The present invention provides a new high-strength and Si—Cr-containing hot rolled steel plate provided with higher strength as well as excellent workability and a method for manufacturing the steel plate. The high-strength steel plate can be obtained by controlling the particle size of prior austenite to be 10 µm or less, and properly selecting the coiling temperature. The steel plate obtained includes a retained austenite phase in a volume fraction of 5% to 20%; a martensite phase in a volume fraction equal to or less than 10%; and a bainite phase in the remaining volume fraction. The particle size of the retained austenite particle is 1 µm or less and the retained austenite particles are dispersed uniformly.
Graph for schematically showing a temperature history of hot rolling in a manufacturing process of one embodiment of the present invention.

**FIG. 1**
PHOTOGRAPH OF PRIOR AUSTENITE PARTICLES OF A CROP AT A FINISH ENTRANCE.

FIG. 2
PHOTOGRAPH OF PRIOR AUSTENITE PARTICLES,

FIG. 3
GRAPH SHOWING A RELATIONSHIP OF FINISH ROLLING TEMPERATURES AND ANISOTROPY OF ELONGATION.

FIG. 4
ROLLING SIZE : 1.2 m/m

Each numerical value in parentheses denotes a reduction ratio.

EXIT THICKNESS AT ROUGHING MILL 34 m/m
GREATER REDUCTION AMOUNT ON A FINISH REAR STAGE

EXIT THICKNESS AT ROUGHING MILL 24 m/m
NORMAL REDUCTION AMOUNT ON A FINISH REAR STAGE

GRAPH SHOWING TEMPERATURE CHANGES IN EACH ROLLING MILL
FROM A FINISH ENTRANCE SIDE UP TO A FINISH DISCHARGING SIDE.

FIG. 5
FIG. 6

RELATIONSHIP OF THE DISLOCATION DENSITY AND THE PARTICLE SIZE OF PRIOR AUSTENITE PARTICLES.
a) MARTENSITE–RICH STRUCTURE

b) LOWER–MARTENSITE–CONTAINING STRUCTURE

c) FERRITE–CONTAINING STRUCTURE

STRUCTURE IMAGE OBTAINED BY AN OPTICAL MICROSCOPE

STRUCTURE IMAGE OBTAINED BY AN SEM MICROSCOPE

PHOTOGRAPH OF TYPICAL STRUCTURES OF SECTIONS

FIG. 7
PHOTOGRAPH OF SHOWING THE IMAGE QUALITY

PHOTOGRAPH OF SHOWING THE PHASE DISTRIBUTION
(WHITE: RETAINED AUSTENITE, GRAY: FERRITE)

PHOTOGRAPH OF STRUCTURES OF SECTIONS EACH OBTAINED BY AN EBSM METHOD FOR A STEEL PLATE MANUFACTURED UNDER COMPOSITIONAL AND ROLLING CONDITIONS ACCORDING TO THE PRESENT INVENTION.

FIG. 8
FIG. 9

FIG. 10

RELATIONSHIP OF A COEFFICIENT OF FRICTION AND A REDUCTION RATIO, DEPENDING ON TYPES OF ROLLS.
(a) Balance between the strength and the ductility (a), and relationship of the ductility and the amount of retained austenite (b)

FIG. 11
HIGH-STRENGTH HOT ROLLED STEEL PLATE AND MANUFACTURING METHOD THEREOF

CROSS REFERENCE TO RELATED APPLICATIONS

[0001] This application is based upon the prior Japanese Patent Application No. 2007-108759 filed on Apr. 17, 2007, the entire contents of which are incorporated herein by reference.

BACKGROUND OF THE INVENTION

[0002] 1. Field of the Invention

[0003] The present invention relates to a high-strength hot rolled steel plate having high tension strength and superior workability, and also relates to a manufacturing method thereof.

[0004] 2. Description of the Related Art

[0005] Recent demands for a high-strength steel plate which can exhibit superior workability will be described, with respect to cars, by way of example. In view of environmental protection for the earth, it should be required to reduce the amount of exhaust gas, such as CO₂, or the like, in the field of automobiles. To this end, it is quite essential to further reduce the weight of each car body. In order to achieve the weight reduction of the car body, it is necessary to enhance the strength of steel plate used for the car body so as to reduce the thickness of the steel plate. In addition, safety for the users should be secured in the car. Also for this purpose, the strength of the steel plate must be further improved.

[0006] However, increase of the strength of the steel plate may tend to degrade its workability, and it will be difficult to apply a higher strength steel plate to cold working, such as general press molding or the like.

[0007] Hot pressing is a hot press working process and usually generates a quite small amount of spring back, thus exhibiting preferable shape freezing properties. In addition, due to a hardening effect upon the hot pressing, this method can present parts having significantly higher strength, with high accuracy. However, this process requires heating the steel plate prior to subjecting it to the hot press working, and also requires reduction of the manufacturing scale after the hot press working. Thus, this process has possibility to significantly deteriorate the working efficiency. Furthermore, shorter life of the mold, which should contact with a heated steel plate, inevitably increase the manufacturing cost.

[0008] The elongation of the steel plate after the hot press working will be decreased, as such a hot pressed member may tend to be broken only due to slight deformation caused by being subjected to some external force, impact or the like. Therefore, the steel plate of this type is generally assessed to be of poor impact absorbing ability. Accordingly, it is quite difficult to use such hot pressed parts as key components for securing the safety for cars or the like.

[0009] As a method for enhancing the strength, reinforcement by a solid solution treatment, reinforcement utilizing precipitation, reinforcement by grain refinement, and reinforcement utilizing a low-temperature transforming phase can be mentioned. It is not possible to manufacture the steel plate, for which significantly enhanced strength is required, only by employing a reinforcing mechanism, including the solution treatment or precipitation requiring addition of a greater amount of alloys. Even in the case of utilizing the reinforcement by grain refinement, the improvement of the strength is limited although achieved to some extent. While the reinforcement by utilizing a low-temperature transforming phase is highly effective for manufacturing the steel plate exceeding 1200 MPa, this method can not be expected, in order to enhance ductility which can be balanced with such improvement of the strength.

[0010] Generally, higher strength of the steel plate may tend to lower ductility, as such degrading the workability.

[0011] As known materials having enhanced ductility among high-strength steel plates, there are a dual phase steel plate consisting of ferrite and martensite phases, and a transformation induced plasticity (TRIP) steel plate consisting of ferrite, bainite and retained austenite phases.

[0012] The dual phase steel plate is formed by finely dispersing the hard martensite phase in the ferrite phase. Due to the highly hard martensite phase, significant work hardening is caused upon transformation, thus providing higher ductility to the steel plate.

[0013] Examples of the TRIP steel plate are described in Patent Documents 1 and 2. The steel plate of this type containing the retained austenite phase exhibits highly excellent ductility and moldability both attributed to working induced transformation, depending on the amount the retained austenite phase and the stability to the deformation.

[0014] However, if attempting to obtain the steel plate having strength greater than 1200 MPa, delayed fracture may be caused. The term “delayed fracture” means a phenomenon wherein while cracking and/or fracture is not generated upon working and assembly for respective members, it appears suddenly during use of them. A high-strength steel plate disclosed in Patent Document 3 is intended to provide more preferable anti-delayed fracture properties, by reducing a soft phase, such as the ferrite phase, as much as possible, and by controlling the volume fraction of the retained austenite phase to be less than 4%, relative to the low-temperature transforming phase, such as the bainite phase and/or the tempered martensite phase.


[0018] As the steel plate whose elongation properties in the cold working are enhanced while keeping higher strength, the dual phase steel plate and TRIP steel plate described above can be mentioned.

[0019] In the dual phase steel plate, a higher strength can be achieved even in the case of a smaller additional amount of alloys, as well as more uniform elongation properties can be obtained due to the work hardening.

[0020] The TRIP steel plate exhibits higher ductility and has more excellent deep drawing properties. Therefore, this material is suitable for providing a part or member for which a complicated shape, higher workability and more enhanced strength are required.

[0021] The TRIP steel plate described in the Patent Document 1 is manufactured by a method comprising: creating the ferrite phase in the austenite phase by holding a raw material at 450 to 650°C. for 4 to 20 seconds in a cooling step after rolling, cooling it to a temperature lower than 350°C., and coiling it around a rod material.

[0022] In the Patent Document 2, in order to promote formation of the ferrite phase in the austenite phase in a cooling process after the rolling, a raw material is gently cooled at Ar3 to Ar1 or subjected to a rolling completion temperature of
approximately Ar₃ then cooled to a temperature within a range of 350 to 500°C, and is wound around a rod material.

[0023] Such a TRIP steel plate has a structure in which the martensite phase, retained austenite phase and/or bainite phase is dispersed in the ferrite parent phase, and exhibits excellent strength and elongation properties.

[0024] However, under the condition of C≤0.20% which can ensure the spot welding properties, only the tensile strength as high as 800 MPa can be obtained, as such more enhanced workability should be desired. Accordingly, it is difficult to manufacture the steel plate having significantly higher strength, under such conditions.

[0025] Even in a method of gradually cooling the raw material to a temperature lower than 500°C, without providing the gentle cooling process on route after the rolling, the promotion of creating a fine ferrite phase can be achieved, if setting the rolling completion temperature at a point of approximately Ar₃. With respect to the material quality of a hot rolled steel plate which has been subjected to rolling at a temperature of approximately A₃, however, anisotropy of the material may tend to be undesirably greater.

[0026] Moreover, the hot rolled steel plate described in the Patent Document 1 exhibits lower rolling workability and has a metallic structure in which coarse ferrite particles and retained austenite particles are present continuously because of the temporary stopping for the cooling at a point of approximately A₁.

[0027] Hydrogen dissolved in the steel plate, which is likely to be a cause of the delayed fracture, is a factor of determining the crystal phase, and is trapped preferentially in the retained austenite phase. Especially, the interface between the martensite phase and ferrite phase having been subjected to the impact or working, i.e., the working induced transformation site, is considered to be a highly possible trapping site for hydrogen.

[0028] Coarser retained austenite particles will provide a more reduced ratio of the area of the interface between the martensite phase and ferrite phase having been experienced the working induced transformation, as compared with the volume of the retained austenite particles. Consequently, the concentration of hydrogen to be trapped is increased, as such presenting a greater risk of the delayed fracture. If the martensite phase and the retained austenite phase coexist continuously (in an M-A state), propagation of the fracture is likely to be promoted, thus providing a further increased risk of the fracture.

[0029] The high-strength steel plate described in the Patent Document 3 is intended to enhance the anti-delayed fracture properties by limiting the amount of the retained austenite. However, in order to obtain excellent workability while keeping higher strength, utilization of the retained austenite is substantially effective. Accordingly, it is desirable if the presence of the retained austenite will not detrimentally affect the anti-delayed fracture properties without providing any limitation as described above.

SUMMARY OF THE INVENTION

[0030] To address this challenge, the present inventors have developed a new low-alloy and higher-strength steel plate and a method of manufacturing thereof; the steel plate having a bainite phase in which seven or more of the retained austenite particles having a particle size of 1 μm or less are finely dispersed per 10 μm² (the volume fraction is within the range of from 5% to 20%), thereby exhibiting higher strength as well as more preferred workability and secure anti-delayed fracture properties.

[0031] Through many experiments, we have found that a preferably high-strength steel plate can be obtained by employing appropriate rolling conditions and selecting a proper composition of components for the steel plate. Namely, higher strength and excellent ductility as well as secure anti-delayed fracture properties can be provided to a low-alloy steel plate, by subjecting a slab having a proper composition of components to rough hot rolling under high pressure conditions, completing rear-stage higher strain rolling in a finish rolling process under high temperature conditions, starting a cooling process after air-cooling for several seconds, and coiling the processed material at an appropriate temperature.

[0032] A high-strength hot rolled steel plate of the present invention comprises: a retained austenite phase in a volume fraction of 5% to 20%; a martensite phase in a volume fraction of 0% to 10%; and a bainite phase in the remaining volume fraction, wherein particles constituting the retained austenite phase have a particle size of 1 μm or less. More preferably, the particle size of prior austenite is 10 μm or less, and the average aspect ratio of the particles is 2.0 or less.

[0033] With the control of the particle size of the austenite crystal after hot rolling to be 10 μm or less (FIG. 3), the lamellar structure of the bainite phase can be made fine. In addition, with completion of uniform bainite transformation, the retained austenite particles having a particle size of 1 μm or less can be finely and effectively dispersed in the phase with the density of seven or more particles per 10 μm² (FIG. 8).

[0034] In this manner, superior anti-delayed fracture properties can be obtained, even in the case of steel provided with a higher ductility by utilizing working induced plasticity to be caused by a relatively great amount of retained austenite.

[0035] With the control of the aspect ratio of the prior austenite particles to be 2.0 or less (FIG. 3), anisotropy of the material, which is drawn in both of the rolling direction and the direction vertical to the rolling direction, can be reduced, as such enhancing the workability (FIG. 4).

[0036] Preferably, the high-strength hot rolled steel plate of the present invention has a composition comprising: C (0.15 to 0.21 (% by weight)), Si (0.5 to 2.0), Mn (0.2 to 1.0), Cr (1.0 to 4.0), Ni (0.02 to 1.0), Mo (0.05 to 0.4), V (0.05 to 0.12), W (0.01 to 0.010), N (0.005 to 0.015), and the remaining components including Fe and other inevitable impurities.

[0037] Such a chemical composition comprising proper types and amounts of the selected components can facilitate formation of the high-strength steel plate which can include the phases described above and exhibit desired mechanical properties.

[0038] Since the alloy elements described above can constitute the desired steel plate structure of the present invention in the steps of cooling after hot rolling, and cooling the cooled material, Cr and Si having greater influence on the bainite transformation are included as major elements. With controlling of amounts of these elements, the bainite transformation can be promoted, and formation of the martensite phase can be suppressed, thereby to control the entire phase to have an aimed strength.

[0039] The effect of each component will be described below.

[0040] It is preferred that the high-strength hot rolled steel plate has the structure as described above, and also a plate
Namely, this steel plate can possess the structure described above, and hence exhibit greater strength and more excellent elongation properties.

A method for manufacturing a high-strength hot rolled steel plate according to the present invention comprises the steps of:

1. preparing a slab (rolling material) having a composition containing: C (0.13 to 0.21 % by weight), Si (0.5 to 2.0), Mn (0.2 to 1.0), Cr (1.0 to 4.0), Ni (0.02 to 1.0), Mo (0.05 to 0.4), P (0 to 0.010), S (0 to 0.003), N (0.005 to 0.015), and the remaining components including Fe and other inevitable impurities;

2. roughly rolling a steel material, under the conditions of: 1250 °C or higher of an extraction temperature of a reheating furnace; 1030 °C or higher of a discharging-side temperature of roughly rolling mills; and 30% or higher of a reduction ratio for each of roughly rolling final three passes;

3. finish rolling of the steel material under the conditions of: 950 °C or higher of a discharging-side temperature of finish rolling mills; 40% or higher of a reduction ratio for each of finish front-stage first to third rolling mills (this is the case of using six rolling mills, but first to fourth rolling mills are used in the case of using seven rolling mills) and 0.5 or greater of accumulated strain in the pressed state due to three rolling mills on a finish rear-stage; and

4. cooling the steel material by air-cooling for 2 to 6 seconds, followed by water-cooling, and cooling the steel material at a cooling temperature of 550 to 650 °C.

With the purpose of enhancing strength by obtaining a low-alloy bainite phase due to employment of a temperature history (FIG. 1) for maintaining the temperature, in the steps of rapid cooling after hot rolling by utilizing hot strip milling and cooling the material at predetermined temperature conditions, a uniform phase of bainite, in which martensite and retained austenite are finely dispersed, can be obtained, by adding Cr and Si as major alloy elements and selecting a composition containing lower Mn and Ni (FIG. 7(b)).

With control of precipitation of carbides due to addition of Si and with formation of a more uniform bainite phase, the austenite having carbon density of 0.8% or higher can be retained in a greater amount. In this way, a steel plate having enhanced strength and more excellent workability can be obtained (FIG. 11).

By controlling the hot rolling finish temperature to be 950 °C or higher, the aspect ratio of the prior austenite particles can be controlled at 2.0 or less (FIG. 3).

In order to prevent biting failure of a topmost portion of the rolling material into a roll, it is preferred that, upon the finish rolling, a reduction amount of a topmost portion of the steel material is reduced, as needed, as compared with an expected reduction amount (or reduction amount originally set for a predetermined rolling), in first to fifth rolling mills (in the case of using six stages of finish rolling mills, while first to sixth rolling mills are used in the case of using seven stages of finish rolling mills), wherein the reduction amount is increased by 10% or less, as compared with the expected amount of each rolling mill. It is also preferred that a length to be rolled in the increased reduction amount is within 5 m as measured from a biting position of the topmost portion of the rolling material.

In order to prevent slip occurrence between the rolling material and the roll during the rolling process, it is also preferred that a special high-grip roll is used as a working roll for each of finish first to third rolling mills including the final rolling mill.

Our test on the manufacture, which will be described below, demonstrates that the aforementioned high-strength steel plate can be obtained readily by employing the conditioned as provided above.

In the high-strength steel plate of the present invention, the retained austenite is incorporated in the bainite phase in a volume fraction of 5% to 20% such that it is finely dispersed with the density of seven or more particles per 10 μm². Therefore, both strength and workability, which are contrary to each other, can be provided to the steel plate, and excellent anti-delayed fracture properties can also be provided thereto.

According to a method of the present invention for manufacturing a high-strength steel plate, the high-strength steel plate described above can be readily and securely manufactured.

BRIEF DESCRIPTION OF THE DRAWINGS

The above and other objects, features and advantages of the present invention will become more apparent from the following description taken in connection with the accompanying drawings, in which:

FIG. 1 is a graph for schematically showing a temperature history of hot rolling in a manufacturing process of one embodiment of the present invention;

FIG. 2 is a photograph of prior austenite particles of a crop at a finish entrance;

FIG. 3 is a photograph of prior austenite particles;

FIG. 4 is a graph showing a relationship of finish rolling temperatures and anisotropy of elongation;

FIG. 5 is a graph showing a relationship of a rolling schedule and a rolling temperature;

FIG. 6 is a graph showing a relationship of a dislocation density and a particle size of prior austenite particles;

FIG. 7 is a photograph of typical phases of sections;

FIG. 8 is a photograph of phases of sections each obtained by an EBSP method for a steel plate manufactured under compositional and rolling conditions according to the present invention, and bright or light color portions designate retained austenite;

FIG. 9 is a graph showing a deviation of a plate thickness at a distal end of a rolling material;

FIG. 10 is a graph showing a relationship of a coefficient of friction and a reduction ratio, depending on types of rolls; and

FIG. 11 is a graph showing a balance between the strength and the ductility, and a relationship of the ductility and the amount of retained austenite.

DESCRIPTION OF THE PREFERRED EMBODIMENTS

Hereinafter, one embodiment of a sheet steel used to produce parts by working the same, for which excellent workability and anti-delayed fracture properties are required while
keeping tensile strength of 1200 MPa or higher, and a manufacturing method of the sheet steel will be described.

The steel plate has a composition containing the following components: C (0.13 to 0.21 % by weight), Si (0.5 to 2.0), Mn (0.2 to 1.0), Cr (1.0 to 4.0), Ni (0.02 to 1.0), Mo (0.05 to 0.4), P (0 to 0.010), N (0 to 0.003), and the remaining components include Fe and other inevitable impurities.

As used herein, the term “sheet plate” means a steel plate having a thickness of from 1.0 mm to 3.0 mm. The steel plate to be manufactured under the above compositional conditions can be mainly used for parts of cars, consumer electrical appliances, electronic equipment and the like, which require higher workability and strength. In addition, the steel plate can also be applied to materials for steel pipes.

First, components of the steel plate will be described.

The amount of carbon (C) should be within the range of 0.13 to 0.21%.

C is the most important component for stabilizing the retained austenite. If the amount of C is less than 0.13%, sufficient stability can not be obtained, thus an amount of C of 0.13% or greater should be required. However, if it exceeds 0.21%, a welded portion becomes too hard and is likely to be broken. Such a situation provides some limitation of use of steel plate to be formed. Therefore, the upper limit described above is provided to the amount of C. Namely, by setting the amount of C within the range of 0.13 to 0.21%, it has been found that a composite structure which accords with an intention of the present invention can be obtained.

The amount of silicon (Si) should be within the range of 0.5 to 2.0%. Si also serves to stabilize the retained austenite. In addition, Si enhances the strength to be obtained by reinforcement due to solid solution. If the amount of Si is 0.5% or greater, a preferred composite structure and material quality can be obtained. A greater amount of Si can increase more retained austenite as well as enhance the stability. However, if the amount of Si exceeds 2.0%, properties for balancing the strength and the ductility will be saturated, thus the upper limit of the Si amount should be set at 2.0% in view of reduction of the cost.

The amount of chromium (Cr) should be within the range of 1.0 to 4.0%. Cr can create the bainite phase, and enhance the strength of the steel plate to be formed therewith.

If the amount of Cr is less than 1.0%, the amount of ferrite is undesirably increased, as such making it difficult to obtain a steel plate with a desirably higher strength. Therefore, the Cr amount should be 1.0% or greater. However, if it exceeds 4.0%, the martensite phase is likely to be produced, thus making the steel plate strength too high and hence rendering the anti-delayed fracture properties insufficient. Therefore, 4.0% is set as the upper limit.

The amount of manganese (Mn) should be within the range of 0.2 to 1.0%. If the Mn amount is less than 0.2%, the manufacture of the steel plate will be difficult. Therefore, it should be 0.2% or greater.

In order to attain higher strength, it is preferred to add Mn as much as possible. However, if it is excessively added, the martensite phase may tend to be produced, thus making it impossible to obtain the intended structure according to this invention. Therefore, the upper limit of the Mn amount should be set at 1.0%.

The amount of nickel (Ni) should be within the range of 0.02 to 1.0%. Ni can enhance the strength of the steel plate by reinforcement due to solid solution. However, if the amount of Ni is too increased, the martensite phase is likely to be produced. Moreover, inadvertent addition would lead to increase of the production cost. Thus, the upper limit should be set at 1.0%.

Molybdenum (Mo) can create the bainite phase as is similar to Cr, and enhance the strength of the steel plate to be formed therewith. In addition, a hydrogen trapping effect due to Mo carbides is useful for providing anti-delayed fracture properties to the steel plate. However, inadvertent addition would cause undesirably restrained recrystallization as well as lead to increase of the cost. Therefore, the Mo amount should be set within the range of from 0.05 to 0.40%.

In order to enhance the weldability, it is necessary to possibly reduce the amount of phosphorus (P). Thus, the upper limit of this element should be set at 0.010%.

Also, in order to enhance the weldability, it is necessary to possibly reduce the amount of sulfur (S). Therefore, the upper limit of this element should be set at 0.003%.

The amount of nitrogen (N) should be within the range of 0.005 to 0.015%. As is similar to carbon, nitrogen is useful to stabilize the austenite phase. However, its excessive existence will cause degradation of the weldability. Thus, the range of this amount should be set at a value of from 0.005 to 0.015%.

A slab produced to have the composition as described above is then subjected to hot rolling after heated again or subjected to hot rolling immediately after casting.

FIG. 1 is a graph for schematically showing a temperature history of hot rolling in a manufacturing process of one embodiment of the present invention, in which particles sizes of prior austenite are also designated. The horizontal axis denotes the elapsed time and the vertical axis denotes the temperature.

Upon providing the hot rolling, the extraction temperature of the reheating furnace was set at 1250°C. This temperature was selected to preferentially secure the surface temperature of 950°C after the finish, even though some inevitable growth of austenite particles would be caused in the reheating furnace due to such a high temperature condition. However, the size or diameter of the austenite particles will be lessened in the following rolling process. Therefore, it is necessary to reduce the particle size of the prior austenite as finely as possible before subjecting it to a finishing rolling mill. Accordingly, in a roughly rolling process, the crystal particle size is reduced in advance to 35 µm or less, by setting the reduction ratio of each of final three passes for roughly rolling at 30% or greater, at a discharging-side temperature of 1030°C or higher on the discharging side of the roughly rolling mills. FIG. 2 shows a particle size of prior austenite after subjected to the roughly rolling, wherein the processed material was cut by a pre-finish crop shearing machine.

For first to third rolling mills on a finish front-stage (in the case of using six finish rolling mills, but first to fourth rolling mills are used in the case of using seven finish rolling mills), the reduction ratio per mill is set at 40% or higher. Accumulated strain in the pressed state for three rolling mills of a finish rear-stage is set at 0.5 or greater, and the finishing rolling mill discharging-side temperature is securely set at 950°C or higher, so as to render the austenite particle size equal to or less than 10 µm. In addition, air-cooling is provided for 2 to 6 seconds after the finish rolling, followed by water-cooling. A cooling temperature is set at 550°C to 650°C. In the above air-cooling step, the size of the austenite
particles is also controlled. Namely, during the hot rolling step, the particle size of the prior austenite is controlled to be 10 \( \mu \text{m} \) or less before post-hot-rolling hot run cooling is started, so as to control the size of the prior austenite particles to eliminate working strain.

[0087] FIG. 3 shows a result of observation for the prior austenite particles of the steel plate according to the present invention by using an SEM phase observation. An average particle size of the prior austenite particles is 9.3 \( \mu \text{m} \), presenting a uniformly granulated structure. An average aspect ratio of the major axis/the short axis is 1.7.

[0088] In the case of using low-temperature rolling, such that a rolling completion temperature is 950°C or lower, and employing the accumulated strain, set at 0.5 or less, of three rolling mills on a finish rear-stage, the austenite particles become larger (10 \( \mu \text{m} \) or larger), and the shape of each austenite particle may tend to be flat due to rolling, thus causing increased anisotropy. FIG. 4 shows a relationship between the finish rolling mill discharging-side temperature (FDT) and anisotropy of elongation. As is seen from FIG. 4, the anisotropy of elongation appears when the FDT is 950°C or lower. This anisotropy is defined by an equation of \( C = (\omega)(L)/(C + L)/2 \) (\( L \) is an elongation in the rolling direction, and \( C \) is an elongation in the direction vertical to the rolling direction). A smaller value to be obtained from this equation shows less anisotropy.

[0089] As used herein, the "strain" means a value designated by \( \varepsilon \) in the following equation:
\[
\varepsilon = (h_0 - h_1)/(h_0 + h_1)/2
\]
wherein \( h_0 \) difference, between a thickness \( h_0 \) of the plate on the inlet side and its thickness \( h_1 \) on the discharging side, for each stand (each stage, or each pass upon rough rolling), is divided by an average thickness of the both thicknesses.

[0090] As used herein, the "accumulated strain" means a value expressed by \( \varepsilon_{\text{acc}} \), in the following equation:
\[
\varepsilon_{\text{acc}} = \sqrt{\varepsilon_1 + \varepsilon_2 + \varepsilon_3}/3
\]
wherein strain of each stage (each pass) of the rear-finish three stands is calculated by using a weighted estimation, considering the strength of each effect to be imposed on the metal phase, and wherein the strain to be generated on a final stage (final pass), front-stage (pre-pass), and pre-front-stage (pre-front-pass) is each expressed by \( \varepsilon_1 \), \( \varepsilon_2 \), and \( \varepsilon_3 \).

[0091] In order to carry out high-temperature finish rolling, a temperature rising process for the steel plate by utilizing heat generated by working due to the rolling is employed. To this end, it is important to provide a schedule of high-strain rolling for each rear-stage rolling mill as well as to set the reduction ratio of the front-stage stand at 40% or higher. As shown in FIG. 5, it can be seen that the surface temperature after finishing will vary, by 80°C, depending on the type of steel, in the case of using the same rolling size with respect to the reduction ratio while there is a difference in exit thicknesses at roughing mill.

[0092] Hot rolling is completed at a temperature of 950°C or higher, and the material is then subjected to air-cooling for 2 to 6 seconds without undergoing the post-hot-rolling hot run cooling, so as to reduce the dislocation density in the crystal particles. In FIG. 6, changes in the austenite particle size and changes in the dislocation density are illustrated, wherein these data are obtained by calculation over a period from a finishing F1 rolling mill to starting the hot run cooling, in the case of changing the rolling temperatures for the same type of steel. From the drawing, it can be seen that the dislocation density is significantly influenced by the rolling temperature. It can also be seen that under high-pressure rolling conditions, the austenite particle size will be smaller under lower temperature conditions, provided that the processing temperature is equal to or higher than that required for AR3 transformation. However, under such lower temperature conditions, the dislocation density will be higher, thus providing a material with further increased anisotropy. Additionally, it can be seen that while the dislocation density is significantly reduced due to the hot run air-cooling after rolling, effective results can be obtained by employing the cooling time within six seconds. In this simulation model, setting the aspect ratio at 2.0 or less can be translated into controlling the dislocation density to be at least 2.50E +10 (\( \mu \text{m}^2 \)) or less (this was confirmed from the results of comparison between actual data and the simulation model). However, the reduction of the dislocation density leads to increase of the size of the prior austenite particle. In order to further reduce the above-described numerical value of the dislocation density while controlling the prior austenite particle size at 10 \( \mu \text{m} \) or less, the aforementioned rolling conditions (rolling temperature: 950°C or higher, and cooling time: 2 to 6 seconds) are required.

[0093] The simulation model described above is based on Yanagimoto, Morimoto, et al., "Iron and Steel", vol. 88 (2002), No. 11, and each coefficient of the numerical formulae was reviewed for this application.

[0094] While the rolling temperature was set at a value within the range of from 550°C to 650°C, the temperature range lower than 550°C may tend to increase the martensitic phase, thus increasing the possibility of delayed fracture. On the other hand, the temperature range higher than 650°C will generate more ferrite and pearlite, as such making it difficult to obtain higher strength. In FIG. 7, sectional phases of three types of high-strength steel plates are shown.

[0095] Either phase shown in FIG. 7 is bainite based, in which a photograph corresponding to FIG. 7(a) shows a martensite-rich structure, FIG. 7(b) shows a lower-martensite and relatively fine structure, and FIG. 7(c) shows a ferrite-containing structure. FIG. 7(b) shows a structure obtained according to the present invention.

[0096] In the bainite-based structure, austenite is retained in each interface between the prior austenite particles as well as in each packet boundary and each block boundary, i.e., in the prior austenite particles themselves. The retained austenite can be closely and uniformly dispersed into a parent phase such that seven or more of the retained austenite particles having a very fine particle size, such as 1 \( \mu \text{m} \) or less, are dispersed per 10 \( \mu \text{m}^2 \), by employing the bainite phase as the parent phase and setting the size of prior austenite particles before transformation at 10 \( \mu \text{m} \) or less. FIG. 8 is a photograph of structures of sections each obtained by an EBSP method, for the steel plate according to the present invention, in which the bainite phase having a body-centered cubic phase and the austenite phase having a face-centered cubic phase are discriminated by colors. The retained austenite phase shown by a bright color constitutes a structure in which seven or more of the retained austenite particles having a particle size of 1 \( \mu \text{m} \) or less are finely and uniformly dispersed per 10 \( \mu \text{m}^2 \).

[0097] Due to such control for the hot rolling, the bainite phase can be obtained, in which the retained austenite particles are finely and uniformly dispersed.

[0098] If rolling a high-strength thin plate material (having a thickness of 2 mm or less) under high-strain and high-reduction-ratio conditions, a biting failure at a plate top por-
tion and/or a slip between a roll and a rolled material during the rolling operation is likely to occur. It has been found that from the rolling results, binding properties at a topmost portion of the rolling material is not problematic in the case of using a material of TS less than 1000 MPa at the reduction ratio of 40 to 50% per each rolling mill. On the other hand, the binding failure at the topmost portion of the rolling material will be likely to frequently occur (rate of occurrence: 50%), at the final rolling mill and the first to second rolling mills of the front-stage rolling mills, if using a material of TS greater than 1000 MPa. As a measure for addressing this problem, we have attempted to elevate the roll grinding finish roughness Ra up to 1 μm (ordinarily 0.5 μm) in order to raise the coefficient of roll friction, so as to obtain the coefficient of friction (μ) during rolling of 0.4 (ordinarily 0.3). In addition, we have reduced the flow amount of the roll cooling water in order not to unduly decrease the temperature at the topmost portion of the rolling material. However, securely effective results could not be obtained. Accordingly, as shown in FIG. 9, we have attempted to render the topmost portion of the rolling material thinner, over a place within the range of 5 m from the discharging side of the rolling mill, so as to make a thinner plate thickness (by 10% of a thickness finally expected). Therefore, an inclination up to the expected plate thickness was provided to the plate material.

0099 As a result, the binding failure was drastically decreased (rate of occurrence: 0%). In addition, the setting reduction amount employed in a range from the finish front-stage rolling mills to the rolling mills located before the finish final rolling mill was set at a value to be obtained by adding 10% or less of an expected set value thereto. The reduction setting time is set within two seconds from a biting site of the topmost portion of the plate material into the rolling mill.

0100 With respect to the slip between the rolling mill and the rolled material during a rolling process, if rolling a material of TS greater than 1000 MPa, under high temperature and high pressure conditions, with the final plate thickness being set less than 2 mm, slip is likely to occur at the final rolling mill and the rolling mill located just before the final rolling mill. As a phenomenon of this situation, a metallic sound is generated during the rolling process, the rolling load of the rolling mill upon occurrence of the slip is drastically showing a state wherein the rolled material and the roll are likely to slip on each other. To address such a situation, we have employed special high-grip rolls. As a result, the occurrence of the slip could be avoided completely. The rolls are each formed by uniformly dispersing micro-carbide particles (particle size: less than 1 μm) over the whole surface of the roll. These carbide particles can be utilized as spikes and supported by a hard base material. In addition, even through the micro-carbide particles will be worn away from the surface, micro-oxide particles will successively appear from below, thus maintaining a stable coefficient of friction, thereby to prevent the occurrence of slip. As shown in FIG. 10, changes of the coefficient of friction due to the rolling process is maintained in a more suitable range (approximately 0.3) as compared with commonly known rolls.

0101 FIG. 11 is a graph showing a relationship between the volume fraction (Vγ) of the retained austenite in the heat rolled steel plate produced by the manufacturing process shown in FIG. 1 and data obtained by the tensile test. FIG. 11(a) shows a relationship between the volume fraction Vγ and (the tensile strength×elongation). FIG. 11(b) shows a relationship between the volume fraction Vγ and the elongation. As is seen from the drawing, in the range of 5 to 20% of the volume fraction of the retained austenite, as the volume fraction Vγ is increased, the data of the tensile strength×elongation and the elongation are improved. The metallic phase corresponding to the data can be considered as the lower-martensite fine bainite phase as shown in FIG. 7(b).

0102 The present invention was made on the above empirical basis.

EXAMPLES

0103 Hereinafter, examples of the present invention will be described.

0104 Slab materials (rolling materials) were prepared from melted steel having each chemical composition shown in Table 1 by using a forging method or continuous casting method. Subsequently, these slab materials were heated again, and subjected to hot rolling, so as to obtain hot rolled steel plates, respectively. Table 2 shows respective conditions of the hot rolling and properties of the materials.

| TABLE 1 | Chemical composition (%) |
|---|---|---|---|---|---|---|---|---|---|---|---|---|---|
| Classes/Types | C | Si | Mn | P | S | Cu | Ni | Cr | Mo | (ppm) |
| Developed Steel A | 0.180 | 0.51 | 0.72 | 0.008 | 0.002 | 0.09 | 0.51 | 2.63 | 0.35 | 104 |
| Steel B | 0.180 | 1.00 | 0.40 | 0.009 | 0.001 | 0.07 | 0.16 | 2.97 | 0.33 | 137 |
| Steel C | 0.182 | 1.44 | 0.33 | 0.008 | 0.001 | 0.08 | 0.11 | 2.99 | 0.31 | 101 |
| Comparative Steel D | 0.188 | 0.26 | 0.42 | 0.008 | 0.002 | 0.11 | 3.87 | 1.96 | 0.65 | 123 |
| Steel E | 0.180 | 0.24 | 0.76 | 0.005 | 0.011 | 0.30 | 1.03 | 2.22 | 0.46 | 67 |
| Steel F | 0.097 | 1.45 | 0.41 | 0.002 | 0.005 | 0.11 | 0.16 | 2.95 | 0.29 | 79 |
| Steel G | 0.120 | 2.00 | 0.36 | 0.002 | 0.005 | 0.11 | 0.17 | 3.02 | 0.31 | 69 |
| Steel H | 0.199 | 1.97 | 0.41 | 0.010 | 0.005 | 0.11 | 0.17 | 4.42 | 0.31 | 166 |
| Steel I | 0.241 | 1.53 | 0.40 | 0.010 | 0.005 | 0.11 | 0.16 | 2.97 | 0.31 | 82 |

0105 With respect to steel types shown in Table 1, A, B, C designate steel plates prepared in accordance with the present invention, while D, E, F, G, H are provided as comparative examples.
The steel type D as one comparative example contains significantly lower Si and is excessively rich in Ni, thus departing from the preferred range of the present invention.

The steel type F contains significantly lower Si, thus also departing from the range defined according to the present invention.

The steel types F and G contain lower C, as such departing from the preferred range of the present invention, and the steel type H exhibits an unduly high content of C, thus also departing from the desired range of the present invention. The steel type H is excessively rich in Cr, as such departing from the preferred range of the present invention.

<table>
<thead>
<tr>
<th>Steel No.</th>
<th>Steel Type</th>
<th>%Si</th>
<th>%C</th>
<th>%Fe</th>
<th>Delayed fracture</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>A</td>
<td>0.85</td>
<td>0.15</td>
<td>98.5</td>
<td></td>
</tr>
<tr>
<td>2</td>
<td>A</td>
<td>0.80</td>
<td>0.16</td>
<td>99.5</td>
<td></td>
</tr>
<tr>
<td>3</td>
<td>B</td>
<td>0.59</td>
<td>0.24</td>
<td>97.5</td>
<td></td>
</tr>
<tr>
<td>4</td>
<td>B</td>
<td>0.45</td>
<td>0.25</td>
<td>96.0</td>
<td></td>
</tr>
<tr>
<td>5</td>
<td>B</td>
<td>0.45</td>
<td>0.25</td>
<td>96.0</td>
<td></td>
</tr>
<tr>
<td>6</td>
<td>C</td>
<td>0.84</td>
<td>0.16</td>
<td>97.0</td>
<td></td>
</tr>
<tr>
<td>7</td>
<td>D</td>
<td>0.92</td>
<td>0.20</td>
<td>96.0</td>
<td></td>
</tr>
<tr>
<td>8</td>
<td>E</td>
<td>0.67</td>
<td>0.35</td>
<td>95.5</td>
<td></td>
</tr>
<tr>
<td>9</td>
<td>F</td>
<td>0.67</td>
<td>0.35</td>
<td>95.5</td>
<td></td>
</tr>
<tr>
<td>10</td>
<td>G</td>
<td>0.67</td>
<td>0.35</td>
<td>95.5</td>
<td></td>
</tr>
<tr>
<td>11</td>
<td>H</td>
<td>0.67</td>
<td>0.35</td>
<td>95.5</td>
<td></td>
</tr>
<tr>
<td>12</td>
<td>I</td>
<td>0.67</td>
<td>0.35</td>
<td>95.5</td>
<td></td>
</tr>
</tbody>
</table>

These examples exhibit negative delayed fracture while satisfying the strength and the elongation.

No. 6 was prepared by using the steel type C containing 1.44% Si and by employing the hot rolling coiling temperature of 610°C. This example exhibits excellent properties in both of the strength and the elongation, and is superior to the No. 3 example in both of the strength and the elongation.

Nos. 7 to 12 were respectively prepared by carrying out hot rolling, using steel types of comparative examples departing from the desired range of the composition used in the present invention.

No. 7 was prepared by rolling, using the steel type D containing lower Si and higher Ni. This comparative example is insufficient in the spot welding properties (S/W properties) as well as in the delayed fracture properties.

No. 8 was prepared by using the steel type E containing lower Si, thus exhibiting insufficient strength and poor balance of strength/ductility.

No. 9 and No. 10 were prepared by using the steel types F and G both containing lower C, respectively, as such exhibiting unduly lower strength and poor balance of strength/ductility. However, these comparative examples are insufficient in the spot welding properties as well as in delayed fracture properties.

The volume fraction of the ferrite particles was measured by observation using an optical microscope, after polishing a section cut along the rolling direction of each steel plate and then subjecting the polished surface to nital corrosion. The measurement also used a commercially available image analyzer.
The volume fraction of the martensite was obtained by measuring the martensite phase expressed by a white color in an image analysis process during observation using an optical microscope for a position directed to 1/4 of the plate thickness direction, after polishing a section cut along the rolling direction of each steel plate and then etching the polished surface by using a liquid formed by mixing 1:1 of 4% picric acid-alcohol and 2% sodium pyrophosphate.

The measurement of the retained austenite was carried out by employing the X-ray diffraction by using Kα ray of Cu. The volume fraction was determined as an average of the volume fraction of the retained austenite to be calculated from a combination of data obtained by respectively measuring integrated intensities of (200), (220) and (311) faces of the austenite phase and those of (200), (211) faces of the ferrite phase, after electrolytic polishing for a position directed to 1/4 of the plate thickness direction.

The tensile properties (tensile strength (TS) and elongation (EL)) were measured by subjecting each sample to a tensile test, the sample being formed into the shape in accordance with the JIS No. 5 test piece.

The delayed fracture properties were assessed by observation of each sample dipped in a 1N hydrochloric acid solution for a predetermined period of time, after forming 10 mm punch holes with a 12.2% clearance in a central portion subjected to the tensile test, onto which 8% or more of strain had been loaded.

As described above, the high-strength steel plates obtained by the examples, which exhibit high strength and high ductility properties in a lower alloy composition are suitable for use as components for manufacturing car structures.

For example, the high-strength steel plates according to the present invention can be used as quite preferred materials, such as center pillars for cars, which require highly excellent properties, including sufficient tensile strength for supporting doors and preventing deformation upon collision or the like, bendability for press molding, deep drawability, hole extending workability for forming an attachment hole to be used for associated equipment, and weldability for welding the material to another car component.

Although the invention has been described in its preferred embodiments with a certain degree of particularity, obviously many changes and variations are possible therein. It is therefore to be understood that the present invention may be practiced otherwise than as specifically described herein without departing from the scope and spirit thereof.

What is claimed is:

1. A high-strength hot rolled steel plate comprising: a retained martensite phase in a volume fraction of 5% to 20%; a martensite phase in a volume fraction of 0% to 10%; and a bainite phase in a remaining volume fraction, wherein a particle size of retained austenite particles is 1 μm or less, and the retained austenite particles are dispersed in a density of seven or more particles per 10 μm².

2. The high-strength hot rolled steel plate according to claim 1, wherein a particle size of prior austenite particles is 10 μm or less, and an average aspect ratio of the prior austenite particles is 2.0 or less.

3. The high-strength hot rolled steel plate according to claim 1, wherein the steel plate has a composition comprising: C (0.13 to 0.21 (% by weight)), Si (0.5 to 2.0), Mn (0.2 to 1.0), Cr (1.0 to 4.0), Ni (0.02 to 1.0), Mo (0.05 to 0.4), P (0 to 0.010), S (0 to 0.003), N (0.005 to 0.015), and remaining components including Fe and other inevitable impurities.

4. The high-strength hot rolled steel plate according to claim 1, the steel plate has a plate thickness of 1.0 to 3.0 mm and a tensile strength of 1200 MPa or greater.

5. A method for manufacturing a high-strength hot rolled steel plate, comprising the steps of:
   preparing a steel material having a composition containing: C (0.13 to 0.21 (% by weight)), Si (0.5 to 2.0), Mn (0.2 to 1.0), Cr (1.0 to 4.0), Ni (0.02 to 1.0), Mo (0.05 to 0.4), P (0 to 0.010), S (0 to 0.003), N (0.005 to 0.015), and remaining components including Fe and other inevitable impurities;
   roughly rolling the steel material under conditions of: 1250°C or higher of an extraction temperature of a reheating furnace; 1030°C or higher of a discharging-side temperature of roughly rolling mills; and 50% or higher of a reduction ratio for each of roughly rolling final three passes;
   finish rolling of the steel material under conditions of: 950°C or higher of a discharging-side temperature of finish rolling mills; 40% or higher of a reduction ratio for each mill on a finish front-stage, and 0.5 or greater of accumulated strain due to a reduction by three rolling mills on a finish rear-stage; and
   cooling the steel material by air-cooling for 2 to 6 seconds, followed by water-cooling, and cooling the steel material at a cooling temperature of 550 to 650°C.

6. The method for manufacturing a high-strength hot rolled steel plate according to claim 5, wherein upon the finish rolling, a reduction amount of a topmost portion of the steel material is set greater than an expected reduction amount, in one or more rolling mills other than a mill on a final stage, wherein the reduction amount of the top portion is set at a value increased by less than 10% of the expected reduction amount of the rolling mill, and wherein a length to be rolled in an increased reduction amount is within 5 m as measured from a biting site of the topmost portion of the steel material, and thereafter the reduction amount is returned to the expected reduction amount.

7. The method for manufacturing a high-strength hot rolled steel plate according to claim 5, wherein a high-grip roll having micro-carbide particles dispersed on a surface of the high-grip roll is used as a working roll for each finish rear-stage rolling mill including a final rolling mill.

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