/s.

[0076] Meanwhile, in order to evaluate the LME characteristics, electroplating was performed on a surface of the steel sheet to form a zinc plating layer with a thickness of 5 to 7 μm.

[Table 1]

Steel type	C	Si	Mn	P	S	Cr	Ti	Nb	Ac1 (°C)	Ar3 (°C)
1	0.07	0.18	2.4	0.01	0.0052	0.98	0.015	0.08	719.1	679
2	0.092	0.27	2.41	0.011	0.004	1.02	0.042	0.04	722.3	671

[Table 2]

Steel type	Hot rolling conditions			Cold rolling reduction ratio (%)	Temperature for each step of annealing process (°C)					Remarks
	Finish rolling outlet temperature	Whether to spray water during finish rolling	Steel sheet surface temperature during finish rolling		HS	SS	SCS	RCS	OAS	
2	903	Sprayed	Higher than Ar3	40	730	730	650	450	360	Comparative example 1
1	902	Not sprayed	Higher than Ar3	40	750	750	650	450	360	Comparative example 2
2	903	Sprayed	Lower than Ar3	40	790	790	650	450	360	Comparative example 3
1	902	Not sprayed	Higher than Ar3	40	800	800	650	450	360	Comparative example 4
2	903	Not sprayed	Higher than Ar3	40	820	820	650	450	360	Comparative example 5
1	902	Not sprayed	Higher than Ar3	40	840	840	650	450	360	Comparative example 6
1	902	Sprayed	Lower than Ar3	40	860	860	650	450	360	Comparative example 7
1	902	Sprayed	Lower than Ar3	80	730	730	650	450	360	Inventive example 1
2	903	Sprayed	Lower than Ar3	80	750	750	650	450	360	Inventive example 2
1	902	Sprayed	Lower than Ar3	80	770	740	650	450	360	Inventive example 3

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(continued)

Steel type	Hot rolling conditions			Cold rolling reduction ratio (%)	Temperature for each step of annealing process (°C)					Remarks
	Finish rolling outlet temperature	Whether to spray water during finish rolling	Steel sheet surface temperature during finish rolling		HS	SS	SCS	RCS	OAS	
2	903	Sprayed	Lower than Ar3	80	790	790	650	450	360	Comparative example 8
1	902	Not sprayed	Higher than Ar3	80	810	810	650	450	360	Comparative example 9
2	903	Not sprayed	Higher than Ar3	80	830	830	650	450	360	Comparative example 10
1	902	Not sprayed	Higher than Ar3	80	850	850	650	450	360	Comparative example 11

[0077] The microstructure of each steel sheet manufactured as described above was observed, the mechanical properties and the plating properties thereof were evaluated, and the results were shown in the Table 3 below.

[0078] At this time, a tensile test for each specimen was performed at a strain rate of 0.01/s after collecting a tensile specimen of a JIS No. 5 size in a direction perpendicular to a rolling direction.

[0079] In order to observe the structure of the manufactured steel sheet, each fraction thereof was measured using an SEM and an image analyzer after nital etching. Meanwhile, a depth of a sound layer in a surface layer portion of the manufactured steel sheet was measured using an optical microscope. LME was observed by performing spot welding under the same condition, and then cutting the spot welding zone and observing a cross-section thereof with an optical microscope to check whether cracks occurred in the surface layer due to LME.

[Table 3]

Steel type	Structure at point of 1/4 of thickness of steel sheet		Steel sheet surface layer portion	Mechanical properties			Whether to occur LME cracks	Division
	Fraction of soft phase (%)	Fraction of hard phase (%)	Thickness of sound layer (μm)	YS (MPa)	TS (MPa)	elongation (%)		
2	72	28	0	611	1325	9.8	Occur	Comparative example 1
1	65	35	0	602	1135	10.5	Occur	Comparative example 2
2	52	48	16	683	1048	11.3	Not occur	Comparative example 3

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(continued)

Steel 1 type	Structure at point of 1/4 of thickness of steel sheet		Steel sheet surface layer portion	Mechanical properties			Whether to occur LME cracks	Division
	Fraction of soft phase (%)	Fraction of hard phase (%)	Thickness of sound layer (μm)	YS (MPa)	TS (MPa)	elongation (%)		
1	46	54	0	736	1072	14	Occur	Comparative example 4
2	35	66	0	779	1087	7.7	Occur	Comparative example 5
1	15	85	0	804	1109	7.6	Occur	Comparative example 6
1	6	95	15	809	1125	8.9	Not occur	Comparative example 7
1	76	24	17	577	782	22.7	Not occur	Inventive example 1
2	70	30	15	408	830	20.7	Not occur	Inventive example 2
1	65	35	18	543	965	18.1	Not occur	Inventive example 3
2	57	43	16	667	1055	13.1	Not occur	Comparative example 8
1	45	55	0	720	1078	12.7	Occur	Comparative example 9
2	25	75	0	769	1089	8.3	Occur	Comparative example 10
1	18	82	0	836	1128	7.3	Occur	Comparative example 11

[0080] In Table 3, YS means yield strength, TS means tensile strength, and whether LME occurred or not was determined by the cracks formed when plated zinc melts during spot welding and penetrates into grain boundaries of a base material.

[0081] As shown in Tables 1 to 3, in Inventive Examples 1 to 3, which satisfy all of the suggestions in the present disclosure, it can be confirmed that excellent physical properties may be secured and excellent spot weldability may be secured by preventing cracking due to LME during spot welding. FIG. 2 is an image obtained by observing the surface layer of Inventive Example 1, and it can be confirmed that a sound layer is formed therein.

[0082] In Comparative Examples 1, 2, 4 to 6 in which it was manufactured using a conventional process without water cooling during hot rolling, so a sound layer was not formed in the surface layer portion, resulting in defects due to sensitivity to LME during spot welding. In addition, in terms of a material, a cold rolling reduction ratio is low, so if an

annealing temperature is lowered, ferrite recrystallization is insufficient and austenite is formed, which ensures strength but has a problem of low elongation. In particular, FIG. 3 is an image obtained by observing the microstructure of the surface layer of Comparative Example 1, and it can be seen that a sound layer is not formed in the surface layer portion.

[0083] In Comparative Examples 3 and 7, water was sprayed during hot rolling to form a sound ferrite layer on the surface, so LME cracks were not observed. However, the reduction ratio was low, recrystallization was slow in the heating section, and a fraction of the hard phase fraction was excessively high, resulting in low elongation.

[0084] In Comparative Example 8, water was sprayed during hot rolling to form a sound layer on the surface, and no LME cracks were observed. Cold rolling was performed at a high reduction ratio, and annealing was performed at a continuous annealing temperature, a high temperature exceeding $Ac1+30^{\circ}C$, and it was confirmed that the fraction of the hard phase therein was high so the elongation was inferior.

[0085] In Comparative Examples 9 to 11 in which it was manufactured using a conventional process without water cooling during hot rolling, so a sound layer was not formed in the surface layer portion, so that defects were formed due to sensitivity to LME during spot welding. In addition, cold rolling was performed at a high reduction ratio, but the continuous annealing temperature exceeded $Ac1 + 30^{\circ}C$, and the internal hard phase fraction was high, resulting in poor elongation.

Claims

1. A high-strength and high-formability steel sheet having excellent spot weldability, comprising by weight:

0.05 to 0.10% of carbon (C), 0.3% or less (excluding 0%) of silicon (Si), 2.0 to 2.5% of manganese (Mn), 0.05% or less (excluding 0%) of titanium (Ti), 0.1% or less (excluding 0%) of niobium (Nb), 1.5% or less (excluding 0%) of chromium (Cr), 0.1% or less of phosphorus (P), and 0.01% or less of sulfur (S), with a balance of Fe and inevitable impurities,

wherein a microstructure of the steel sheet at a point of $1/4$ of a thickness (t) of the steel sheet includes, by volume fraction, 65 to 85% of a soft phase and a remainder of a hard phase,

wherein a surface layer portion of the steel sheet includes a sound layer, and a thickness of the sound layer is 5 to 50 μm .

2. The high-strength and high-formability steel sheet having excellent spot weldability of claim 1, wherein the sound layer is a ferrite main phase with a grain size of 6 to 20 μm .

3. The high-strength and high-formability steel sheet having excellent spot weldability of claim 1, wherein the soft phase comprises, by area fraction, 60% or more of recrystallized ferrite and 5% or less of non-recrystallized ferrite.

4. The high-strength and high-formability steel sheet having excellent spot weldability of claim 1, wherein the hard phase comprises martensite or a mixed structure of martensite and a trace amount of bainite.

5. The high-strength and high-formability steel sheet having excellent spot weldability of claim 1, wherein the hard phase has an aspect ratio of 1.2 or less.

6. The high-strength and high-formability steel sheet having excellent spot weldability of claim 1, further comprising: a plating layer on a surface of the steel sheet.

7. The high-strength and high-formability steel sheet having excellent spot weldability of claim 1, wherein the steel sheet has a tensile strength (TS) of 780 MPa or more and an elongation (EI) of 18% or more.

8. A method for manufacturing a high-strength and high-formability steel sheet having excellent spot weldability, the method comprising:

heating a steel slab comprising, by weight: 0.05 to 0.10% of carbon (C), 0.3% or less (excluding 0%) of silicon (Si), 2.0 to 2.5% of manganese (Mn), 0.05% or less (excluding 0%) of titanium (Ti), 0.1% or less (excluding 0%) of niobium (Nb), 1.5% or less (excluding 0%) of chromium (Cr), 0.1% or less of phosphorus (P), and 0.01% or less of sulfur (S), with a balance of Fe and inevitable impurities, to a temperature within a range of 1100 to $1300^{\circ}C$; hot rolling the heated steel slab, and cooling and hot rolling the steel slab so that a surface temperature of a material thereof is $Ar3$ or lower for a predetermined period of time during finish hot rolling; cooling and cooling the hot-rolled steel sheet after the hot rolling;

cold rolling the cooled hot-rolled steel sheet at a reduction ratio of 70 to 90%;
heating and maintaining the cold-rolled steel sheet to a temperature range of A_{c1} to $A_{c1}+30^{\circ}\text{C}$ after the cold
rolling; and
slowly cooling the cold-rolled steel sheet at an average cooling rate of 1 to 10°C/s to a temperature within a
range of 650 to 700°C , and then rapidly cooling the cold-rolled steel sheet at an average cooling rate of 5 to
 50°C/s to a temperature within a range of 300 to 580°C .

9. The method for manufacturing a high-strength and high-formability steel sheet having excellent spot weldability of
claim 8, wherein the temperature of the material during the finish hot rolling is A_{r3} to 1000°C .
10. The method for manufacturing a high-strength and high-formability steel sheet having excellent spot weldability of
claim 8, wherein the coiling is performed at a temperature within a range of 400 to 700°C and the cooling is performed
at a cooling rate 0.1°C/s or less.
11. The method for manufacturing a high-strength and high-formability steel sheet having excellent spot weldability of
claim 8, further comprising:
performing an over-aging treatment for 200 to 800 seconds after the rapid cooling.
12. The method for manufacturing a high-strength and high-formability steel sheet having excellent spot weldability of
claim 8, wherein, during the finish hot rolling, the surface is cooled by spraying water at least once between rolling
passes.
13. The method for manufacturing a high-strength and high-formability steel sheet having excellent spot weldability of
claim 8, further comprising:
forming a plating layer.

FIG. 1

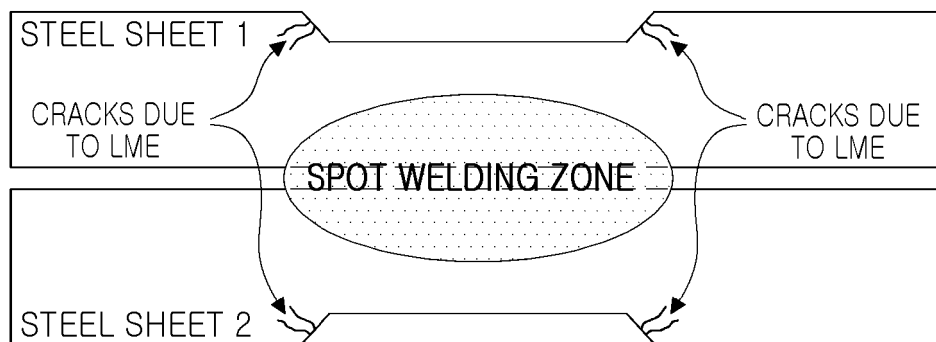


FIG. 2

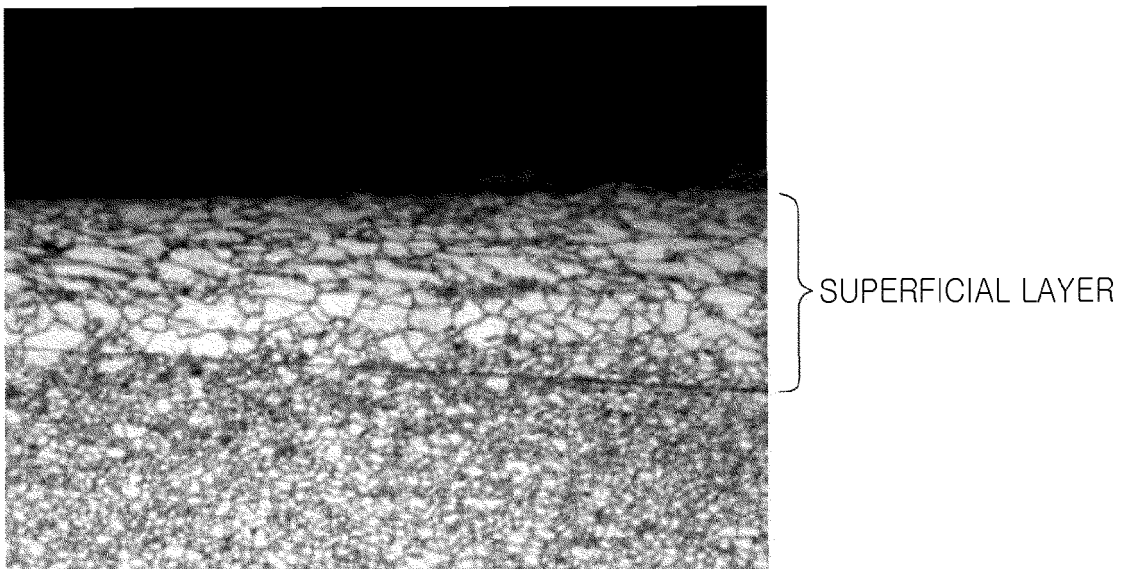


FIG. 3

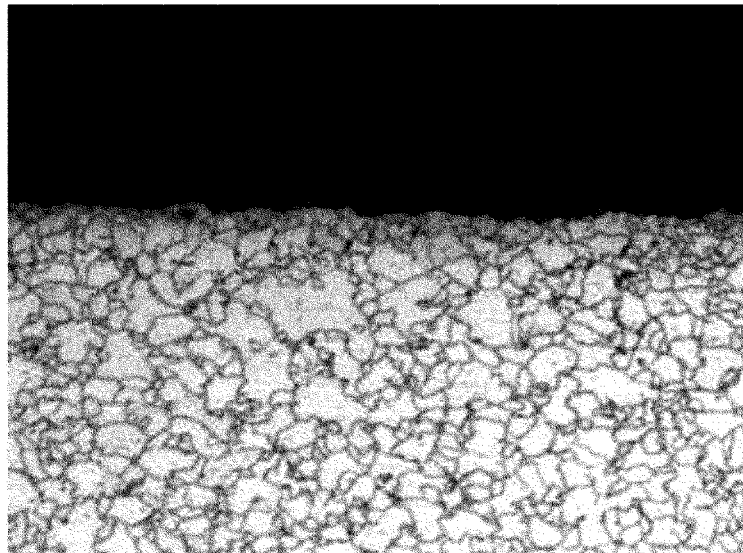


FIG. 4

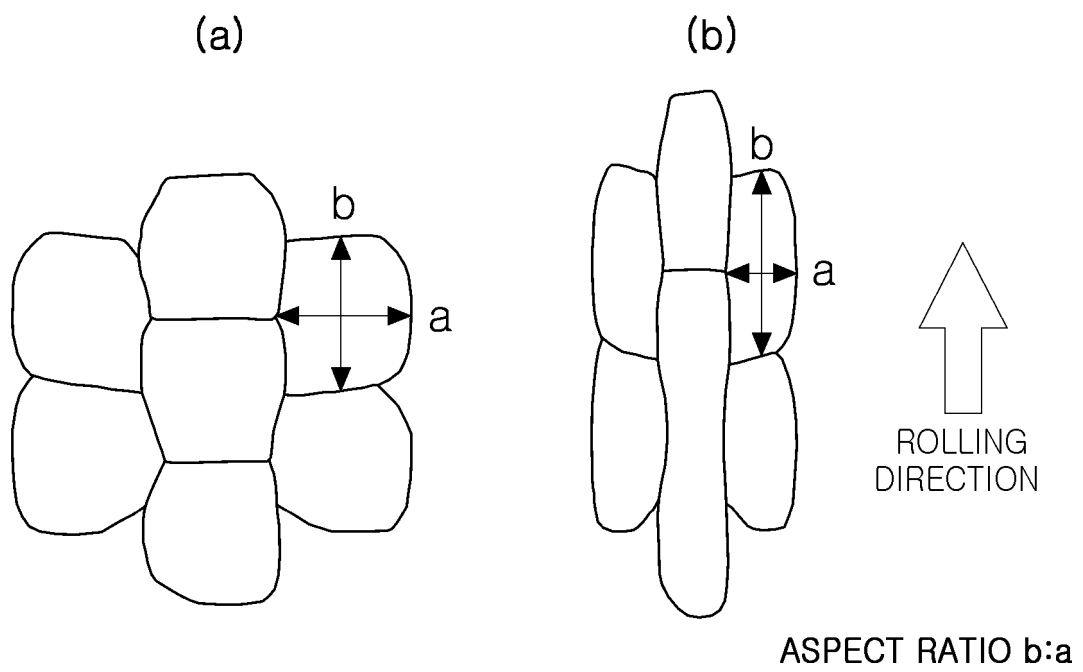


FIG. 5

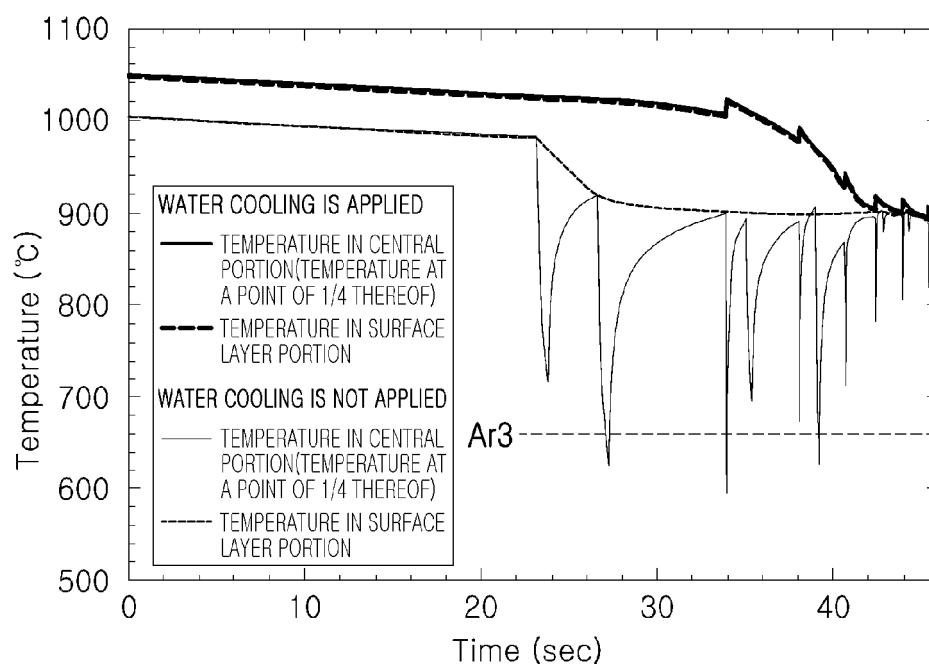


FIG. 6

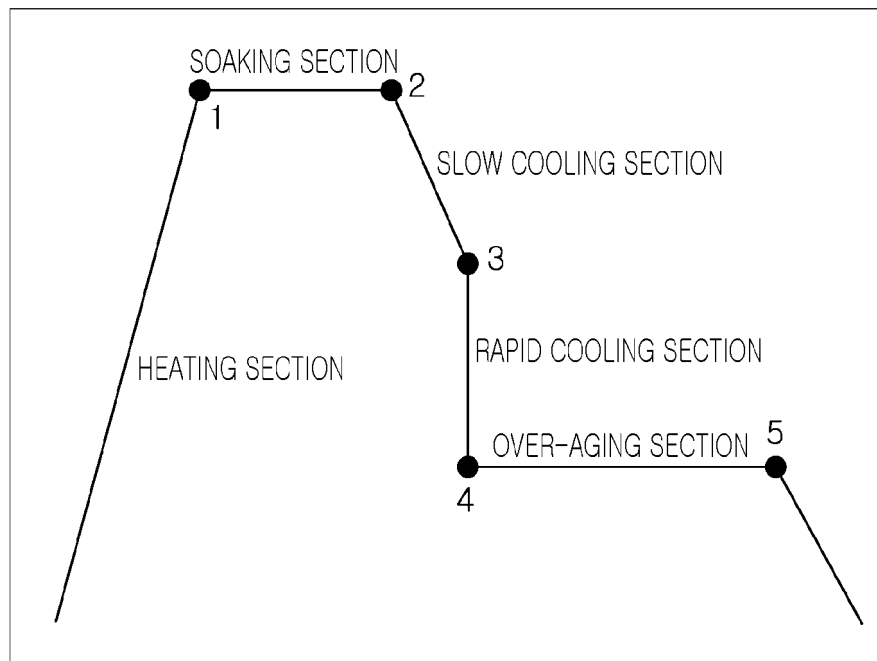
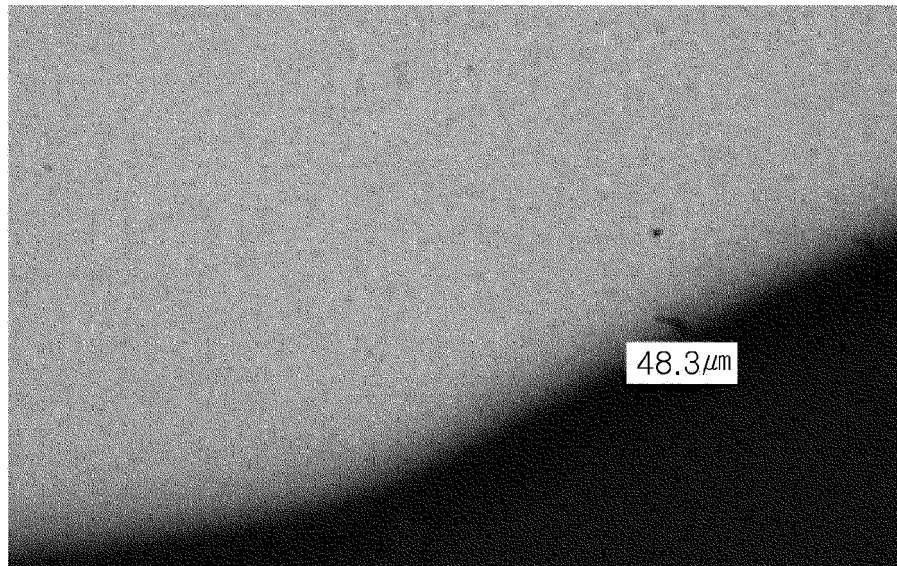


FIG. 7



INTERNATIONAL SEARCH REPORT

International application No.

PCT/KR2022/020731

A. CLASSIFICATION OF SUBJECT MATTER C22C 38/38(2006.01)i; C22C 38/28(2006.01)i; C22C 38/26(2006.01)i; C21D 8/02(2006.01)i; C21D 1/667(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); B21B 1/22(2006.01); B21B 1/26(2006.01); B21B 3/00(2006.01); B21B 45/02(2006.01); C21D 8/02(2006.01); C22C 38/00(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: 강판(steel plate), 점 용접 성(spot welding characteristic), 냉각(cooling), 열간압연(hot rolling), 권취(winding), 냉간압연(cold rolling), 소둔(annealing), 물 분사(water jetting)																		
C. DOCUMENTS CONSIDERED TO BE RELEVANT <table border="1"> <thead> <tr> <th>Category*</th> <th>Citation of document, with indication, where appropriate, of the relevant passages</th> <th>Relevant to claim No.</th> </tr> </thead> <tbody> <tr> <td>A</td> <td>KR 10-2014-0041833 A (NIPPON STEEL & SUMITOMO METAL CORPORATION) 04 April 2014 (2014-04-04) See claims 1-9.</td> <td>1-13</td> </tr> <tr> <td>A</td> <td>KR 10-2016-0041850 A (JFE STEEL CORPORATION) 18 April 2016 (2016-04-18) See claims 1-4.</td> <td>1-13</td> </tr> <tr> <td>A</td> <td>JP 2005-105367 A (NIPPON STEEL CORP.) 21 April 2005 (2005-04-21) See paragraphs [0017]-[0029].</td> <td>1-13</td> </tr> <tr> <td>A</td> <td>KR 10-2013-0074503 A (POSCO) 04 July 2013 (2013-07-04) See paragraph [0079]; and claim 4.</td> <td>1-13</td> </tr> <tr> <td>A</td> <td>KR 10-0778680 B1 (POSCO) 22 November 2007 (2007-11-22) See paragraph [0021]; claims 1-3; and figures 1-2.</td> <td>1-13</td> </tr> </tbody> </table>	Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.	A	KR 10-2014-0041833 A (NIPPON STEEL & SUMITOMO METAL CORPORATION) 04 April 2014 (2014-04-04) See claims 1-9.	1-13	A	KR 10-2016-0041850 A (JFE STEEL CORPORATION) 18 April 2016 (2016-04-18) See claims 1-4.	1-13	A	JP 2005-105367 A (NIPPON STEEL CORP.) 21 April 2005 (2005-04-21) See paragraphs [0017]-[0029].	1-13	A	KR 10-2013-0074503 A (POSCO) 04 July 2013 (2013-07-04) See paragraph [0079]; and claim 4.	1-13	A	KR 10-0778680 B1 (POSCO) 22 November 2007 (2007-11-22) See paragraph [0021]; claims 1-3; and figures 1-2.	1-13
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☐ Further documents are listed in the continuation of Box C. ☒ See patent family annex.																		
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<table border="1"> <tr> <td>Date of the actual completion of the international search 23 March 2023</td> <td>Date of mailing of the international search report 24 March 2023</td> </tr> </table>	Date of the actual completion of the international search 23 March 2023	Date of mailing of the international search report 24 March 2023																
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<table border="1"> <tr> <td> Name and mailing address of the ISA/KR Korean Intellectual Property Office Government Complex-Daejeon Building 4, 189 Cheongsaro, Seo-gu, Daejeon 35208 Facsimile No. +82-42-481-8578 </td> <td> Authorized officer Telephone No. </td> </tr> </table>	Name and mailing address of the ISA/KR Korean Intellectual Property Office Government Complex-Daejeon Building 4, 189 Cheongsaro, 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/KR2022/020731

Patent document cited in search report	Publication date (day/month/year)	Patent family member(s)	Publication date (day/month/year)
KR 10-2014-0041833 A	04 April 2014	BR 112014001994 A2	21 February 2017
		CA 2842897 A1	07 February 2013
		CA 2842897 C	20 September 2016
		CN 103717773 A	09 April 2014
		CN 103717773 B	04 May 2016
		EP 2738280 A1	04 June 2014
		EP 2738280 A4	05 August 2015
		EP 2738280 B1	20 March 2019
		ES 2727865 T3	21 October 2019
		JP 5273324 B1	28 August 2013
		KR 10-1597473 B1	24 February 2016
		MX 2014000882 A	20 February 2014
		MX 360332 B	29 October 2018
		PL 2738280 T3	30 August 2019
		RU 2014-106991 A	10 September 2015
		RU 2569615 C2	27 November 2015
		TW 201309815 A	01 March 2013
		TW I493055 B	21 July 2015
		US 2014-0212684 A1	31 July 2014
		US 9234268 B2	12 January 2016
		WO 2013-018739 A1	07 February 2013
KR 10-2016-0041850 A	18 April 2016	CN 105102662 A	25 November 2015
		EP 2987887 A1	24 February 2016
		EP 2987887 A4	14 September 2016
		EP 2987887 B1	11 September 2019
		JP 2014-205889 A	30 October 2014
		JP 2014-205890 A	30 October 2014
		JP 5867444 B2	24 February 2016
		JP 5870955 B2	01 March 2016
		MX 2015014437 A	03 February 2016
		US 2016-0076124 A1	17 March 2016
		WO 2014-171063 A1	23 October 2014
JP 2005-105367 A	21 April 2005	BR PI0414674 A	28 November 2006
		BR PI0414674 B1	01 November 2016
		CA 2540762 A1	07 April 2005
		CA 2540762 C	18 September 2012
		CA 2747654 A1	07 April 2005
		CA 2747654 C	21 April 2015
		CN 102011053 A	13 April 2011
		CN 102011053 B	24 July 2013
		CN 1860249 A	08 November 2006
		CN 1860249 B	19 September 2012
		EP 1681363 A1	19 July 2006
		EP 1681363 A4	25 November 2009
		EP 1681363 B1	11 January 2012
		EP 2309012 A1	13 April 2011
		EP 2309012 B1	12 September 2012
		ES 2391164 T3	22 November 2012
		JP 2005-105361 A	21 April 2005
		JP 4486334 B2	23 June 2010

Form PCT/ISA/210 (patent family annex) (July 2022)

INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.

PCT/KR2022/020731

Patent document cited in search report	Publication date (day/month/year)	Patent family member(s)	Publication date (day/month/year)
		JP 4486336 B2	23 June 2010
		KR 10-1094594 B1	15 December 2011
		KR 10-1165166 B1	11 July 2012
		KR 10-1165168 B1	11 July 2012
		KR 10-2006-0096002 A	05 September 2006
		KR 10-2008-0035017 A	22 April 2008
		KR 10-2011-0018463 A	23 February 2011
		KR 10-2011-0028643 A	21 March 2011
		MX 344641 B	04 January 2017
		MX PA06003566 A	14 June 2006
		TW 200516158 A	16 May 2005
		TW I302572 B	01 November 2008
		US 2007-0029015 A1	08 February 2007
		US 2011-0232807 A1	29 September 2011
		US 8084143 B2	27 December 2011
		US 8747577 B2	10 June 2014
		WO 2005-031024 A1	07 April 2005
KR 10-2013-0074503 A	04 July 2013	CN 104024452 A	03 September 2014
		CN 104024452 B	02 March 2016
		JP 2015-507094 A	05 March 2015
		JP 5865516 B2	17 February 2016
		KR 10-1353787 B1	22 January 2014
		WO 2013-100485 A1	04 July 2013
KR 10-0778680 B1	22 November 2007	None	

Form PCT/ISA/210 (patent family annex) (July 2022)