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(54) LEAN DUPLEX STAINLESS STEEL AND METHOD FOR PRODUCING SAME

(57) The present invention relates to lean duplex stainless steel having a dual-phase structure of an austenite phase and a ferrite phase, and a method for producing the lean duplex stainless steel, the lean duplex stainless steel according to one embodiment of the present invention, as a ferrite-austenite stainless steel, having the preferred stacking fault energy (SFE) value of the austenite phase, expressed by the formula 2 below, of 19-37 and critical strain value range, within which the strain-induced martensite phases occurs, of 0.1-0.25.

$$\text{Formula 2: } SFE = 25.7+1.59\times Ni/[K(Ni)-K(Ni)\times V(\gamma)+V(\gamma)]+0.795\times Cu/[K(Cu)-K(Cu)\times V(\gamma)+V(\gamma)]-0.85\times Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)]+0.001\times(Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)])^2+38.2\times(N/[K(N)-K(N)\times V(\gamma)+V(\gamma)])^{0.5}-2.8\times Si/[K(Si)-K(Si)\times V(\gamma)+V(\gamma)]-1.34\times Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)]+0.06\times(Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)])^2$$

where Ni, Cu, Cr, N, Si and Mn indicate the overall content (wt.%) of the respective constituent element, and K(x) is the distribution index of respective constituent element (x) and is expressed by the formula 3 below, and $V(\gamma)$ is the component ratio of austenite (in the 0.45-0.75 range).

$$\text{Formula 3: } K(x) = [\text{amount of element x in ferrite phase}]/[\text{ amount of element x in austenite phase}]$$

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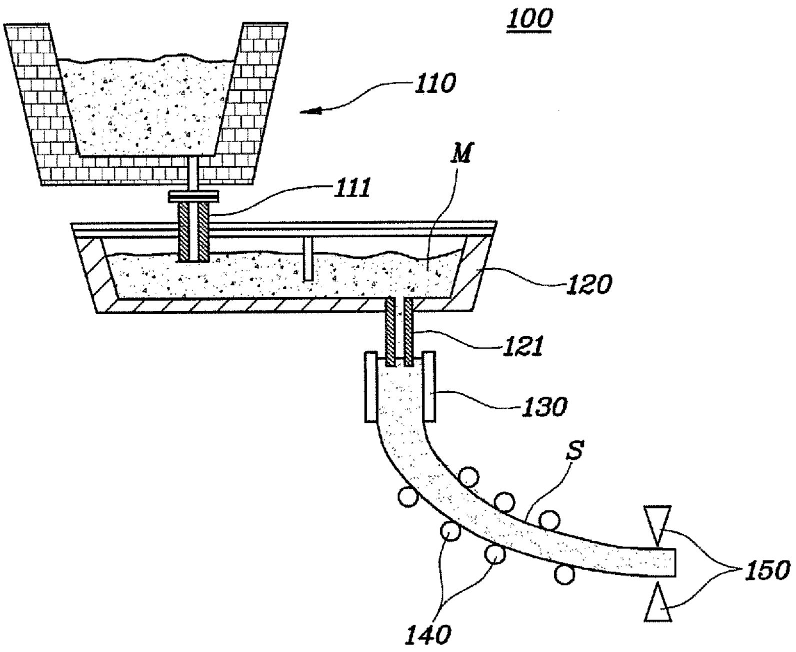
[FIG.4]



Description

TECHNICAL FIELD

[0001] The present invention relates to lean duplex stainless steel having a ferritic-austenitic structure and a method for producing the same.

BACKGROUND ART

[0002] In general, the austenitic stainless steel with good workability and corrosion resistance contains iron (Fe) as a base metal and chromium (Cr) and nickel (Ni) as main raw materials, and has been developed as a variety of steel to meet various applications by adding other elements such as molybdenum (Mo) and copper (Cu).

[0003] 300 series stainless steel, which is excellent in corrosion resistance and workability, contains expensive raw materials such as Ni and Mo, and thus, as an alternative to this, 200- and 400-series stainless steels have been discussed. However, 200-series and 400-series stainless steels have disadvantages respectively in that their formability and corrosion resistance do not reach to 300-series stainless steel.

[0004] On the other hand, the duplex stainless steel in which an austenite phase and a ferrite phase are mixed has all the advantages of the austenitic stainless steel and the ferritic stainless steel, and thus, to date, various types of duplex stainless steels have been developed.

[0005] In addition, in recent years, in order to compensate for the shortcoming of price competitiveness, there is an increasing interest in lean duplex stainless steels, in which high-cost alloying elements such as Ni and Mo contained in the duplex stainless steels are eliminated and low-cost alloying elements are added to further enhance cost advantages.

[0006] However, such lean duplex stainless steels have a disadvantage in that they have a poor hot workability depending on the difference in strength between the ferrite phase and the austenite phase, resulting in a large number of surface cracks and edge cracks.

[0007] In addition, if alloy elements whose components were adjusted for lean duplex stainless steel are processed according to a conventional and general continuous casting process, a large number of the pores are generated inside the slab due to the difference in nitrogen solubility when solidified from a liquid phase to a solid phase. These internal pores are the cause of generating a large number of defects on the surface of the product during the reheating and hot rolling processes which are subsequent processes, and also causes the disadvantage of generating a large number of cracks on the edge of the hot rolled coil.

[0008] The foregoing description of the background art is merely for the purpose of promoting an understanding of the background of the present invention, and should not be accepted as an admission that it corresponds to the prior art known to a person having ordinary skill in the art.

DISCLOSURE

TECHNICAL PROBLEM

[0009] The present invention provides lean duplex stainless steel capable of ensuring excellent elongation and corrosion resistance and a method of manufacturing the same, by controlling the content of alloy components to reduce costs and also satisfactorily control the stacking fault energy value present in the lean-duplex stainless steel,

[0010] Also, the present invention provides a lean duplex stainless steel capable of ensuring excellent elongation and corrosion resistance and a method of manufacturing the same, by satisfactorily controlling the value of the critical strain for strain induced martensite formation.

[0011] Further, the present invention provides a lean duplex stainless steel capable of solving the problem of massive release of nitrogen gas due to a sharp reduction in nitrogen solubility during solidification from a liquid phase to a solid phase in casting, and a method of manufacturing the same.

TECHNICAL SOLUTION

[0012] The lean duplex stainless steel according to one embodiment of the present invention is a ferrite-austenitic stainless steel wherein it is preferable that the stacking fault energy (SFE) value of the austenite phase represented by the following formula 2 is 19 to 37 and the range of the value of the critical strain for strain induced martensite formation is 0.1 to 0.25.

$$SFE = 25.7+1.59\times Ni/[K(Ni)-K(Ni)\times V(\gamma)+V(\gamma)]+0.795\times Cu/[K(Cu)-K(Cu)\times V(\gamma)+V(\gamma)]-0.85\times Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)]+0.001\times(Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)])^2+38.2\times(N/[K(N)-K(N)\times V(\gamma)+V(\gamma)])^{0.5}-2.8\times Si/[K(Si)-K(Si)\times V(\gamma)+V(\gamma)]-1.34\times Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)]+0.06\times(Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)])^2 \text{ ....Formula 2}$$

wherein Ni, Cu, Cr, N, Si and Mn refer to the overall content (wt.%) of respective constituent element respectively, and K(x) is represented by the following formula 3 as the distribution coefficient of respective constituent element (x), and V(γ) is the fraction of the austenite phase (the range of 0.45 to 0.75).

$$K(x) = [\text{content of element x in ferrite phase}] / [\text{content of element x in austenite phase}] \text{ .... Formula 3}$$

wherein regarding the K(x), K(Cr) = 1.16, K(Ni) = 0.57, K(Mn) = 0.73, and K(Cu) = 0.64, and K(N) and K(Si) may have the following values depending on the content (wt.%) of N and Si:

when N is 0.2 to 0.32%, K (N) = 0.15;
when N < 0.2%, K(N) = 0.25;
when Si ≤ 1.5%, K(Si) = 2.76-0.96×Si; and
when Si > 1.5%, K(Si) = 1.4.

[0013] It is preferable that the elongation of the stainless steel is 45% or more.

[0014] It is preferable that the stainless steel comprises, by weight, C: 0.08% or less (excluding 0%), Si: 0.2 to 3.0%, Mn: 2 to 4%, Cr: 18 to 24%, Ni: 0.2 to 2.5%, Cu: 0.2 to 2.5%, balance Fe and the other unavoidable impurities.

[0015] The stainless steel may further comprise, by weight, at least one of W: 0.1 to 1.0% and Mo: 0.1 to 1.0%.

[0016] The stainless steel may further include, by weight, at least one of Ti: 0.001 to 0.1%, Nb: 0.001 to 0.05%, and V: 0.001 to 0.15%.

[0017] Meanwhile, the method of manufacturing the lean duplex stainless steel according to an embodiment of the present invention is a method of manufacturing a ferritic-austenitic lean duplex stainless steel, comprising preparing a molten steel; and treating the molten steel to form the stainless steel wherein the molten steel is treated so that the stacking fault energy (SFE) value of the austenite phase represented by the following formula 2 is 19 to 37 and the range of the value of the critical strain for strain induced martensite formation is 0.1 to 0.25:

$$SFE = 25.7+1.59\times Ni/[K(Ni)-K(Ni)\times V(\gamma)+V(\gamma)]+0.795\times Cu/[K(Cu)-K(Cu)\times V(\gamma)+V(\gamma)]-0.85\times Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)]+0.001\times(Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)])^2+38.2\times(N/[K(N)-K(N)\times V(\gamma)+V(\gamma)])^{0.5}-2.8\times Si/[K(Si)-K(Si)\times V(\gamma)+V(\gamma)]-1.34\times Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)]+0.06\times(Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)])^2 \text{ ....Formula 2}$$

wherein Ni, Cu, Cr, N, Si and Mn refer to the overall content (wt.%) of respective constituent element respectively, and K(x) is represented by the following formula 3 as the distribution coefficient of respective constituent element (x), and V(γ) is the fraction of the austenite phase (the range of 0.45 to 0.75):

$$K(x) = [\text{content of element x in ferrite phase}]/[\text{content of element x in austenite phase}] \text{ ....Formula 3.}$$

[0018] In particular, the process of treating the molten steel to form the stainless steel may comprise temporarily storing the molten steel in the tundish while maintaining the temperature of the molten steel at the temperature higher than the theoretical solidification temperature by 10 to 50 °C; primarily cooling the molten steel by injecting the molten steel in the tundish into the mold and passing the molten steel through the mold while maintaining a cooling rate of 500 to 1500 °C/min; and secondarily cooling the molten steel having the solidified shell formed by the primary cooling process while drawing it into a segment and passing through.

[0019] At this time, in the secondary cooling process, it is preferable to spray the cooling water of 0.25 to 0.35 L/kg on the molten steel having the formed solidified shell.

[0020] Further, the method may further comprise, tertiarily cooling, after the secondary cooling process, by spraying the cooling water of 100 to 125 L/kg · min on the surface of the cast-slab in the range of the surface temperature of the cast-slab being drawn of 1100 to 1200 °C wherein the cooling water is mixed with air such that the ratio of air to cooling water (air/cooling water) is 1.0 to 1.2

[0021] Meanwhile, the process of treating the molten steel to form the stainless steel may comprise producing a strip by solidifying the molten steel while passing it between a pair of casting rolls wherein nitrogen contained in the molten steel in the process of producing the strip and exceeding a nitrogen solubility limit can be discharged through the casting roll to the outside of the solidified shell.

[0022] At this time, in the process of producing the strip, it is preferable to use a casting roll, in which a gas discharge channel is formed in the circumferential direction on the outer peripheral surface, as at least one of a pair of casting rolls

[0023] In addition, the gas discharge channel formed in the casting roll used in the process of producing the strip has a width of 50 to 500 μm and a depth of 50 to 300 μm, a plurality of gas discharge channels is formed in the casting roll, the gap between adjacent gas discharge channels is 100 to 1000 μm and unevennesses of 15 to 25 μm are formed on the surface of the casting roll.

[0024] Also, in the process of preparing the molten steel, the molten steel comprises, by weight, C: 0.08% or less (excluding 0%), Si: 0.2 to 3.0%, Mn: 2 to 4%, Cr: 18 to 24%, Ni: 0.2 to 2.5%, N: 0.15 to 0.32%, Cu: 0.2 to 2.5%, balance Fe and the other unavoidable impurities.

[0025] Also, in the process of preparing the molten steel, the molten steel may further comprise, by weight, at least one of W: 0.1 to 1.0% and Mo: 0.1 to 1.0%.

[0026] Also, in the process of preparing the molten steel, the molten steel may further comprise, by weight, at least one of Ti: 0.001 to 0.1%, Nb: 0.001 to 0.05%, and V: 0.001 to 0.15%.

ADVANTAGEOUS EFFECTS

[0027] According to the embodiment of the present invention, it is possible to save resources and significantly reduce the cost of raw materials by controlling the contents of the alloy components of Ni, Si and Cu which are high-priced elements, and in particular, it is sufficiently usable as an alternative to the 200 and 300 series (STS 304, 316) used for molding, by ensuring corrosion resistance and excellent elongation at the same level or higher compared to STS 304.

[0028] Further, in the case of continuous casting of the alloying elements according to the embodiments of the present invention, the temperature of the molten steel and the cooling rate can be controlled to suppress the pinholes generated inside the cast-slab.

[0029] Further, in the case of strip casting of the alloying element according to the embodiment of the present invention, the generation of the internal pores and the occurrence of surface defects can be prevented by smoothly inducing the discharge of the nitrogen gas generated when solidified from a liquid phase to a solid phase, through the improvement of the casting roll.

BRIEF DESCRIPTION OF THE DRAWINGS

[0030]

FIG. 1 is a view showing a value of critical strain for strain induced martensite phase formation as stress-strain curves of an inventive steel according to an embodiment of the present invention and a comparative steel.

FIGs. 2a and 2b are photographs showing the formation of strain induced martensite phases as representative transmission electron microscope photographs of a inventive steel according to an embodiment of the present invention and a comparative steel.

FIG. 3 is a graph showing the relationship between the elongation and the value of critical strain for strain induced martensite phase formation.

FIG. 4 is a schematic view illustrating a manufacturing process of continuous casting method of the lean duplex stainless steel according to an embodiment of the present invention.

FIG. 5 is a schematic view illustrating a manufacturing process of strip casting method of the lean duplex stainless steel according to an embodiment of the present invention.

FIG. 6 is a schematic view illustrating a casting roll required in a manufacturing process of strip casting method of the lean-duplex stainless steel according to an embodiment of the present invention.

FIG. 7 is a photograph of the structure of the comparative material I and the inventive material A.

FIG. 8 is a photograph of surface defects of the comparative material H.

FIG. 9 is a photograph of surface defects of the comparative material F.

BEST MODE FOR THE INVENTION

[0031] Hereinafter, embodiments of the present invention will be described in more detail with reference to the accompanying drawings. However, the present invention is not limited to the embodiments described below, but may be embodied in many different forms. Rather, these embodiments are provided so that the disclosure of the invention will be thorough and complete, and will fully convey the scope of the invention to those skilled in the art.

[0032] First, the present invention relates to lean duplex stainless steel having ferritic-austenitic structures wherein the ferritic-austenitic structures referred to in the present invention means that ferrite phase and austenite phase occupy most of the structures, and does not means that the stainless steel is formed only in ferrite phase and austenite phase. For example, the description that the ferrite phase and the austenite phase occupy most of the structures means that the sum of the ferrite phase and the austenite phase in the structures for forming stainless steel accounts for 90% or more, and the remainder, except for the ferrite phase and the austenite phase, may be occupied by the austenite phase transformed martensite phase.

[0033] FIG. 1 is a view showing a value of critical strain for strain induced martensite formation as stress-strain curves of the inventive steel according to an embodiment of the present invention and the comparative steel.

[0034] The present invention relates to lean duplex stainless steel having two-phase structures formed of a ferrite phase and an austenite phase comprising, by weight (unless otherwise specified below, the content of the ingredients is % by weight), C: 0.08% or less (excluding 0%), Si: 0.2 to 3.0%, Mn: 2 to 4%, Cr: 18 to 24%, Ni: 0.2 to 2.5%, N: 0.15 to 0.32 %, Cu: 0.2 to 2.5%, balance Fe, and the other unavoidable impurities. The lean duplex stainless steel may further comprise by weight at least one of W: 0.1 to 1.0%, Mo: 0.1 to 1.0%, Ti: 0.001 to 0.1%, Nb: 0.001 to 0.05%, and V: 0.001 to 0.15%.

[0035] C is an element for forming an austenite phase and is an effective element for increasing the strength of a material by solid solution strengthening. However, since C, when added excessively, easily bonds with carbide-forming elements such as Cr, which is effective for corrosion resistance at the ferrite-austenite phase boundary, thereby reducing the Cr content around the grain boundary and thus reducing the corrosion resistance, C is preferably added in a range of more than 0% to 0.08% or less in order to maximize the corrosion resistance.

[0036] Si is partially added for the deoxidation effect, is an element for forming a ferrite phase and is an element which is concentrated on the ferrite phase during the annealing heat treatment. Therefore, 0.2% or more should be added to ensure a proper ferrite phase fraction. However, the excessive addition of Si exceeding 3.0% increases the hardness of the ferrite phase drastically and thus affects the lowering of the elongation of the two-phase steel, and also makes it difficult to ensure the austenite phase for ensuring sufficient elongation. Further, Si, when being added excessively, lowers the slag fluidity at the time of steelmaking and bonds with oxygen to form inclusions, thereby deteriorating the corrosion resistance. Therefore, the content of Si is preferably limited to 0.2 to 3.0%.

[0037] Mn is an element that increases the deoxidizing agent and nitrogen solubility, and is an austenite phase forming element and can be used for replacing expensive Ni. When Mn is added in a large amount, excessive Mn is effective for solubility of nitrogen, but combines with S in the steel to form MnS, thereby deteriorating the corrosion resistance. Therefore, when its content exceeds 4%, it becomes difficult to ensure the corrosion resistance of the level of 304 steel. In addition, when the content of Mn is less than 2%, even if Ni, Cu, N and the like which are the austenite phase forming elements are controlled, it is difficult to ensure a proper austenite phase fraction, and the solubility of N to be added is low and thus sufficient solution of nitrogen at normal pressure cannot be obtained. Therefore, the content of Mn is preferably limited to 2 to 4%.

[0038] Cr is an element for stabilizing the ferrite phase together with Si and plays a major role in ensuring the ferrite phase of the two-phase stainless steel, as well as it is an essential element for ensuring corrosion resistance. When the Cr content is increased, the corrosion resistance increases, but in order to maintain the phase fraction, the content of expensive Ni or other austenite phase forming elements must be increased. Accordingly, in order to ensure corrosion resistance equal to or higher than that of 304 steel while maintaining the phase fraction of the two-phase stainless steel, it is preferable to limit the content of Cr to 18 to 24%.

[0039] Ni is an element for stabilizing the austenite phase together with Mn, Cu and N, and plays a main role in ensuring the austenite phase of the duplex stainless steel. By increasing Mn and N, which are other austenite phase forming elements, instead of maximally decreasing the content of expensive Ni for cost reduction, it is possible to sufficiently maintain the phase fraction balance that can be influenced by the reduction of Ni. However, in order to suppress the formation of strain induced martensite phase which occurs during the cold processing and accordingly to ensure sufficient austenite phase stability, it is necessary to add at least 0.2%. When a large amount of Ni is added, it is difficult to ensure a proper austenite phase fraction because the fraction of the austenite phase increases and in particular, it is difficult to ensure competitiveness compared to 304 steel due to an increase in the manufacturing cost of products by expensive Ni. Therefore, the content of Ni is preferably limited to 0.2 to 2.5%.

[0040] N is an element that contributes greatly to the stabilization of the austenite phase together with Ni in the duplex stainless steel and is one of the elements which is mostly concentrated on the austenite phase due to the high diffusion rate on the solid phase during the annealing heat treatment. Therefore, the increase of the N content can incidentally induce an increase in corrosion resistance and an increase in strength. However, the solubility of N varies depending on the content of Mn added. When the N content exceeds 0.32% in the range of Mn of the present invention, it is difficult to stably produce steel due to generation of surface defects such as blow holes or pinholes during casting due to excessive nitrogen solubility. On the other hand, N is added in the amount of 0.15% or more in order to ensure the corrosion resistance of the level of 304 steel. If the content of N is too low, it becomes difficult to ensure a proper phase fraction. Therefore, the content of N is preferably limited to 0.15 to 0.32%.

[0041] Cu is an element for stabilizing the austenite phase together with Mn, Ni and N, and it is desirable that the content of Cu, which plays the same role as Ni, should be minimized for cost reduction. However, it is preferable to add at least 0.2% in order to ensure the stability of the austenite phase sufficient to suppress the excessive formation of the strain induced martensite phase occurring during the cold processing. On the other hand, if the content of Cu exceeds 2.5%, it becomes difficult to process the product due to hot brittleness. Therefore, the content of Cu is preferably adjusted to 0.2 to 2.5%.

[0042] W and Mo are elements for forming the austenite phase and elements for improving corrosion resistance, and are elements that promote the formation of an intermetallic compound at 700 to 1000 °C during heat treatment, resulting in deterioration of corrosion resistance and mechanical properties. When the content of W and Mo exceeds 1% respectively, an intermetallic compound is formed, which may lead to a rapid decrease in corrosion resistance and particularly in elongation. In addition, 0.1% or more can be added in order to exhibit the effect of improving the corrosion resistance. Therefore, the content of W and Mo is preferably limited to 0.1 to 1.0% respectively, and at least one of W and Mo may be contained.

[0043] Ti, Nb, and V are elements that react with nitrogen to form nitrides, and they are crystallized as TiN, NbN, and VN respectively in the molten steel and act as nucleation sites on the ferrite phase during solidification, so that sufficient solidification can proceed even when the cooling rate is increased, thereby suppressing the breakage of the slab. In addition, these elements are sufficiently dissolved during the manufacturing process, i.e., reheating or hot rolling, and react with carbon and nitrogen during cooling to form carbonitride and thus inhibit the formation of Cr carbide, thereby contributing to improvement of corrosion resistance. In particular, they are elements that inhibit the formation of Cr carbide in the heat affected zone during welding. When each of these elements is excessively added, that is, when Ti is more than 0.1%, Nb is more than 0.05%, and V is more than 0.15%, these crystals in the solidification form large clusters and cause clogging phenomenon that block the casting nozzle and further, when existing in the surface layer portion of the slab, they act as the cause of defects during rolling and breakage during processing. In addition, since most of these are expensive alloying elements, the addition of a large amount causes an increase in manufacturing cost. Therefore, it is preferable to limit to the ranges of Ti: 0.001 to 0.1%, Nb: 0.001 to 0.05%, and V: 0.001 to 0.15%, and at least one of Ti, Nb and V may be contained.

[0044] Meanwhile, the present invention maintains elongation and corrosion resistance excellently by controlling the content of the alloy elements, the distribution coefficient and the phase fraction to control the stacking fault energy

[0045] For example, the following formula 1 is a formula for deriving the stacking fault energy by utilizing the content of all components in the alloy.

$$\mathrm{SFE} = 25.7 + 1.59\,(\mathrm{Ni} + 0.5\mathrm{Cu}) - 0.85\mathrm{Cr} + 0.001\mathrm{Cr}^2 + 38.2\mathrm{N}^{0.5} - 2.8\mathrm{Si} - 1.34\mathrm{Mn} + 0.06\mathrm{Mn}^2 -$$

- Formula 1

wherein Cr, Ni, Cu, Si, Mn, and N mean the overall content (wt.%) of respective constituent elements.

[0046] However, the applicants of the present invention have found that as a result of measuring and calculating the stacking fault energies of the inventive steel by various methods, it is more accurate to predict the properties of the alloy by calculating the stacking fault energy while utilizing the content of the components of the austenite structure rather than calculating it using only the content of the components of the overall alloy composition as in formula 1. The applicants of the present invention have found that for this purpose, calculating the stacking fault energy taking into account the interstitial distribution coefficient of the alloying elements rather than calculating the stacking fault energy by using only the component content of the overall alloy composition can obtain a more approximate approximation of the actually measured stacking fault energy value.

[0047] Therefore, the applicants of the present invention have made the following formula 2 by supplementing formula 1 above so as to deduce the stacking fault energy of the austenite phase by utilizing the distribution coefficient of the austenite phase.

$$\begin{aligned} SFE = {} & 25.7+1.59\times Ni/[K(Ni)-K(Ni)\times V(\gamma)+V(\gamma)]+0.795\times Cu/[K(Cu)-K(Cu)\times V(\gamma)+V(\gamma)]- \\ & 0.85\times Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)]+0.001\times(Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)])^{2}+38.2\times(N/[K(N)- \\ & K(N)\times V(\gamma)+V(\gamma)])^{0.5}-2.8\times Si/[K(Si)-K(Si)\times V(\gamma)+V(\gamma)]-1.34\times Mn/[K(Mn)- \\ & K(Mn)\times V(\gamma)+V(\gamma)]+0.06\times(Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)])^{2} \text{ ....Formula 2} \end{aligned}$$

wherein Ni, Cu, Cr, N, Si and Mn refer to the overall content (wt.%) of respective constituent elements, and K(x) is represented by the following Formula 3 as the distribution coefficient of respective constituent element (x).

$$K(x) = [\text{content of element x in ferrite phase}]/[\text{content of element x in austenite phase}] \text{ ....Formula 3.}$$

[0048] The applicants of the present invention measured the distribution coefficients for respective alloying elements in various annealing conditions and alloy systems by using Fe-EPMA and FE-TEM which are more accurate than EDAX analysis by conventional scanning electron microscopy. At this time, it was confirmed that the distribution coefficients for most of the measured alloying elements are not changed depending on the change of temperature at 900 to 1200 °C, which is the temperature range of hot rolling annealing or cold rolling annealing.

[0049] That is, regarding the K(x), K(Cr) = 1.16, K(Ni) = 0.57, K(Mn) = 0.73, and K(Cu) = 0.64, and it was confirmed that K(N) and K(Si) vary depending on the content of N and Si (wt.%). However, when N: 0.2 to 0.32%, K(N) = 0.15, when N < 0.2%, K(N) = 0.25, when Si ≤ 1.5%, K(Si) = 2.76-0.96×Si, and when Si > 1.5, K(Si) = 1.4. At this time, the alloying elements N and Si mean the entire components of the stainless steel. However, in the present embodiment, since N: 0.15 to 0.32%, when N: 0.15 or more and less than 0.2%, K (N) = 0.25 will be applied and when N: 0.2 to 0.32, K (N) = 0.15 will be applied, and since Si: 0.2 to 3.0%, when Si: 0.2 to 1.5%, K(Si) = 2.76-0.96×Si will be applied and when Si: more than 1.5% and 3.0% or less, K(Si) = 1.4 will be applied.

[0050] In addition, in the formula 2, V(γ) is the austenite phase fraction, and the austenite phase fraction is defined by the following relation function.

$$1 = V(\alpha) + V(\gamma)$$

wherein V(α) is the ferrite phase fraction, and V(γ) has the range of 0.45 to 0.75.

[0051] On the other hand, the reason for limiting the stacking fault energy value of the austenite phase to 19 to 37 is explained below.

[0052] The stacking fault energy of the austenite phase is known to control the transformation mechanism of the austenite phase. Typically, the stacking fault energy of the austenite phase in the case of single-phase austenitic stainless steel represents the extent to which the externally applied plastic strain energy contributes to the strain of the austenite phase. Generally, the lower the stacking fault energy, the greater the degree of formation of the strain induced martensite phase that contributes to the work hardening of the steel after formation of the epsilon martensite phase on the austenite phase. If the stacking fault energy is moderate, mechanical twin is formed in the austenite phase. In the case of moderate stacking fault energies, the strain induced martensite phase is formed at the intersection of these twinning, so that the applied plastic strain energy causes mechanical phase change and thus transformation from the austenite phase to martensite phase. Therefore, the stainless steel is known to form the strain induced martensite phase in a very broad range, except for difference in the mesophase (the epsilon martensite phase or the mechanical twin). Accordingly, when the stacking fault energy is less than 50 mJ/m^2, the strain induced martensite phase is formed after the epsilon martensite phase is formed in the austenite phase, or the strain induced martensite phase is formed after the mechanical twin is formed in the austenite phase.

[0053] However, it is known that when the stacking fault energy is 50 mJ/m^2 or more, since the strain proceeds by dislocation movement without formation of the mechanical twin or the epsilon martensite phase, transformation from the austenite phase to the martensite phase does not work well.

[0054] On the other hand, it was confirmed that when Formula 1 is applied to deduce the stacking fault energy only from the overall components of the alloy, formation of the strain induced martensite phase is easy when being at 11 or less and thus in the early stage of the strain, rapid work hardening is caused, that is, the strain induced martensite phase is formed, thereby resulting in a sharp decrease in elongation. However, it was confirmed that since the alloying elements distributed to the austenite phase vary depending on the heat treatment or the manufacturing process, the formation behavior of strain induced martensite phases varies in some component systems.

[0055] Therefore, the applicants of the present invention corrected the formula as in formula 2 above, taking into account the distribution coefficients of the alloying elements distributed to the austenite phase after various manufacturing processes and heat treatments. As a result, when the calculated stacking fault energy of the austenite phase is less than 19, the epsilon martensite phase is first formed as the medium phase and the martensite phase is formed at the intersection of the formed epsilon martensite phases. However, these martensite phases are rapidly formed at the beginning of the strain, and thus a phenomenon in that the elongation is lowered due to the rapid work hardening is observed. On the other hand, it was confirmed that when the stacking fault energy of the austenite phases calculated by using the corrected formula is more than 37, as a result of investigation with a transmission electron microscope, formation of the martensite phase is not observed after plastic strain. Therefore, it can be seen that the preferable range of the stacking fault energy of the austenite phases is 19 to 37.

[0056] Meanwhile, the lean duplex stainless steel according to the present invention is preferably formed as austenite phases of 45 to 75% and ferrite phases of 25 to 55% by volume fraction.

[0057] When the fraction of the austenite phases is less than 45%, an excessive concentration phenomenon of the austenite phase forming elements in the austenite phases occurs during the annealing. Thereby, the austenite phases are sufficiently stabilized to suppress transformation of the strain induced martensite phase which is generated during the strain, and the amount of the austenite phase due to sufficient solid solubility of the alloying elements is increased and thus the tensile strength of the material can be also sufficiently ensured. However, there is a phenomenon in which the ductility is lowered, and the desired sufficient elongation and strength cannot be obtained. Therefore, in view of the high ductility, the fraction of the austenite phase is preferably 45% or more.

[0058] However, when the fraction of the austenite phase exceeds 75%, surface cracks, etc. occur during hot rolling, thereby resulting in deterioration of the hot workability and loss of properties as two-phase structure steel. Therefore, the fraction of the austenite phase is preferably 75% or less.

[0059] On the other hand, in the case of the present invention, it is preferable that the range of the critical strain value for strain induced martensite phase format in during the cold processing or the tensile strain is maintained at 0.1 to 0.25.

[0060] The amount of critical strain for strain induced martensite phase formation was measured from the inflection point of the stress-strain curve, and this usually refers to the strain value at the time of contributing to the work hardening of the martensite phases in the steel in which the strain induced martensite phase is formed.

[0061] In other words, a method of obtaining a critical strain value for strain induced martensite phase formation is described below.

[0062] First, after the specimens is taken in accordance with ASTM sub-size standard in parallel with the rolling direction in the cold-rolled annealed material, a tensile test is performed at room temperature (for example, 20 to 25 °C) in a strain rate of 1.0×10^{-3}/s using the tensile tester until the material is broken. The slope change of the true strain-true stress curve obtained at this time is the work hardening rate. The change in the work hardening rate is closely related to the formation of the strain induced martensite phase. In the case of the work hardening rate, it gradually decreases as the tensile strain progresses after yielding, and then an inflection point is formed at the time point at which the strain induced martensite phase is generated and begin to contribute to the work hardening, i.e. at the critical strain. Then, when the tensile strain progresses due to the strain beyond the inflection point, and at the same time, when the formation of the strain induced martensite phase is increased, the work hardening rate is increased again.

[0063] Therefore, the critical strain value is a strain value at a point where the strain induced martensite phase is formed and begins to contribute to work hardening, and refers to the strain value at the point corresponding to the inflection point in the stress-strain curve obtained by the tensile test and mathematically refers to the point at which the second derivative value of the curve becomes zero.

[0064] Thus, when the value of the critical strain is less than 0.1, the strain induced martensite phase is easily formed at the time of strain, and the ductility of the material is rapidly lowered due to the rapid work hardening at the beginning of strain. In addition, when the strain induced martensite phase is formed too late, i.e., when the value of the critical strain exceeds 0.25, the elongation rate is lowered due to occurrence of necking which is a local stress concentration by lack of work hardening of the material. Therefore, there is an appropriate range of work hardening rate. Therefore, in the present invention, the range of the critical strain value for strain induced martensite phases formation is preferably 0.1 to 0.25.

[0065] In addition, it is very important to control the stability of the austenite phase in the lean duplex stainless steel according to the present invention. Generally, the strain induced martensite phase is a mild phase formed when the unstable austenite phase is strained, and induces work hardening and thus contributes to the increase in the elongation of the steel. In the case of the steel of the present invention, which is a duplex stainless steel consisting of austenite phase and ferrite phase, the stability of the austenite phase can be controlled using the proper distribution of alloying elements. In the present invention, a rapid solidification method is utilized as a method for enabling proper distribution of alloying elements. In the case of rapid solidification, since the time for diffusion occurring in the solid phase is insufficient, the austenite phase and the ferrite phase to be formed are solidified in a non-equilibrium state. When these non-equilibrium solidification phases are subjected to hot-rolling annealing for a short period of time, it is possible to control the stability of the austenite phases sufficiently within a desired range by utilizing the distribution of the generated alloying elements. As a method for achieving this, the alloy was designed so that most of nitrogen is segregated on the austenite phases by keeping the content of nitrogen, which has a fast diffusion rate in the solid phase, at the content higher than the normal content.

Examples

[0066] Hereinafter, elongation and corrosion resistance will be described in detail through various embodiments of the lean duplex stainless steel according to the present invention.

[0067] The specimens were prepared using a molten steel whose component contents were adjusted as shown in the following table 1. Thereafter, the phase fraction of the material was controlled by proceeding with hot rolling, hot rolling, cold rolling and then cold rolling annealing, and then the elongation and corrosion resistance were measured.

[0068] The tensile test specimens were measured by adjusting the tensile strain rate to a rate of 1.0×10^{-3}/s at room temperature after processing the specimens of ASTM-sub size in parallel to the rolling direction. The following table 1 shows the alloy composition (wt.%) of the test steel.

[0069] In addition, the following table 2 below shows the phase fractions of the ferrite phases and the austenite phases of some experimental steels in table 1 above measured after annealing heat treatment thereof at 1100 °C.

[0070] In addition, table 3 below shows the results of the stacking fault energy values, the differences in Gibbs free energies, the presence or absence of the strain induced martensite phase, the critical strain values, and the elongations for the inventive steels used in the description of the present invention and comparative steels, which were calculated by formula 2 while taking into account the stacking fault energy values, distribution coefficients and phase fraction calculated by formula 1 without considering the distribution coefficient

[0071] At this time, the reason for utilizing the Gibbs Free energy difference is that if the difference of the thermodynamic Gibbs free energies calculated when the crystal structure of the phase having the same components is austenite with FCC and when it is martensite with BCC satisfies the condition of $\Delta G = G^M - G^\gamma \leq 0$ (Gibbs energy of martensite phase - Gibbs energy of austenite phase), a strain induced martensite phase is formed. Thus, the Gibbs free energy difference and formation of strain induced martensite phase are closely related. For example, it means that when the Gibbs free energy difference (ΔG) is positive, a strain induced martensite phase is not formed, and when the Gibbs free energy difference (ΔG) is negative, a strain induced martensite phase is formed.

[0072] In this example, the Gibbs free energies of the austenite phases and the martensite phases were calculated using commercial software FACTsage 6.4 (Thermfact and GTT-Technologies). In particular, in order to calculate the Gibbs free energies, it is first necessary to know the components present in the austenite phase in the steel which is present in two phases of ferrite-austenite phases. The amount of the alloy component present in the austenite phase can be calculated by using the distribution coefficient and the phase fraction shown in the present invention. For example, it is possible to calculate using the component $X = X/[K(X) - K(X) \times V(\gamma) + V(\gamma)]$ (X: total X component, K(X): distribution coefficient, $V(\gamma)$: austenite phase fraction) in the austenite phase.

[0073] In addition, the presence or absence of formation of the strain induced martensite phase was measured by using a ferrite scope (commercial product) in the crack stretching section before necking during tensile strain

Table 1:

Steel	C	Cr	Ni	Mn	Si	Cu	N	Mo	W
Comp. steel 1	0.025	21.84	2.51	1.76	0.54	0.47	0.19	0.58	-
Comp. steel 2	0.021	20.3	0.198	5.05	0.217	-	0.102	-	-
Comp. steel 3	0.048	19.97	-	3.02	0.201	1	0.284	-	-
Comp. steel 4	0.041	23.1	0.3	3.2	2.9	0.55	0.15	-	-
Comp. steel 5	0.003	19.8	2.4	1.7	0.3	1.6	1.28	-	-
Inventive steel 1	0.051	19.87	0.5	2.91	0.865	1	0.24	-	-
Inventive steel 2	0.05	20.12	0.205	3.03	2	0.8	0.234	-	-

(continued)

Steel	C	Cr	Ni	Mn	Si	Cu	N	Mo	W
Inventive steel 3	0.021	19.9	0.8	3.05	0.6	1.04	0.261	-	-
Inventive steel 4	0.052	19.7	1.01	3.1	1.2	2	0.256	-	-
Inventive steel 5	0.051	21	1.02	3.02	1.1	2.03	0.254	-	-
Inventive steel 6	0.034	20	1.51	3	1.95	1.5	0.251	-	-
Inventive steel 7	0.049	19.95	1.95	2.7	0.9	2.02	0.251	-	-
Inventive steel 8	0.05	19.95	1.01	2.97	2.6	1	0.235	-	-
Inventive steel 9	0.0514	19.93	1.04	2.96	1.53	1	0.32	-	0.9
Inventive steel 10	0.047	21.33	1.02	3.04	1.53	1	0.23	-	0.47

Table 2:

Steel	Fraction of ferrite phase	Fraction of austenite phase
Comp. steel 1	51	49
Comp. steel 2	83	17
Comp. steel 3	35	65
Comp. steel 4	54	46
Comp. steel 5	41	59
Inventive steel 1	45	55
Inventive steel 2	37	63
Inventive steel 3	40	60
Inventive steel 4	38	50
Inventive steel 5	28	53
Inventive steel 6	33	67
Inventive steel 7	42	58
Inventive steel 8	47	53
Inventive steel 9	42	58
Inventive steel 10	48	52

Table 3:

Steel	Stacking fault energy (formula 1) without considering distribution (mJ/m^2)	Stacking fault energy (formula 2) of austenite phase with considering distribution (mJ/m^2)	Gibbs free energy	presence or absence of strain induced martensite phase	Critical strain value	Elongation, %
Comp. steel 1	29.94	33.3	positive	absence	-	31
Comp. steel 2	15.53	33.23	negative	presence	0.29	38

(continued)

Steel	Stacking fault energy (formula 1) without considering distribution (mJ/m^2)	Stacking fault energy (formula 2) of austenite phase with considering distribution (mJ/m^2)	Gibbs free energy	presence or absence of strain induced martensite phase	Critical strain value	Elongation, %
Comp. steel 3	26.21	30.87	negative	presence	0.07	35
Comp. steel 4	10.51	17.88	negative	presence	0.08	33
Comp. steel 5	31.62	37.75	positive	absence	-	36
Inventive steel 1	17.51	23.47	negative	presence	0.12	52
Inventive steel 2	19.33	25.60	negative	presence	0.143	49
Inventive steel 3	25.59	31.54	negative	presence	0.24	48
Inventive steel 4	24.93	33.17	negative	presence	0.238	53
Inventive steel 5	24.20	32.03	negative	presence	0.219	51
Inventive steel 6	22.89	28.62	negative	presence	0.167	59
Inventive steel 7	27.28	33.37	negative	presence	0.22	57
Inventive steel 8	19.33	27.40	negative	presence	0.182	50.5
Inventive steel 9	25.49	32.78	negative	presence	0.24	52
Inventive steel 10	20.96	28.59	negative	presence	0.189	54

[0074] In the case of the lean duplex stainless steel, the phase fraction varies depending on the alloy components and the heat treatment temperature.

[0075] Thus, table 2 shows the fractions of ferrite phases and austenite phases when heat-treating comparative steels 1 to 5 and inventive steels 1 to 10 at 1100 °C, respectively. It can be seen that in inventive steels 1 to 10, the fractions of the ferrite phases are in the range of about 25 to 55% and the fractions of the austenite phase are in the range of 45 to 75%. Meanwhile, in comparative steel 2, when heat-treating at 1100 °C, the fraction of ferrite phases was 83%, and at this time, the fraction of the austenite phases was 17%. That is, it can be seen that comparative steel 2 is not included in the range of the fractions of the ferrite phase and the austenite phase of the present invention.

[0076] In addition, it can be seen that comparative steel 4 has the stacking fault energy of the austenite phase of 17.88 mJ/m^2 when considering distribution, and thus is not included in the appropriate range of the stacking fault energy (SFE) value of the austenite phase.

[0077] On the other hand, fig. 1 is a representative nominal strain-nominal stress comparison curve obtained by the present invention wherein the results of the tensile test after annealing each material at 1100 °C are shown.

[0078] In the case of comparative steel 1, strain induced martensite phases were not formed during uniform strain. As a result, there were no strain induced martensite phases that can suppress localized necking due to work hardening during plastic strain, and thus a decrease in elongation was predicted. In fact, the elongation of comparative steel 1 was about 30% which is very inferior.

[0079] In addition, in the case of comparative steel 3, it was confirmed that the value of the critical strain for strain induced martensite phase formation at the time of plastic strain is 0.1 or less (inflection point; indicated by an arrow). As a result, it was predicted that the elongation would be lowered due to rapid work hardening in accordance with rapid formation of strain induced martensite phases. In fact, comparative steel 3 is two-phase stainless steel composed of ferrite phase and austenite phase, but the elongation was about 35% which is very inferior.

[0080] On the other hand, in the case of the steels of the present invention, it has been shown that when the critical strain range for strain induced martensite phase formation in the stress-strain curve is in the range of 0.1 to 0.25, various values of elongation can be obtained by suitably controlling the rate of formation of the strain induced martensite phase which is formed during processing. That is, when inventive steel 8 and inventive steel 1 are compared, it can be seen that the larger the critical strain for strain induced martensite phase formation, the greater the strain rate. This is to control the transformation of the austenite phases into the strain induced martensite phases during cold working, and the elongation is mostly 45% or more. This result shows that the steel of the present invention has an excellent elongation rate compared to that of the conventional lean duplex stainless steel of comparative steel 1 to be substituted and that the steel of the present invention has an excellent elongation ratio comparable to the elongation of 304 steel to be substituted.

[0081] In addition, when the critical strain value is less than 0.1, a sharp reduction of the elongation is caused by the hardening of the material due to the rapid work hardening while the strain induced martensite phase is rapidly formed. In addition, when the critical strain value is more than 0.25, the strain induced martensite phase is formed too late and the local necking of the material due to the strain cannot be suppressed. Therefore, in the case of the lean duplex stainless steel made of the austenite phase-ferrite phase of the present alloy system, when the value of the critical strain for strain induced martensite phase formation becomes 0.1 to 0.25, it is possible to ensure an elongation of 45% or more, which is much better than the elongation of the conventional duplex stainless steel of 30% or less, and to ensure the elongation of 45% or more, which is comparable to 304 steel, under the condition of some strains. Accordingly, the value of the critical strain for strain induced martensite phase formation during the cold processing is preferably 0.1 to 0.25.

[0082] In addition, as shown in comparative steel 1 and comparative steel 5 in table 3, it was confirmed that even if the stacking fault energy according to formula 2 is in the range of 19 to 37, the value of the Gibbs free energy at which the martensite phase is formed on the austenite shows a positive value, so that formation of strain induced martensite phase is not observed in the microstructure, observed by the transmission electron microscope, and in this case, it was observed that the elongation is lowered.

[0083] Also, as shown in the comparative steel 2 and the comparative steel 3 in table 3, it was observed that even if the value of Gibbs free energy is negative and thus the strain induced martensite phase is formed, the elongation is lowered even when the critical strain value is not in the range of 0.1 to 0.25.

[0084] Figs. 2a and 2b show transmission electron microscopic microstructures of comparative steel 1 and inventive steel 1, respectively. In the case of comparative steel 1, as shown in Fig. 2a, it can be seen that strain bands or mechanical twins due to the strain are observed, but the strain induced martensite phase is not observed. In the case of invention steel 1, as shown in Fig. 2a, it can be seen that the strain induced martensite phase is formed at the intersection of the strain bands or the mechanical twins (the strain induced martensite phase is indicated by an arrow).

[0085] Fig. 3 is a graph showing the relationship between the elongation and the value of the critical strain for strain induced martensite phase formation. The results of measuring the critical strain for martensite phase formation referring to the stress-strain curve of Fig. 1 are shown in Fig. 3.

[0086] Referring to Fig. 3, it can be confirmed that when the values of the critical strain are less than 0.1 or more than 0.25, the elongation of 45% is not ensured, but when the value range of critical strain is 0.1 to 0.25, the elongation of 45% or more can be ensured.

[0087] Meanwhile, the lean duplex stainless steels according to the present invention can be manufactured by both the continuous casting method and the strip casting method by solving the problem of nitrogen gas generation or emission depending on the difference in nitrogen solubility when solidified from a liquid phase to a solid phase.

[0088] First, a method for manufacturing lean duplex stainless steel by a continuous casting method will be described.

[0089] Fig. 4 is a schematic view illustrating the manufacturing process of the continuous casting method of the lean duplex stainless steel according to an embodiment of the present invention.

[0090] The lean duplex stainless steel according to one embodiment of the present invention is manufactured in a conventional continuous facility 100 in which ladle 110, tundish 120, mold 130, and a plurality of segments 140 are sequentially disposed. Additionally, the rear end of the segment 140 is further provided with a spraying mean 150 for spraying air and cooling water mixed with each other.

[0091] In order to manufacture the lean duplex stainless steel by the continuous casting method, first, a molten steel having the above-described alloy components is prepared and moved to the ladle 110, and then temporarily stored in the tundish 120 using the shrouding nozzle 111. At this time, it is preferable that the molten steel temporarily stored in the tundish 120 is maintained at a temperature higher than the theoretical solidification temperature by 10 to 50 °C.

[0092] In other words, ΔT (°C.), which is the difference between the temperature of the molten steel and the theoretical solidification temperature in the tundish 120, has a lower limit of 10 °C and an upper limit of 50 °C. The reason is that if ΔT is lower than the lower limit of 10 °C, solidification of the molten steel M can proceed in the tundish 120, thereby causing a problem in continuous casting, and if ΔT exceeds the upper limit of 50 °C, the solidification rate is lowered during solidification and the solidification structure is coarsened, so that solidification cracks in the continuous casting cast slab and linear defects during hot rolling are liable to occur.

[0093] Then, in the tundish 120, the molten steel M is injected into the mold 130 using the immersion nozzle 121. At this time, in the mold 130, the molten steel M is passed through the mold 130 to be primary cooled while maintaining the cooling rate of the molten steel M at 500 to 1500 °C/min.

[0094] At this time, when the cooling rate becomes less than 500 °C/min, nitrogen gas generated by the difference in nitrogen solubility due to the initially formed delta ferrite solidification is discharged through the solidified shell of the mold 130, so that coarse nitrogen pinholes are generated and thus a large number of nitrogen pinholes are generated in the continuously cast slab. In addition, coarsening of the initially formed delta ferrite occurs and becomes vulnerable to external stress. Also, when the cooling rate is less than 500 °C/min, the amount of cooling (primary cooling) in the mold 130 and the amount of cooling (secondary cooling) in the segment 140 during the continuous casting are reduced and thereby, during casting, the heat transfer of the cast slab S is delayed, the strength of the cast slab solidified layer is lowered, and the cast slab is bulged, resulting in the deterioration of operation and quality.

[0095] In addition, when the cooling rate is controlled to exceed 1500 °C/min, although it is very advantageous from the viewpoint of nitrogen pinhole, continuous operation is not possible due to the current limitations of the continuous casting facility, and the time for diffusing the segregation of solute elements remaining between dendrites during continuous casting is shortened, thereby generating surface cracks of the cast slab. There is a problem in that an overlapping phenomenon in which the shell of the cast slab is temporarily broken inside the mold 130 due to such a phenomenon occurs. Therefore, it is preferable to set the cooling rate during the primary cooling in the mold 130 to 500 to 1500 °C/min. The molten steel M, that is, the cast slab S having the solidified shell formed in the casting mold 130 is drawn into the segment 140 and thus cooled secondarily, and at this time, it is preferable to spray the cooling water of 0.25 to 0.35 L/kg to the cast slab S. The reason for limiting the spraying amount in the segment 140 in this way is as follows.

[0096] If the amount of the sprayed water in the segment 140, i.e., the secondary cooling zone is set to a relatively large value, the solidified structure can be finely formed, but if the amount of the sprayed water exceeds 0.35 L/kg, since the period of time for the segregated impurities to diffuse between the solidified structures in the continuous casting process is reduced, the cast slab is present in a sigma phase, and cracks are generated on the surface of the cast slab. In addition, since not only cracks due to thermal stress but also residual stress is generated on the surface excessively, surface cracks occur during grinding of the cast slab. In addition, when the amount of the sprayed water is less than 0.25 L/kg, there are problems in that the solidification structures become excessive and thus solidification cracks are generated by the sigma phase generated in the grain boundary and in that the strength of the solidification shell of the cast slab is lowered during the continuous casting and the cracks due to bulging of the cast slab is generated.

[0097] Therefore, the range of the amount of the sprayed water in the segment 140 is preferably 0.25 to 0.35 L/kg.

[0098] In addition, while drawing to the segment 140, a tertiary cooling is performed on the secondarily cooled cast slab S. The tertiary cooling is carried out by spraying the cooling water of 100 to 125 L/kg·min on the entire surface of the cast slab S in the surface temperature range of the cast slab S at 1100 to 1200 °C wherein the air and cooling water are mixed so that the ratio thereof is 1.0 to 1.2 (air/cooling water), while continuing to draw into the segment 140.

[0099] The tertiary cooling is controlled so as to ensure a uniform scale on the surface of the cast slab S. The reason for this is that in the case of the lean duplex stainless steel, since the oxidation amount in the heating furnace is very small, the lubrication effect by the scale during the hot rolling is small and thus it is very difficult to reduce surface cracks. Therefore, in order to prevent the reduction of temperature due to the contact between the roll and the steel sheet during rolling and to reduce the frictional force between the roll and the steel sheet to prevent surface cracking, a dense and thick scale should be formed on the surface of the steel sheet and a peeling should not occur easily during rolling. As mentioned above, the reason for limiting the surface temperature of the cast slab S, the amount of the cooling water, and the ratio of the cooling water to air (air/cooling water) is that if the above conditions are not satisfied, a scale having a desired level of thickness (approximately $35\mu m \pm 2\mu m$) is not formed on the surface of the slab S, and the generated scale is not uniformly formed.

[0100] Hereinafter, the lean duplex stainless steels having the composition according to the present invention were produced by producing a cast slab while changing the molten steel temperature in the tundish, the cooling rate in the casting mold, and the amount of the sprayed water in the secondary cooling zone as shown in table 4, and the degrees of occurrence of pinholes and cracks on the surface of the cast slab generated therefrom are shown together in table 4. At this time, the occurrence or not of pinholes in the cast slab was measured by grinding the surface of the slab by about 0.5 mm and observing the grinded surface.

Table 4:

Material	Degree of overheating of molten steel in tundish (°C)	Cooling rate in mold (°C/min)	Amount of sprayed water in the cooling zone (L/kg)	Degree of occurrence of pinhole	Degree of cracking on surface of continuous casting cast-slab
Inventive material A	15	1350	0.29	non	non
Inventive material B	20	1100	0.32	non	non
Inventive material C	15	1100	0.27	non	non
Inventive material D	25	850	0.29	non	non
Inventive material E	22	550	0.3	non	non
Comp. material F	19	1100	0.4	non	serious
Comp. material G	13	1100	0.2	non	weak
Comp. material H	20	400	0.3	weak	non
Comp. material I	15	60	0.28	serious	non
Comp. material J	19	40	0.29	serious	non

[0101] As can be seen from table 4, inventive materials A to E which satisfy all of the control conditions of the present invention did not cause pinholes in continuous casting of the cast-slab due to nitrogen and did not cause bulging and defects on the surfaces of the hot-rolled coils.

[0102] In addition, since the cooling rates of comparative materials F and G in the mold were within the range of the present invention, pinholes due to nitrogen were not generated in the cast slab. However, in the case of comparative material F, since the amount of the sprayed water was larger than the range of the present invention, no bulging occurred during casting, but thermal stress was severely exerted on the surface of the cast slab, thereby causing occurrence of cracks.

[0103] Also, in comparative material G, since the range of the amount of the sprayed water in the secondary cooling zone was less than the range of the present invention, and thus bulging occurred in the cast-slab, thereby causing cracks on the surface of the cast-slab. As a result, linear defects occurred on the surface of the hot-rolled coil due to the formation of a local excessive scale during hot rolling.

[0104] Also, in comparative materials I and J, the cooling rate in the mold was lower than the range of the present invention, and thus serious pinholes were generated in cast-slab. However, the amount of the sprayed water in the secondary cooling zone is within the scope of the present invention, and thus the surface of the continuous casting cast-slab is good, but a large number of linear defects occurred during the hot rolling due to the pinholes present in the cast-slab.

[0105] On the other hand, Fig. 7 is a photograph of the structure of comparative material I and inventive material A produced according to the continuous casting method of the present invention, Fig. 8 is a photograph of surface defects of the comparative material H produced according to the continuous casting method of the present invention and Fig. 9 is a photograph of surface defects of comparative material F produced according to the continuous casting method of the present invention. At this time, Figs. 8 and 9 are photographs of surface defects of the surface of the hot-rolled coils found after the hot rolling of comparative materials H and F.

[0106] As can be seen from FIG 7, it is confirmed that no pinholes were found on the surface of the inventive cast A-slab, but a large number of pinholes were found in comparative material I. Further, as can be seen from Fig. 8, when the surface of the hot-rolled coil after the hot rolling of comparative material H having relatively good occurrence of pinholes are observed, a large number of pinhole-like defects drawn in the rolling direction are observed. In addition,

Fig. 9 shows that when the surface of the hot-rolled coil after hot rolling of comparative material F is observed, a large number of crack-like surface defects of cast-slab are observed.

[0107] Therefore, it has been confirmed through various embodiments that by properly controlling the cooling rate in the mold and the amount of the sprayed water in the secondary cooling zone during continuous casting according to the present invention, excellent cast-slab quality for the lean duplex stainless steel composed of the austenite phase and the ferrite phase with suppressed occurrence of pinholes, cracking during casting, and bulging can be obtained as well as stable continuous casting operation is possible.

[0108] In addition, cast-slab was produced using the lean duplex stainless steel having the composition according to the present invention and subjecting to primary cooling and secondary cooling, while changing the amount of cooling water, the period of spraying time, the air/cooling water ratio, and the surface temperature of cast-slab as shown in table 5, and the thicknesses and the degree of uniformity of the scale obtained therefrom are shown together in Table 5.

Table 5:

Material	Amount of cooling water (L/kg.min)	Period of spraying time (min)	Air/cooling water	Surface temperature of cast-slab (°C)	Thickness of scale (mm)
Inventive material 1	100	28	1.0	1100	35 (uniform)
Inventive material 2	110	22	1.1	1160	34 (uniform)
Inventive material 3	120	20	1.0	1156	37 (uniform)
Inventive material 4	100	22	1.1	1121	33 (uniform)
Com. material 1	50	20	1.0	1111	22 (non-uniform)
Com. material 2	80	20	1.0	1121	30 (non-uniform)
Com. material 3	100	20	0.5	1082	10 (non-uniform)
Com. material 4	100	20	0.6	1198	12 (non-uniform)
Com. material 5	100	20	0.8	1145	23 (non-uniform)
Com. material 6	100	15	1.0	1220	22 (non-uniform)
Com. material 7	100	10	1.0	1230	12 (non-uniform)
Com. material 8	100	20	1.0	932	15 (non-uniform)
Com. material 9	100	20	1.0	1062	26 (non-uniform)

[0109] As in inventive materials 1 to 4, it can be seen that when spraying the cooling water for 20 to 30 minutes at a cooling rate of 100 to 120 L/kg · min while keeping the ratio of air/cooling water at 1.0-1.2 at the surface temperature point of cast-slab of 1000 to 1200 °C, the scale becomes very uniform and thick.

[0110] However, as in comparative materials 1 and 2, when the cooling water was sprayed at 50 L/kg · min or 80 L/kg · min, the amount of cooling water was insufficient and thus the generation of the oxidized scale was not promoted and a uniform oxidized scale could not be obtained.

[0111] Also, as in comparative materials 3 to 5, it was confirmed that when the oxidized scale is investigated while varying the ratio of air to cooling water (air/cooling water), the greater the amount of air, the greater the thickness of the scale layer. Accordingly, in order to form a scale layer having a desired thickness, it is desirable to maintain the ratio of air to cooling water (air/cooling water) to 1.0 or more. However, if the ratio of air exceeds the upper limit of 1.2, a sufficient scale layer can be obtained, but there is a concern that the entire cooling water system will be hindered

[0112] As in the comparative materials 6 and 7, when the cooling water is sprayed for the period of spraying time of 15 minutes and 10 minutes, although the temperature of the spraying point of cooling water and the ratio of air/cooling water are in the vicinity of the conditions of the inventive material, a uniform and sufficiently thick scale layer cannot be obtained due to the shortage of the period of spraying time. Therefore, it has been confirmed that in order to obtain a uniform and thick scale layer, the sufficient period of spraying time is required for the slab to react with the air. However, if the period of spraying time of cooling water exceeds a certain period, there is a concern that the cast-slab may become stagnant and the yield may decrease.

[0113] In addition, as in the comparative materials 8 and 9, when the cooling water is sprayed at the surface temperature points of cast-slab at 932 and 1062 °C, the thicknesses of the oxidized scales were 15 and 26 μm and the oxidized scales were non-uniform. However, it can be predicted that as the temperature of the cast-slab increases, the formation of the oxidized scale is accelerated and a uniform scale can be obtained. It is also predicted that as the surface temperature of continuous casting cast-slab increases, a uniform scale layer can be obtained.

[0114] As described above, in the cooling spraying process after the completion of the continuous casting process, when the optimal amount of cooling water is sprayed at the optimal ratio of cooling water and air at the optimum spraying position, the scale formation can be optimized, the surface quality can be improved, the cost for process for removing defects can be minimized, and thus the added value can be improved.

[0115] Next, a method of manufacturing a lean-duplex stainless steel by a strip casting method will be described.

[0116] Fig. 5 is a schematic view showing a manufacturing process of the strip casting method for the lean-duplex stainless steel according to an embodiment of the present invention, and fig. 6 is a schematic view of a nitrogen discharge channel formed in the casting roll of the present invention.

[0117] The lean duplex stainless steel according to an embodiment of the present invention is manufactured in a conventional strip casting facility 200 in which the ladle 210, the tundish 220, a pair of casting rolls 230, the inline rollers 260, and the winding rolls 270 are sequentially disposed. Additionally, the gas discharge channel 231 is formed on the surface of the casting roll 230

[0118] In order to produce the lean duplex stainless steel by the strip casting method, first, a molten steel M having the above-described alloy components is prepared and moved to the ladle 210, and then temporarily stored in the tundish 220 using the shrouding nozzle 211. Then, the molten steel M is solidified while passing between a pair of casting rolls 230 through the injection nozzle 221 to produce the strip S, and the manufactured strip S is rolled in an inline roller 260 disposed continuously with a casting roll 230 and wound around a winding roll 270.

[0119] On the other hand, on the upper side of the casting roll 230, a manifold shield 250 is mounted to prevent the surface of the molten metal from being oxidized by contact with air, and an appropriate gas is added to the inside of the manifold shield 250 to appropriately form an anti-oxidizing atmosphere.

[0120] In this way, the molten steel M is rolled through the inline roller 260 while exiting the roll nip where a pair of casting rolls 230 meet, and then passes through the process such as a heat treatment process and a cold rolling process to form a strip S of 10 mm or less.

[0121] One of the most important technical elements in the twin-roll strip casters which directly manufacture the above-mentioned strip S of 10 mm or less is that the molten steel M is provided between the side dam 240 and the inner water-cooled twin-drum rolls 230 rotating at a high speed through the injection nozzle 250 in opposite directions wherein the molten steel M is provided so that the molten steel is rapidly cooled by releasing a large amount of heat through the surface of the water-cooled casting roll 230 and the actual yield is improved without cracking the thin plate of desired thickness.

[0122] In the method of producing the highly ductile lean duplex stainless steel of the present invention, the problems associated with nitrogen exceeding the solubility limit contained in the molten steel, which is the cause of edge cracking and surface cracking, have been solved, and the problem of lowering hot workability due to the nitrogen content has been solved.

[0123] That is, by completing the rapid casting while discharging nitrogen exceeding the solubility limit at the time of solidifying the molten steel M through the casting roll 230, and completing the rapid casting by using the inline roller 260, the above-described problem has been solved by producing a thin strip S of about 2 to 5 mm.

[0124] Various means can be suggested in order to remove nitrogen exceeding the solubility limit contained in the molten steel M during the strip casting process. As one example, in the manufacturing method of a highly ductile duplex stainless steel according to the present invention, the nitrogen discharge channel 231 is formed on the surface of the casting roll 230, thereby discharging nitrogen exceeding the solubility limit in solidifying the molten steel.

[0125] The problem of internal porosity due to nitrogen is mostly generated in the process of rapidly cooling the molten steel M while passing between a pair of casting rolls 230.

[0126] Therefore, the discharge of nitrogen exceeding the solid solubility limit in the molten steel M must be performed simultaneously with the passage of the molten steel M through the casting roll 230. For this purpose, it is preferable that the gas discharge channel 231 is formed on the surface of the casting roll 230 so that nitrogen can be discharged during casting.

[0127] The gas discharge channel 231 is a fine channel to which only the nitrogen gas can be discharged while the molten steel M cannot pass. These gas discharge channels 231 may be formed in the casting roll 230 in various ways and may be formed in the circumferential direction on the surface of the casting roll 230 to guide and discharge the nitrogen gas toward the outside of the casting roll 230 in accordance with the rotation of the casting roll 230.

[0128] It is preferable that the gas discharge channel 231 corresponds to a fine channel having a width of 50 to 500 μm and a depth of 50 to 300 μm, and a plurality of gas discharge channels 231 are formed in the circumferential direction of the casting roll 230 wherein the distance between adjacent gas discharge channels 231 is about 100 to 1000 μm.

[0129] The shape, structure, and application position of the gas discharge channel 231 may be variously modified as long as the function thereof can be achieved.

[0130] On the other hand, when a plurality of such gas discharge channels 231 are formed, the contact area between the casting roll 230 and the molten steel M passing through the casting roll 230 can be reduced and thus in order to prevent this, it is preferable that unevennesses are preferably formed on the surface of the casting roll. These unevennesses have an average size of 15 to 25 μm.

[0131] Hereinafter, in order to confirm the effect of nitrogen exceeding the solubility limit in molten steel on the thin plate, a lean duplex stainless steel was produced while changing the composition and casting method of molten steel as shown in the following table 6. Here, comparative example 1 is an example in which molten steel having a specific composition is cast by using a general continuous casting method, comparative example 2 is an example in which molten steel having a specific composition is cast by using a general strip casting (rapid casting) method, and examples 1 to 5 are examples in which casting is performed by a strip casting process while discharging nitrogen exceeding the solubility limit in molten steel by using the casting roll according to the present invention.

Table 6:

Ex.	C	Si	Mn	Cr	Ni	Cu	N	Casting method	Nitrogen discharge	Internal pore
Comp. ex. 1	0.05	1.35	2.8	20.3	1.06	1.0	0.23	continuous casting	X	O
Comp. ex. 2	0.05	1.35	2.8	20.3	1.06	1.0	0.23	rapid casting	X	O
Ex. 1	0.045	1.08	3.02	19.63	0.98	0.98	0.272	rapid casting	O	X
Ex. 2	0.021	1.3	3.2	19.89	1.14	0.8	0.28	rapid casting	O	X
Ex. 3	0.031	0.6	3.0	20.02	0.9	0.7	0.24	rapid casting	O	X
Ex. 4	0.033	1.2	3.09	20.21	0.8	0.7	0.24	rapid casting	O	X
Ex. 5	0.021	0.8	2.63	20.13	0.85	0.9	0.22	rapid casting	O	X

[0132] As shown in table 6, it was confirmed that in the case of comparative example 1, nitrogen was not discharged during the continuous casting process, and pores were generated inside the cast-slab.

[0133] In addition, it was confirmed that in the case of comparative example 2, nitrogen was not discharged during the conventional strip casting process and pores were generated inside the strip.

[0134] This is due to the difference in the solubility of nitrogen generated when the molten steel passes through the mold or the casting roll and solidifies

[0135] The nitrogen composition of the high ductility duplex stainless steel of the present invention ranges from 1500 to 3200 ppm.

[0136] Meanwhile, the process of solidification of the molten steel from the liquid phase to the solid phase proceeds in the order of liquid phase -> liquid phase + delta phase -> delta phase -> delta phase + austenite phase, and when the liquid phase changes into the delta phase, the solubility of nitrogen is about 1164 ppm and thus the solubility difference of about 836 to 1836 ppm occurs. Therefore, some of the supersaturated nitrogen in the liquid phase is gasified during solidification to form various pores in the solidified material and also to form many pores in the solidified shell formed on the surface of the material.

[0137] As described above, there are many pores in the actual solidified material. Some of these pores are squeezed during hot rolling, but the non-squeezed pores produce internal defects and develop into a various surface defect form when exposed to the outside during heating in the heating furnace.

[0138] On the other hand, it was confirmed that examples 1 to 5 are strip casting processes according to the present invention in which nitrogen is discharged during the process and pores are not generated in the strip.

[0139] Although the present invention has been described with reference to the accompanying drawings and the preferred embodiments described above, the present invention is not limited thereto but is limited by the following claims. Therefore, those skilled in the art can variously change and modify the present invention within the technical spirit of the following claims.

Explanation of symbols

[0140]

100: Continuous casting equipment, 110: Ladle
120: Tundish, 130: Mold
140: Segment, 150: Spraying means
200: Strip casting facility, 210: Ladle
220: Tundish 230: Casting roll
260: Inline roller 270: Winding roll

Claims

1. A ferritic-austenitic lean duplex stainless steel wherein the stacking fault energy (SFE) value of the austenite phase represented by the following formula 2 is 19 to 37 and the range of the value of the critical strain for strain induced martensite formation is 0.1 to 0.25:

$$SFE = 25.7+1.59\times Ni/[K(Ni)-K(Ni)\times V(\gamma)+V(\gamma)]+0.795\times Cu/[K(Cu)-K(Cu)\times V(\gamma)+V(\gamma)]-0.85\times Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)]+0.001\times(Cr/[K(Cr)-K(Cr)\times V(\gamma)+V(\gamma)])^{2}+38.2\times(N/[K(N)-K(N)\times V(\gamma)+V(\gamma)])^{0.5}-2.8\times Si/[K(Si)-K(Si)\times V(\gamma)+V(\gamma)]-1.34\times Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)]+0.06\times(Mn/[K(Mn)-K(Mn)\times V(\gamma)+V(\gamma)])^{2} \text{ ....Formula 2}$$

   wherein Ni, Cu, Cr, N, Si and Mn refer to the overall content (wt.%) of respective constituent element respectively, and K(x) is represented by the following Formula 3 as the distribution coefficient of respective constituent element (x), and $V(\gamma)$ is the fraction of the austenite phase (the range of 0.45 to 0.75):

$$K(x) = [\text{content of element x in ferrite phase}] / [\text{content of element x in austenite phase}] \text{ .... Formula 3.}$$

2. The lean duplex stainless steel according to claim 1, wherein regarding the K(x), K(Cr) = 1.16, K(Ni) = 0.57, K(Mn) = 0.73, and K(Cu) = 0.64, and K(N) and K(Si) have the following values depending on the content (wt.%) ofN and Si:

   when N is 0.2 to 0.32%, K(N) = 0.15;
   when N < 0.2%, K(N) = 0.25;
   when Si ≤ 1.5%, K(Si) = 2.76-0.96×Si; and
   when Si > 1.5%, K(Si) = 1.4.

3. The lean duplex stainless steel according to claim 1 or 2, wherein the elongation of the stainless steel is 45% or more.

4. The lean duplex stainless steel according to claim 1, wherein the stainless steel includes, by weight, C: 0.08% or less (excluding 0%), Si: 0.2 to 3.0%, Mn: 2 to 4%, Cr: 18 to 24%, Ni: 0.2 to 2.5%, Cu: 0.2 to 2.5%, balance Fe and the other unavoidable impurities.

5. The lean duplex stainless steel according to claim 4, wherein the stainless steel further includes, by weight, at least one of W: 0.1 to 1.0% and Mo: 0.1 to 1.0%.

6. The lean duplex stainless steel according to claim 4, wherein the stainless steel further includes, by weight, at least one of Ti: 0.001 to 0.1%, Nb: 0.001 to 0.05%, and V: 0.001 to 0.15%.

7. A method of manufacturing a ferritic-austenitic lean duplex stainless steel, comprising
preparing a molten steel; and
treating the molten steel to form the stainless steel wherein the molten steel is treated so that the stacking fault energy (SFE) value of the austenite phase represented by the following formula 2 is 19 to 37 and the range of the value of the critical strain for strain induced martensite formation is 0.1 to 0.25:

$$\begin{aligned} \mathrm{SFE} = {} & 25.7+1.59\times \mathrm{Ni}/[K(\mathrm{Ni})-K(\mathrm{Ni})\times V(\gamma)+V(\gamma)]+0.795\times \mathrm{Cu}/[K(\mathrm{Cu})-K(\mathrm{Cu})\times V(\gamma)+V(\gamma)]- \\ & 0.85\times \mathrm{Cr}/[K(\mathrm{Cr})-K(\mathrm{Cr})\times V(\gamma)+V(\gamma)]+0.001\times(\mathrm{Cr}/[K(\mathrm{Cr})-K(\mathrm{Cr})\times V(\gamma)+V(\gamma)])^2+38.2\times(\mathrm{N}/[K(\mathrm{N})- \\ & K(\mathrm{N})\times V(\gamma)+V(\gamma)])^{0.5}-2.8\times \mathrm{Si}/[K(\mathrm{Si})-K(\mathrm{Si})\times V(\gamma)+V(\gamma)]-1.34\times \mathrm{Mn}/[K(\mathrm{Mn})- \\ & K(\mathrm{Mn})\times V(\gamma)+V(\gamma)]+0.06\times(\mathrm{Mn}/[K(\mathrm{Mn})-K(\mathrm{Mn})\times V(\gamma)+V(\gamma)])^2 \ldots\ldots\ldots\ldots\ldots\text{Formula 2} \end{aligned}$$

wherein Ni, Cu, Cr, N, Si and Mn refer to the overall content (wt.%) of respective constituent element respectively, and K(x) is represented by the following Formula 3 as the distribution coefficient of respective constituent element (x), and V(γ) is the fraction of the austenite phase (the range of 0.45 to 0.75):

$$K(x) = [\text{content of element } x \text{ in ferrite phase}]/[\text{content of element } x \text{ in austenite phase}] \ldots\ldots\ldots\text{Formula 3.}$$

8. The method of manufacturing the lean duplex stainless steel according to claim 7, wherein the process of treating the molten steel to form the stainless steel comprises
temporarily storing the molten steel in the tundish while maintaining the temperature of the molten steel at the temperature higher than the theoretical solidification temperature by 10 to 50 °C;
primarily cooling the molten steel by injecting the molten steel in the tundish into the mold and passing the molten steel through the mold while maintaining a cooling rate of 500 to 1500 °C/min; and
secondarily cooling the molten steel having the solidified shell formed by the primary cooling process while drawing it into a segment and passing through.

9. The method of manufacturing the lean duplex stainless steel according to claim 8, wherein in the secondary cooling process, the cooling water of 0.25 to 0.35 L/Kg is sprayed on the molten steel having the formed solidified shell.

10. The method of manufacturing the lean duplex stainless steel according to claim 8, wherein the method further comprises tertiarily cooling, after the secondary cooling process, by spraying the cooling water of 100 to 125 L/kg · min on the surface of the cast-slab in the range of the surface temperature of the cast-slab being drawn of 1100 to 1200 °C wherein the cooling water is mixed with air such that the ratio of air to cooling water (air/cooling water) is 1.0 to 1.2.

11. The method of manufacturing the lean duplex stainless steel according to claim 7, wherein the process of treating the molten steel to form the stainless steel comprises producing a strip by solidifying the molten steel while passing it between a pair of casting rolls wherein nitrogen, which is contained in the molten steel in the process of producing the strip and exceeds a nitrogen solubility limit, is discharged through the casting roll to the outside of the solidified shell.

12. The method of manufacturing the lean duplex stainless steel according to claim 11, wherein in the process of producing the strip, at least one of a pair of the casting rolls is a casting roll having a gas discharge channel formed in a circumferential direction on the outer peripheral surface.

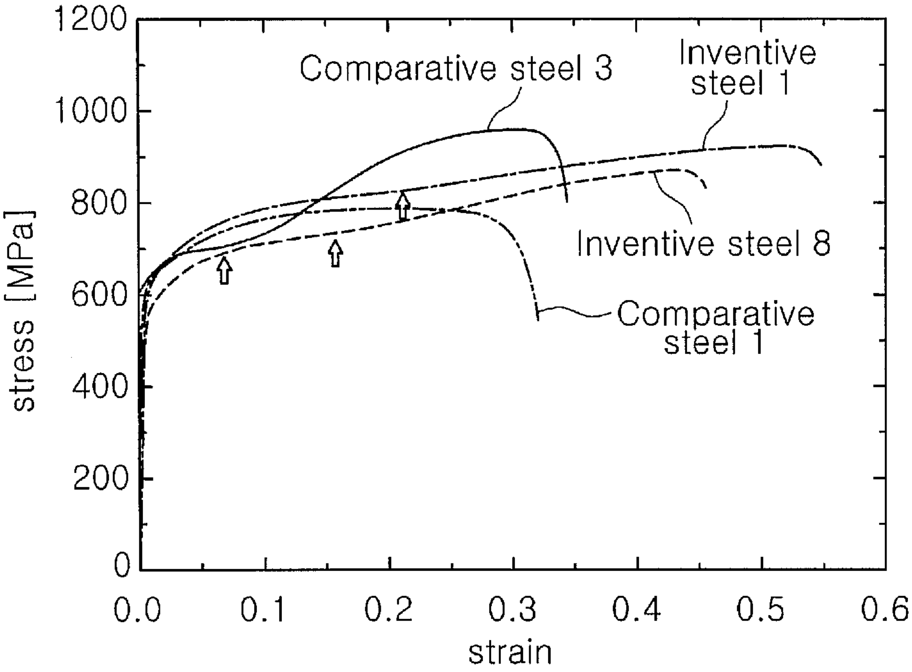
13. The method of manufacturing the lean duplex stainless steel according to claim 12, wherein the gas discharge channel formed in the casting roll used in the process of producing the strip has a width of 50 to 500 μm and a depth of 50 to 300 μm, a plurality of gas discharge channels is formed in the casting roll, the gap between adjacent gas discharge channels is 100 to 1000 μm and unevennesses of 15 to 25 μm are formed on the surface of the casting roll.

14. The method of manufacturing the lean duplex stainless steel according to claim 7, wherein, in the process of preparing the molten steel, the molten steel comprises, by weight, C: 0.08% or less (excluding 0%), Si: 0.2 to 3.0%, Mn: 2 to 4%, Cr: 18 to 24%, Ni: 0.2 to 2.5%, N: 0.15 to 0.32 %, Cu: 0.2 to 2.5%, balance Fe and the other unavoidable impurities.

15. The method of manufacturing the lean duplex stainless steel according to claim 14, wherein, in the process of preparing the molten steel, the molten steel further comprises, by weight, at least one of W: 0.1 to 1.0% and Mo: 0.1 to 1.0%.

16. The method of manufacturing the lean duplex stainless steel according to claim 14, wherein, in the process of preparing the molten steel, the molten steel further comprises at least one ofTi: 0.001 to 0.1%, Nb: 0.001 to 0.05%, and V: 0.001 to 0.15%.

[FIG.1]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1000
800
600
400
200
0
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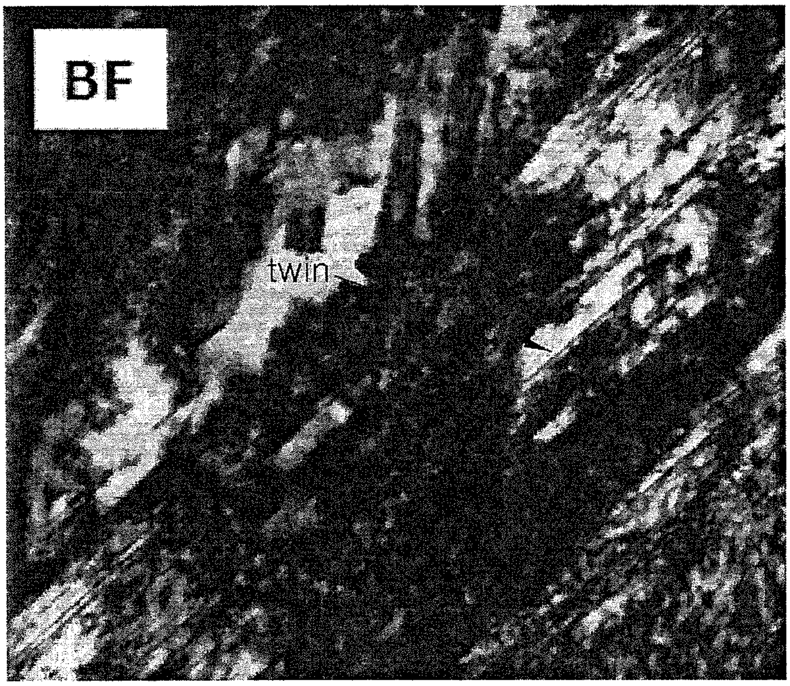
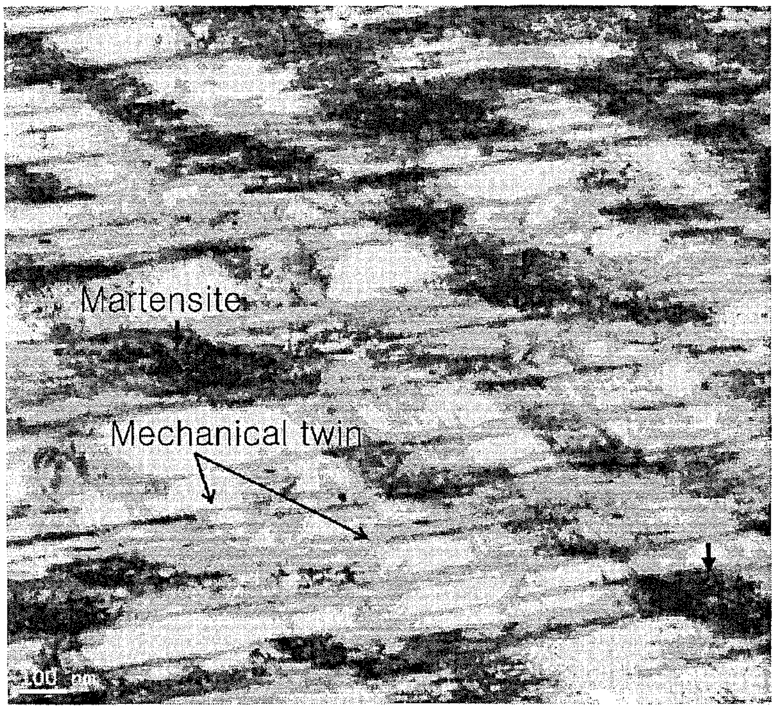
0.0
0.1
0.2
0.3
0.4
0.5
0.6
strain
Comparative steel 3
Inventive steel 1
Inventive steel 8
Comparative steel 1

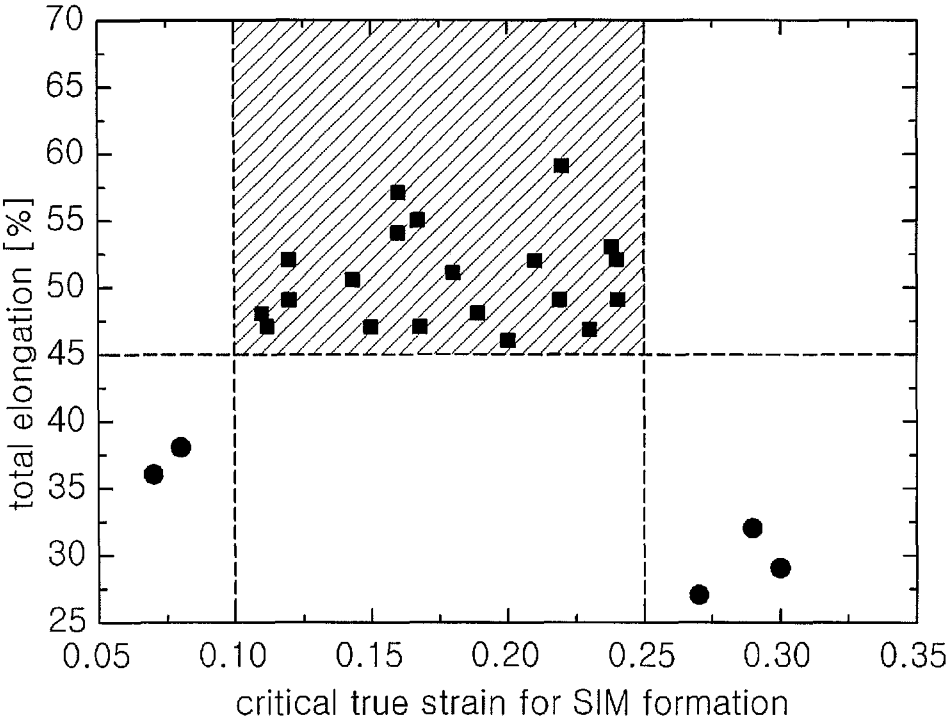
[FIG.2a]


BF
twin

[FIG.2b]


Martensite
Mechanical twin
100 nm

[FIG.3]


70
65
60
55
50
45
40
35
30
25
total elongation [%]
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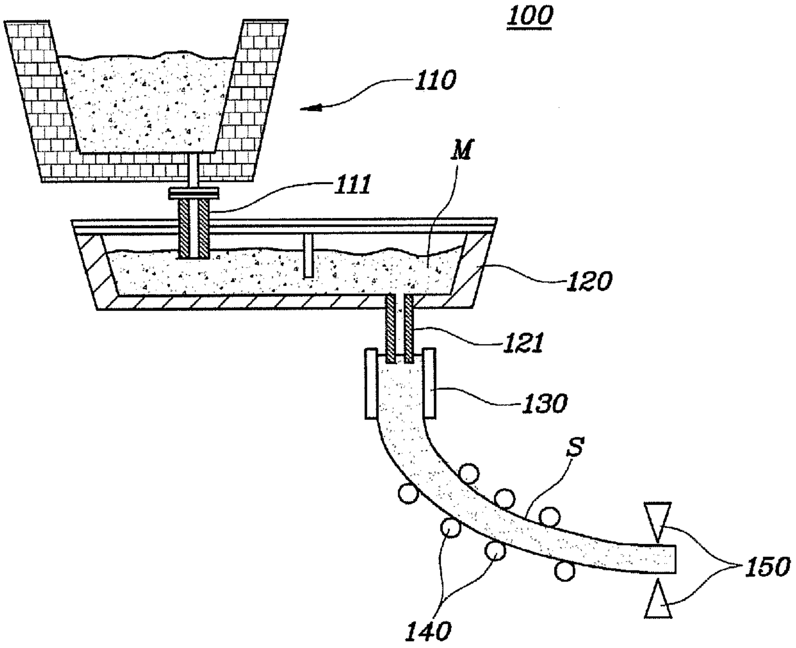
0.10
0.15
0.20
0.25
0.30
0.35
critical true strain for SIM formation

[FIG.4]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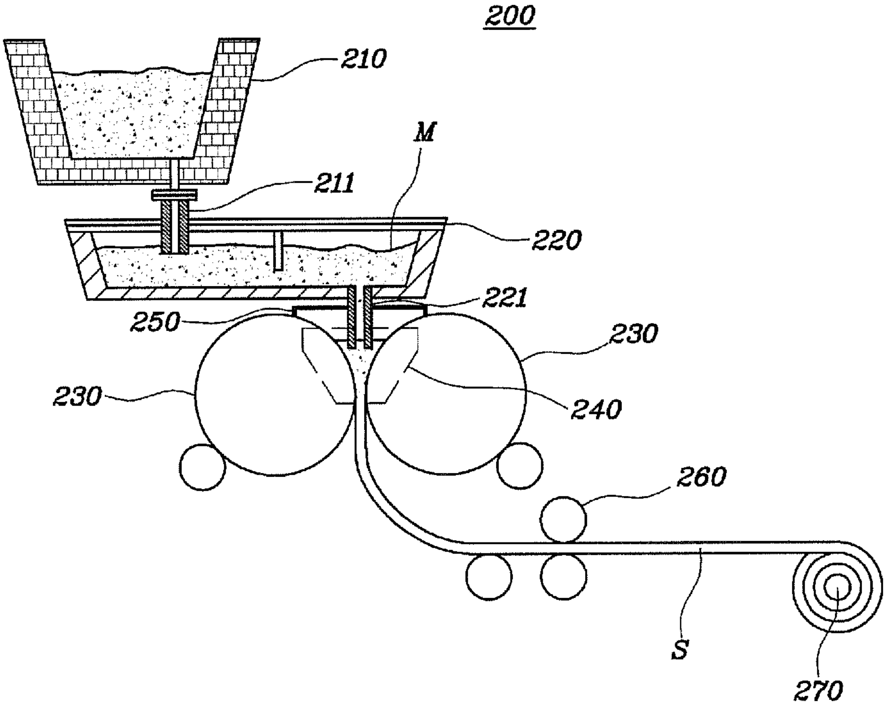
121
130
S
150
140

[FIG.5]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M
211
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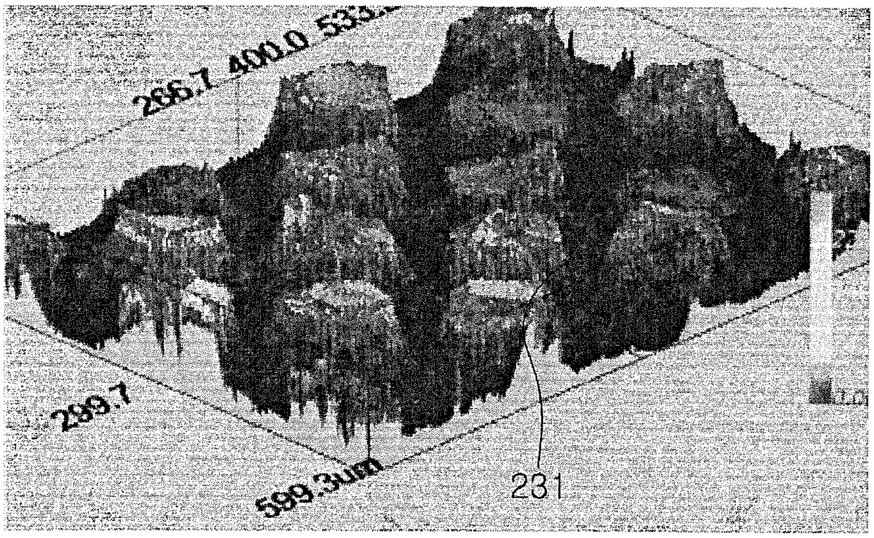
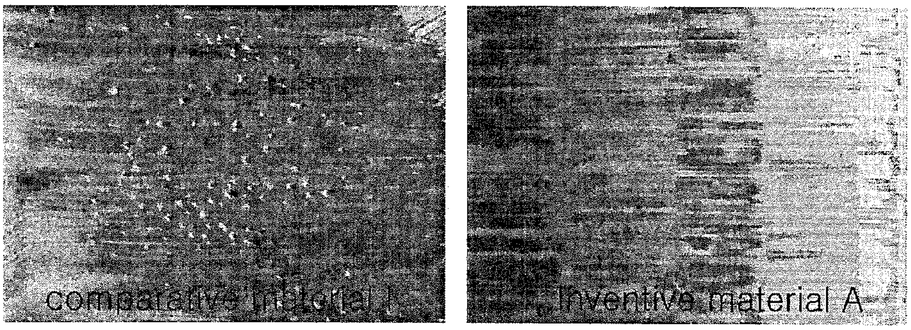
240
260
S
270

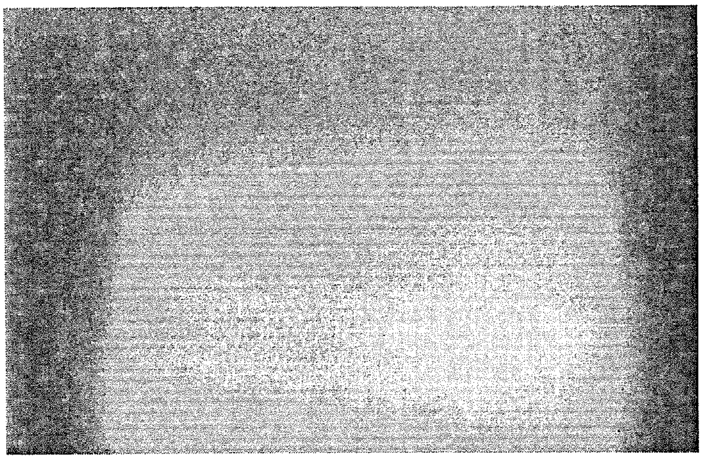
[FIG.6]


266.7
400.0
239.7
599.3um
231

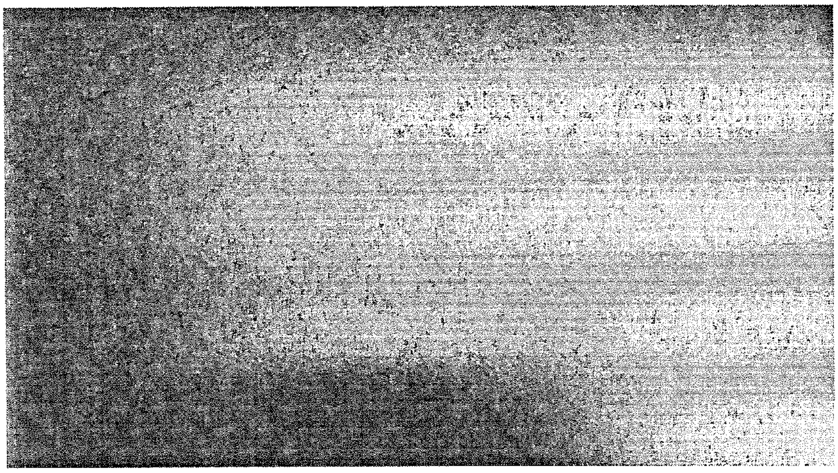
[FIG.7]


comparative material I
Inventive material A

[FIG.8]



[FIG.9]



INTERNATIONAL SEARCH REPORT

International application No.
PCT/KR2015/014235

A. CLASSIFICATION OF SUBJECT MATTER

C22C 38/58(2006.01)i, C22C 38/42(2006.01)i, B22D 11/16(2006.01)i, B22D 11/055(2006.01)i, B22D 11/124(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 38/58; B21B 1/28; C22C 38/00; C22C 38/42; C22C 38/18; C22C 38/02; B21B 3/02; B22D 11/16; B22D 11/055; B22D 11/124

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: duplex, stainless steel, ferrite, austenite, nickel, stack fault energy, critical strain

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
A	KR 10-2014-0052079 A (OUTOKUMPU OYJ) 02 May 2014 See paragraph [0027]; and claims 1, 17.	1-16
A	JP 2010-196103 A (NISSHIN STEEL CO., LTD.) 09 September 2010 See paragraphs [0023]-[0027]; and claim 3.	1-16
A	KR 10-2014-0080347 A (POSCO) 30 June 2014 See paragraphs [0069]-[0074]; and claims 1, 2.	1-16
A	JP 2005-154890 A (NIPPON STEEL & SUMIKIN STAINLESS STEEL CORP.) 16 June 2005 See paragraph [0036]; and claim 1.	1-16
A	KR 10-2014-0082491 A (POSCO) 02 July 2014 See paragraphs [0053]-[0055]; and claims 1, 5.	1-16

☐ Further documents are listed in the continuation of Box C. ☒ See patent family annex.

*	Special categories of cited documents:	"T"	later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention
"A"	document defining the general state of the art which is not considered to be of particular relevance	"X"	document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone
"E"	earlier application or patent but published on or after the international filing date	"Y"	document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art
"L"	document which may throw doubts on priority claim(s) or which is cited to establish the publication date of another citation or other special reason (as specified)	"&"	document member of the same patent family
"O"	document referring to an oral disclosure, use, exhibition or other means		
"P"	document published prior to the international filing date but later than the priority date claimed		

Date of the actual completion of the international search 08 APRIL 2016 (08.04.2016)	Date of mailing of the international search report **08 APRIL 2016 (08.04.2016)**
Name and mailing address of the ISA/KR Korean Intellectual Property Office Government Complex-Daejeon, 189 Seonsa-ro, Daejeon 302-701, Republic of Korea Facsimile No. 82-42-472-7140	Authorized officer Telephone No.

Form PCT/ISA/210 (second sheet) (January 2015)

INTERNATIONAL SEARCH REPORT
Information on patent family members

International application No.
PCT/KR2015/014235

Patent document cited in search report	Publication date	Patent family member	Publication date
KR 10-2014-0052079 A	02/05/2014	AU 2012306232 A1	20/03/2014
		CA 2847076 A1	14/03/2013
		CN 103890214 A	25/06/2014
		EA 201490405 A1	29/08/2014
		EP 2753724 A1	16/07/2014
		EP 2753724 A4	23/09/2015
		FI 20110291 A	08/03/2013
		JP 2014-526613 A	06/10/2014
		KR 10-2015-0046391 A	29/04/2015
		MX 2014002714 A	30/07/2014
		TW 201319275 A	16/05/2013
		US 2014-0219856 A1	07/08/2014
		WO 2013-034804 A1	14/03/2013
JP 2010-196103 A	09/09/2010	JP 5421615 B2	19/02/2014
KR 10-2014-0080347 A	30/06/2014	KR 10-1554771 B1	21/09/2015
JP 2005-154890 A	16/06/2005	CN 100372961 C	05/03/2008
		CN 1788100 A	14/06/2006
		JP 4498847 B2	07/07/2010
		KR 10-1177540 B1	28/08/2012
		KR 10-2006-0099388 A	19/09/2006
		TW 200521251 A	01/07/2005
		TW I268960 B	21/12/2006
		WO 2005-045082 A1	19/05/2005
KR 10-2014-0082491 A	02/07/2014	KR 10-1419878 B1	15/07/2014