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(54) Title: PROCESS FOR PRODUCING HIGH STRENGTH FORMABLE STEEL AND HIGH STRENGTH FORMABLE STEEL PRODUCED THEREWITH

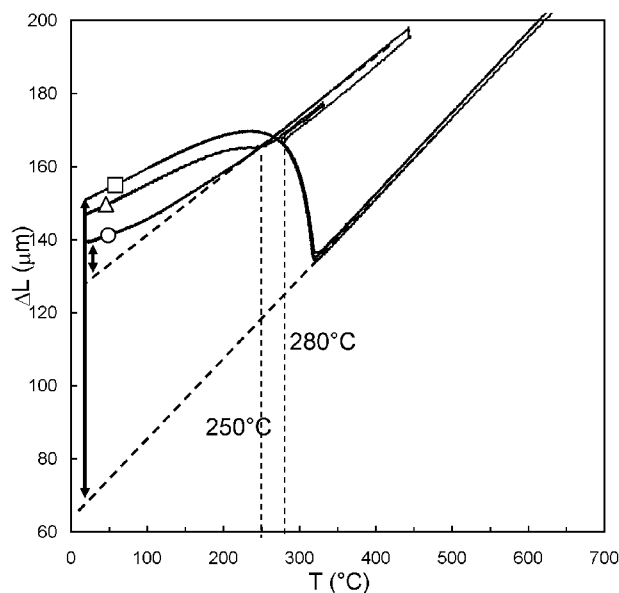


Figure 2(b)

(57) Abstract: The invention relates to a method for producing a steel strip composite and to a steel strip composite produced by said method.



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Process for producing high strength formable steel and high strength formable steel produced therewith.

FIELD OF THE INVENTION

5 The invention relates to a method for producing a steel strip composite and to a steel strip composite produced by said method.

BACKGROUND OF THE INVENTION

Enhancement of ductility at elevated strength is desirable for widespread markets. In the automotive industry in particular, where legislation is driving improvements in fuel economy and safety, there is a move towards stronger, formable high strength steels. High strength and ultrahigh strength strip steel provides automotive manufacturers potential for down weighting of the body in white and the opportunity for countering weight increases arising from the move to electric and hybrid vehicles. In addition high and ultrahigh strength steels play a critical role in determining the crash worthiness of modern passenger vehicles.

Extensive application of high strength and ultrahigh strength steel requires, in many cases, levels of formability that are higher than can be expected for conventional carbon-manganese steels. Enhancement of residual ductility in formed parts is beneficial for integrity in crash. As a first approximation tensile elongation may be considered a simple measure of both formability and impact integrity. Consequently, such effort has been given to developing advanced high strength steels (AHSS) with optimised tensile ductility.

In the case of current commercial AHSS, enhancement of ductility is realised either by using composite effects (Dual Phase or DP steels) or by using deformation induced transformation of retained austenite (TRIP steels). Both composite hardening and transformation induced plasticity mechanisms can measurably enhance the ductility of predominantly ferritic steels. However, the enhancement of ductility which may be achieved in this way is limited to around 1.5 (for DP) to 2 (for TRIP) times that of conventional C-Mn steels at equivalent strength. Furthermore, the strength level that may be achieved in commercial, ferritic-matrix DP and TRIP steels is also limited (TRIP to around 800MPa and DP to around 1000MPa).

Extreme enhancement of tensile ductility may also be achieved by utilising high alloy (predominantly manganese) additions to stabilise austenite (Twinning Induced Plasticity or TWIP steels). Austenite is inherently more ductile than ferrite and twinning provides a very effective work hardening mechanism. Such steels may achieve very high

elongations (typically 50%) at very high strengths (typically 1000MPa). The improvement in elongation may be typically 5x that of a conventional C-Mn steel. However, the yield strength is comparatively low and large strains need to be uniformly imposed to achieve high strength in the formed component. Furthermore, the extremely high levels of alloy
5 make large scale production over conventional process problematic.

Practical experience gained during the implementation of these first and second generations of AHSS have revealed that forming and performance parameters beyond those measured in the tensile test can represent significant barriers to implementation. In particular, although exhibiting high uniform and total elongations, and consequently
10 high resistance to necking during stretch forming, mixed microstructures comprising a distribution of hard phase in a matrix of soft ferrite may be highly susceptible to cracking at stretched edges. Low stretched edge formability is known to impose a practical limit to the formability of otherwise highly ductile AHSS.

A new generation of AHSS that exhibits an advantage in strength ductility-balance with respect to carbon manganese steels but at lower levels of alloy, and consequently lower cost and greater processability, is emerging. For this generation of steels more attention is also being paid to other formability parameters such as stretched edge ductility (hole expansivity) and bendability. In most cases, mixed microstructures comprising hard majority phases such as martensite or bainite are used to develop high
15 to ultrahigh strengths. Such hard, uniformly fine, microstructures tend to exhibit good hole expansivity. Alloy compositions fall into two basic categories: Medium Manganese compositions (Mn typically 7wt%) and Medium Carbon, carbon-manganese steels with additions of Si and or Al.

In the case of the high Mn TRIP steels austenite is simply chemically stabilised by
25 large additions of Mn. Although clearly differentiated strength-ductility combinations can be achieved in this way, processing has proven to be difficult since hard, brittle martensitic phases are developed in intermediate product. This renders further processing, such as cold rolling, difficult at dimensions relevant to commercial processing. Furthermore, alloy costs are high due to the high manganese content.

Therefore the problem arises that it is difficult to provide a steel strip having
30 improved strength and edge ductility.

It is an object of this invention to provide a method for improving the strength and edge ductility of steels.

It is a further object of this invention to provide a steel having improved strength
35 and ductility in combination with an low alloy content.

SUMMARY OF THE INVENTION

According to a first aspect of the invention there is provided a method for producing a quenched and partitioned steel by providing a cold rolled and annealed steel strip containing (in weight %):

- 0.18 - 0.4% C
- 5 • 1.5 - 4.0% Mn
- 0.5 - 2.0% Si
- 0 - 1.5% Al
- 0 - 0.5% Mo
- 0 - 0.5% Ti
- 10 • 0 - 0.4% V
- 0 - 0.010% Nb
- 0 - 0.005% B
- 0 - 0.015% N
- 0 - 0.08% P
- 15 • 0 - 0.01% S
- 0 - 0.06 % Sb
- 0 - 0.05 % Ca
- 0 - 1.0%Cr
- 0 - 1.0% Ni
- 20 • the remainder being iron and unavoidable impurities.

wherein the annealing process comprises the following steps:

- (i) reheating the cold rolled strip to an annealing temperature T_1 of between $A_{c3}-40$ and $A_{c3}+80$;
- (ii) holding the strip at T_1 for an annealing time t_1 of between 10 and 200
25 seconds;
- (iii) cooling the annealed strip at a cooling rate CR1 to a quench temperature T_2 for producing a microstructure in the strip comprising a martensite fraction and a retained austenite fraction;
- (iv) partitioning annealing the cooled strip at a temperature T_3 for enriching the
30 austenite in carbon by repartitioning the carbon from the martensite to the austenite fraction for a repartitioning time t_2 of between 20 and 500 seconds;
- (v) cooling the strip at a cooling rate CR2 to ambient temperature.

According to the invention a cold-rolled strip is provided by the conventional and known processes of casting, hot-rolling and cold-rolling. The casting process may involve
35 thick slab casting (slab thickness between 150 and 350 mm), thin slab-casting (slab thickness below 150 mm, usually between 50 and 100 mm) or even strip casting. Cold rolling is also a conventional and known process.

The composition of the steel according to the invention is a balanced composition comprising medium carbon and moderate manganese additions. The cold rolled steel is heated to form austenite, either partially or fully, followed by quenching to a temperature between the martensite start temperature (M_s) and the martensite finish temperature (M_f), thereby creating a controlled amount of martensite and retained austenite. The manganese additions result in stabilisation of an austenite fraction during cooling from the annealing temperature and the subsequent carbon enrichment further stabilises the austenite fraction. Combined stabilisation via C and Mn additions enables alloying with either element to be restricted to reasonable limits leading to cost and processability advantages. The steel is then subjected to a thermal treatment to partition carbon from the martensite into the austenite. The carbon enrichment of the austenite fraction is achieved by partitioning from martensite or low temperature bainitic transformation or preferably a combination of both. By this partitioning, the formation of carbides is suppressed and the austenite is stabilised rather than decomposed. Combined stabilisation with partitioning and bainitic transformation enables the amount of austenite and the microstructure in which it is embedded to be optimised. The bainitic transformation also leads to carbon enrichment of the remaining austenite because the formation of carbides is suppressed. All compositions are given in weight percentages, unless otherwise indicated. The final microstructure of the steel comprises martensite, bainite and carbon-enriched austenite and, if $T_1 < A_{c3}$, equiaxed ferrite. The stabilisation of austenite results in the steel exhibiting improved ductility relative to traditional high strength steels.

Carbon (C) provides solid solution strengthening, enhances hardenability (thus enabling avoidance of high temperature transformations at cooling rates achievable in conventional annealing lines) and, when dissolved in austenite, promotes the retention of austenite at room temperature. Above 0.4wt% C the propensity for formation of brittle high carbon martensite increases.

Manganese (Mn) delivers substantial solid solution strengthening, stabilises austenite, thus promoting its retention at room temperature, and enhances hardenability promoting the formation of hard transformation products at cooling rates achievable in conventional annealing lines. A preferable upper limit for the manganese content is 3.5%.

Silicon (Si) addition provides solid solution strengthening thus enabling the attainment of high strength and promotes the stabilisation of austenite. Si very effectively retards the formation of carbides during overaging thus keeping carbon in solution for stabilisation of austenite. Ferrite and hard phases such as bainite and martensite exhibit improved ductility in the absence of carbides. Free carbon, not trapped in carbides, may be partitioned to austenite. For acceptable coatability the imposed addition of Si should be below a certain level: Si may be added in the range of between 0.5 to 1wt% when

combined with Al addition. In the absence of Al additions beyond the level needed for deoxidation Si should be maintained in the range of between 1 to 2wt%.

Aluminium (Al) is usually already added in small quantities of at most 0.1% and preferably at most 0.05% to liquid steel for the purpose of deoxidation by forming alumina.

5 Ideally, the total aluminium content in the steel is between 0.01 and 0.08% if aluminium is only added for desoxidation. In the right quantity it also provides an acceleration of the bainite transformation. Al also retards the formation of carbides thus keeping carbon in solution for partitioning to austenite and promoting the stabilisation of austenite. A maximum value of 1.5wt% is imposed for castability purposes because higher Al-contents
10 lead to poisoning of casting mould slag and consequently an increase in mould slag viscosity leading to incorrect heat transfer and lubrication during casting. Aluminium alone delivers low strength. Therefore, if Al is used above levels required for deoxidation it should always be in combination with Si.

Niobium (Nb), if added, is added in small amounts of up to 0.1% or more preferably
15 of up to 0.05wt%. It is added for austenitic grain refinement during hot rolling. If combined with a suitable rolling schedule, it promotes transformation on the run-out-table, and thus finer and a more homogeneous grain size in the hot-rolled intermediate product which is then subsequently cold-rolled.

Titanium (Ti) can be used to form fine precipitates in the ferritic component of
20 mixed microstructures thus increasing strength and promoting uniformity of strength at the microstructural scale and in turn good stretched edge ductility. Preferably Ti is below 0.1%.

Molybdenum (Mo) is used to increase hardenability thus retarding the formation of
25 high temperature transformation products on cooling to the quench temperature. Additions of Mo may therefore allow the utilisation of lower cooling rates more readily achieved in full-scale production lines or to achieve the desired structure and properties in heavier gauge strip. The use of Mo is to be preferred over the use of Cr when UHS is required since in addition to enhancing hardenability Mo provides additional solid solution strengthening. Mo is also known to retard coarsening of fine strengthening precipitates thus promoting
30 thermal stability precipitation strengthening variants. For reasons of cost less than 0.5wt% is preferable.

Vanadium (V) is used to increase hardenability thus retarding the formation of high
temperature transformation products on cooling to the quench temperature. Furthermore V may combine with C or N or both to form fine strengthening precipitates thus increasing
35 strength and promoting uniformity in strength at the microstructural scale in mixed microstructures and in turn improved stretched edge ductility. V addition up to 0.4wt% is effective. Higher additions are undesirable for reasons of cost and because excessive levels

of precipitation tie up high amounts of C. Since free C is required for austenite stabilisation too high V requires increased C addition. Preferably V is below 0.1%.

Phosphorus (P) is used to suppress the formation of pearlite during cooling, to suppress carbide formation and thereby promote the partitioning of carbon to austenite resulting in austenite stabilisation. However, too high a P addition is known to cause embrittlement at hot-rolling temperatures and to lead to reduced toughness in martensitic UHSS. P may also lead to problems in spot welding of the final product. For these reasons P is limited to a maximum of 0.08wt% and more preferably to a maximum of 0.02wt%.

Sulphur (S) is an impurity which may embrittle the intermediate or final product. S should be limited to a maximum level of 0.01wt% and more preferably to at most 0.005wt%.

Chromium (Cr) and nickel (Ni) may be used to increase hardenability thus retarding the formation of high temperature transformation products on cooling to the quench temperature. Additions of Cr and/or Ni may therefore allow the utilisation of lower cooling rates more readily achieved in full-scale production lines or to achieve the desired structure and properties in heavier gauge strip. Cr and/or Ni should be limited to a level of less than 1wt% for reasons of cost and preferably to levels of 0.5wt% or less and more preferably below 0.1%.

Boron (B) may be used to improve hardenability and, in particular, to prevent the formation of ferrite on cooling from a fully austenitic soaking temperature. B should be limited to 50ppm because above these levels further addition is ineffective.

Antimony (Sb) may be used to enhance the wettability of zinc during hot dip galvanising. Addition should be limited to 0.06wt% or less.

Calcium (Ca) may be required to avoid clogging during casting and may be beneficial for modifying the morphology of MnS inclusion. Globular inclusions are known to be less detrimental to stretched edge ductility than highly elongated inclusions. If used, then addition should preferably be made to the level of 30ppm or less. Additions above 0.05% lead to the formation of coarse inclusions detrimental to ductility and toughness.

Nitrogen (N) may be used, in combination with V, to form fine strengthening precipitates imparting strength and promoting uniformity in strength at the microstructural scale and in turn good stretched edge ductility. Additions should be limited to 150ppm or less.

The annealing step starts with reheating to an annealing temperature (T_1). T_1 may be above or below A_{c3} . If T_1 is below A_{c3} the resulting ferrite fraction leads to a lower strength and may introduce heterogeneity in strength at a microstructural scale. Local differences in strength lower the ratio of yield to tensile strength and lead to poor stretched

edge ductility and bendability. It is therefore preferable that T_1 is above Ac_3 . In the context of this invention the phrase "above Ac_3 " means that the microstructure is austenitic at T_1 . Although the equilibrium transformation temperature Ae_3 is only determined by the composition, the value of the corresponding Ac_3 temperature is not a constant value as its value depends among others on the heating rate during which Ac_3 is measured and the starting microstructure of the steel. Usually Ac_3 is determined using dilatometry. When the heating rate used during dilatometry and the microstructure of the test specimen are those used in the process according to the invention, the value of Ac_3 is easy to determine. In determining T_1 and the allowable variation around T_1 the narrowing of the transformation temperature range with increasing C content should preferably be taken into account as follows:

$(Ac_3-40) < T_1 \leq (Ac_3+50)$ for carbon between 0.18 - 0.3% C or

$(Ac_3-20) < T_1 \leq (Ac_3+30)$ for carbon between 0.3 to 0.4% C

The rapid cooling to T_2 (CR1) is required to avoid the formation of high temperature transformation phases. The specific rate required depends upon the steel chemistry and corresponds to the critical cooling rate for avoidance of ferrite and pearlite noses in the relevant CCT diagram. In addition, the critical rate decreases with increasing T_1 above Ac_3 . Preferably CR1, i.e. the cooling rate over the temperature interval 800-500°C from T_1 to T_2 is between 30 to 80 °C/s. T_2 should be chosen low enough to deliver partial transformation to martensite, but not so low as to cause complete transformation to martensite. T_2 is chosen to deliver a volume fraction of martensite of between 50 to 90% (in volume) and preferably an austenite fraction of at least vol.5%. The dependency of the martensitic transformation start temperature (M_s) on composition means that T_2 will also depend upon chemistry. Preferably a martensite fraction of 60 - 85 vol.% is chosen.

Holding at T_3 is needed to enrich the remaining austenite in carbon via a bainitic transformation or carbon partitioning or both. This temperature may be the same as the quench stop temperature ($T_2=T_3$) or may be higher ($T_3>T_2$). Higher isothermal holding temperatures may be advantageous since increased rates of carbon diffusion may make feasible shorter isothermal holds. Several processes occur during isothermal holding including tempering of martensite, diffusion of carbon from the martensite fraction to the remaining austenite, the precipitation of carbides and the formation of bainite. The presence of a small fraction of martensite is known to accelerate subsequent transformation to bainite so, for some combinations of alloy composition and processing capability, quench stop temperatures under M_s are desirable. T_3 is chosen so as to give a suitable rate of transformation to bainite or rate of partitioning or both. The specific temperature will be dependent upon alloy composition and will preferably fall in a range expressed by:

$(M_s - 70) < T_3 \leq (M_s + 150)$, and preferably $T_3 \leq (M_s + 50)$

The correct balance of isothermal holding temperature and isothermal holding time must be chosen for each composition. These can be determined by means of dilatometry as described hereinbelow.

5 The formation of carbides may lock up carbon which would otherwise be available for stabilisation of austenite and should therefore preferably be avoided. Furthermore, coarse Fe_3C carbides may lead to a deterioration in tensile ductility and/or stretched edge ductility. The levels of Si or Si/Al must be suitably tuned to retard carbide formation for the duration of the isothermal hold.

10 Since both the carbon partitioning accompanying the martensite tempering and bainitic transformation, in the presence of sufficient Si or Si/Al, each deliver an enrichment of carbon in austenite, exploitation of both mechanisms in tandem is highly beneficial leading to a greater total degree of austenite stabilisation.

15 It should also be noted that the optimum properties are not delivered by simply tuning process to deliver the maximum stabilisation of austenite. Levels of carbon enrichment too low to completely stabilise all austenite but sufficient to deliver high carbon austenite prior to the final cool can lead to the formation of brittle martensite. Excessive tempering of martensite can lead to significant loss in strength and, in combination with insufficient stabilisation of austenite and the formation of new hard martensite on final
20 cooling, to low ratios of the yield to tensile strength.

 Therefore, it is preferred that the partitioning temperature and time are chosen such as to optimise the enrichment of carbon in the austenite but without creation of deleterious microstructures during the isothermal hold.

25 Subsequent to the hold at T_3 the strip is cooled to ambient temperature. The strip may also be coated with zinc or other such metallic layers using a suitable method of deposition either in-line or in a following process step.

 In a preferable embodiment the cold rolled strip contains at least 0.25% C and at least 0.03% Al. A lower limit of 0.25wt% is placed on C because below this level the desired combinations of strength and ductility may not be achieved.

30 In an embodiment the use of Si is preferred to that of Al such that a minimum silicon content of 1wt% and a maximum aluminium content of 0.5 wt% is defined. Si provides substantial strengthening allowing the achievement of ultra high strength, more effectively suppresses carbide formation enabling longer isothermal holds without formation of large volumes of coarse iron carbides, and because it does not accelerate
35 bainite formation to the same extent as Al thus preventing excessive formation of bainite and enabling higher strengths to be achieved.

Strength-Ductility data for a range of production C-Mn steels including ferritic forming steels and quenched martensitic steels have been used to generate a base-line strength-ductility decay for conventional strip steels. The data conform to the expression:

$$e_{\text{calculated}} = k \frac{1}{\sqrt{UTS^3}} \quad (1)$$

where $e_{\text{calculated}}$ is the total elongation (expressed as % engineering strain), UTS is the ultimate tensile strength and k is a constant which for tensile test pieces with 80 mm gauge and thickness 1 mm is 250000.

If elongations are measured at different gauge or thickness then they must be converted to an equivalent elongation at 80mm gauge and 1mm thick or the above expression must be fit to base-line data measured at that alternative gauge/thickness combination using appropriate values of the constants. Conversion of tensile ductility can be performed using accepted procedures (ISO Norm 2566/1-2) when the geometries corresponding to the measured and to be calculated elongations are know:

$$e_2 = e_1 \left(\frac{L_1}{L_2} \cdot \frac{\sqrt{A_2}}{\sqrt{A_1}} \right)^m \quad (2)$$

Where e_2 is the required elongation for a gauge length of L_2 with a cross section of A_2 , e_1 is the known elongation measured for a gauge length of L_1 with a cross section of A_1 and the exponent m is a material constant here assumed to be equal to 0.4.

In an embodiment the quenched and partitioned steel has an e-ratio of at least 1.8 wherein the e-ratio is defined as $e_{\text{measured}}/e_{\text{calculated}}$ and wherein $e_{\text{calculated}}$ is calculated according to equation (1) and wherein e_{measured} is the elongation measured from an 80mm gauge length sample at 1mm thick (or measured at some other geometry and converted to an equivalent elongation on an 80mm gauge at 1mm thick using expression (2)).

In an embodiment the tensile strength of the steel according to the invention is at least 900 MPa. This strength regime is of interest since it provides significant opportunity for down-gauging and is a strength regime for which formability is most limited. Ductility levels are at least 1.8x or more than that of conventional C-Mn steels at equivalent strength.

Preferably the yield to tensile strength ratio is 0.6 or higher. More preferably the ratio is at least 0.65 or even higher. Low yield to tensile strength ratios are associated with poor bendability and edge cracking sensitivity. Performance is often dependent on yield strength, anti-intrusion components for instance require high yield strength. High yield to

tensile strength ratios ensure strength uniformity in the formed part, especially in forming operations which apply localised strain such as bending, or hole expansion.

In a preferred embodiment of the invention the metal or metal alloy coating is zinc, aluminium, magnesium or alloys thereof. Hereby the steel is afforded sacrificial corrosion protection since the zinc and aluminium will oxidise in preference to iron in the steel.

It was stated above that it is preferred that the partitioning temperature and time are chosen such as to optimise the enrichment of carbon in the austenite but without creation of deleterious microstructures during the isothermal hold. The temperature and time can be determined using dilatometry as follows:

First, fully austenitise a sample and apply a fast cool to room temperature to determine the fraction of martensite as a function of temperature below M_s . This allows to select an appropriate fraction of martensite as a function of the quench temperature. For this selected quench temperature the range of time-temperature partitioning combinations that deliver the correct stabilisation of austenite can be determined as follows. Cool a fully austenitized specimen to the selected quench temperature T_2 to produce the desired fraction of austenite and martensite and subsequently reheat the sample to a selected partitioning temperature T_3 and hold the sample at this temperature for a certain partitioning time t_2 . The value of t_2 may be between 10 and 500 seconds, but for practical purposes in commercial annealing lines t_2 is preferably in the range of 20 to 180 s or even 20 to 100 s. After the partitioning step the specimen can be cooled naturally or acceleratedly to room temperature; no fast quench is needed. If during this cooling the dilatation curve shows that the martensite formation re-starts at a temperature in the range from $(M_s - 20)$ to 120°C , then a correct degree of stabilisation has been achieved. In the case that the martensite formation recommences at a temperature $<120^\circ\text{C}$, it means that the stabilisation is too strong and the martensite has a very high carbon content compared to the bulk concentration. In the other case that the martensite formation recommences at a temperature less than 20°C below M_s the stabilisation is not sufficient. This experiments needs to be repeated for different partitioning times to find the times corresponding the two limiting boundary conditions for the chosen partitioning temperature.

By varying the partitioning temperature T_3 and repeat above the required partitioning time will also vary. Increasing the partitioning temperature results in a decrease in partitioning time t_2 and to a higher degree of tempering of the martensite formed during the quench.

In a preferred embodiment of the invention the metallic coating is provided by hot-dip galvanising or by electro-galvanising.

BRIEF DESCRIPTION OF THE DRAWINGS

The invention will now be elucidated by way of example making reference to the accompanying figures and examples.

Figure 1 shows the schematic annealing schedule indicating the meaning of T_1 , T_2 and T_3 , t_1 and t_2 , and of CR1 and CR2.

Figure 2a and 2b show a set of result of the dilatometric experiments to determine the quench temperature and the partitioning time. Figure 2a shows the temperature as a function of time for a steel having 3.5% Mn which was quenched to a quench temperature of 280°C and reheated to a partitioning temperature of 330°C (triangle) and 440°C (circle). The sample was held at the partitioning temperature for 20 seconds. The quenching temperature resulting in the required amount of martensite is determined on the basis of the base curve (NC-III, square). Figure 2b shows the dilatation of the samples for these conditions. The base curve, with a full quench to room temperature allows to determine M_s (about 315°C). The partitioning step for 20s at 330°C shows no dilation of the sample, which means that no bainite is formed, carbon is partitioned and the martensite is only marginally tempered. The transformation to martensite re-starts at temperatures lower than the quench temperature of 280°C, namely at 250°C, which indicates that the austenite has been stabilized due to carbon partitioning. Increased partition times at 330°C show that the transformation to martensite re-starts at lower temperatures than 250°C. The dilation observed during annealing of the sample for 20s at 440°C means that bainite is formed. Concurrently, carbon is partitioned and the martensite is tempered. The consequence of this high partitioning temperature is that the martensite is severely tempered. Due to the bainitic transformation and the carbon partitioning the austenite is strongly enriched in carbon, which is reflected in the relatively low start temperature of the martensite of approximately 150°C during the second quench after the partitioning at 440°C. The latter martensite is very hard and brittle and therefore undesirable. By varying the partitioning time and the partitioning temperature the optimum combination can be determined.

Table 2 shows the results of various thermal cycles with the steels of Table 1. These results show that (I-VI):

I. The desired property balance is not achieved for processing with C and Mn additions at the level of current commercial TRIP Steels (composition G). G delivers a very large extension of ductility (typically 2.2 times that of a conventional C-Mn steel) at strengths ranging from 850 to 1050MPa. However, this is only true when the annealing temperature T_1 is chosen below the preferred range, ($A_{c3}-40$ to $A_{c3}+40$ such that a high fraction of ferrite is retained in the final structure. From table 2 it is apparent that, in

these cases (cycles 31-32, although ductility is at the desired level, the ratio of YS to UTS drops below the desired level to approximately 0.4.

When annealing is performed with T_1 within the claimed range, both direct quenching to room temperature and direct quenching to room temperature followed by an isothermal hold at a higher temperature (i.e. conventional quenching and tempering as in cycle 25) deliver strengths in the desired range but do not deliver ductility above the desired minimum level.

When annealing is performed with T_1 within the claimed range (cycles 26-30 and 33) strengths in the range 1000-1300 MPa may be achieved but the desired level of ductility is not achieved.

Processing variants annealed at T_1 within the claimed range, subsequently quenched to a low T_2 to give high fractions of martensite (in examples 85-93%) after the quench and followed by an isotherm at some higher temperature (table 3 cycles 28-30) to not deliver the desired level of ductility.

Processing variants annealed at T_1 within the claimed range, subsequently quenched to a low T_2 to give high fractions of martensite after the quench followed by an isotherm at the same temperature (cycle 26) do not deliver the desired level of ductility.

Processing variants annealed at T_1 within the claimed range, subsequently quenched to a high T_2 to give low fractions of martensite (in example 50%) after the quench followed by an isotherm at the same temperature (cycle 27) to not deliver the desired level of ductility.

II. Complete replacement of Si with Al delivers strength and ductility below the desired minima: Composition A when directly quenched to room temperature delivers strength in the desired range but ductility below the desired minimum (cycle 4). Composition A when quenched to a T_2 above the M_s temperature and subjected to isothermal holding at the same temperature delivers strength greatly below the desired range and ductility below the desired minimum (cycle 3). Composition A when quenched to a T_2 below the M_s temperature and subjected to isothermal holding at some higher temperature delivers strengths in the approximate range 950 to 1000MPa and ductilities below the desired minimum (thermal cycles 1-2).

III. Addition of C and/or Mn to levels beyond that found in current commercial trip steels does not enable the desired strength ductility balance to be achieved with conventional quench and temper process. For compositions B, C, D, E, F, and H utilising a direct quench or a direct quench followed by isothermal holding at some higher temperature (i.e. conventional quench and temper) delivers strengths in the desired range but ductility lower than the required minimum (cycles 9, 14, 25, 19, 24, 38).

IV. Addition of C and/or Mn enables the desired strength ductility balance to be achieved when T_2 is above the preferred range but only when the duration of the isothermal hold is unacceptably long for conventional process. Compositions B, C, D, F and H each enable the desired property range to be achieved even if T_2 is set such that no martensite is formed during the initial quench (cycles 8, 12, 13, 16, 17, 36 and 37) but in all cases the isothermal holding time at T_3 is unacceptably long to be practical or economical in a continuous annealing process.

V. Addition of C and/or Mn enables the desired strength ductility balance to be achieved when processing is carried out in the preferred manner using a quench temperature (T_2) below M_s and a suitable combination of isothermal holding temperature (T_3) and time. Compositions B, D and F each enable the desired property range to be achieved if T_2 is set such that the desired fraction of martensite is formed during the initial quench (cycles 6, 15, 22) and if an isothermal holding temperature (T_3) and the holding time at this temperature are set at suitable levels.

VI. C addition of higher than 0.4wt% leads to embrittlement when processing is carried out according to the preferred route. Compositions C and H each return effectively zero ductility results when subjected to processing involving a deep quench (low T_2) and isothermal hold times at T_3 sufficiently short for conventional CA process (cycles 10-11 and 34-35) due to the formation of brittle, high carbon martensite in the final structure.

Table 1 - Alloy Compositions (Ca-treated).

	C %	Mn %	Si %	Al _{tot} %	B ppm	Ti %	N ppm	P %	S %	Nb %	Cr %	Mo %	Ni %	Cu %	V %	Sn %
A	0.228	1.68	0.01	1.42	<2	0.001	55	0.004	0.007	<0.002	0.019	0.009	0.015	0.008	0.002	0.003
B	0.310	1.66	1.47	0.04	<2	0.002	63	0.004	0.008	<0.002	0.017	0.003	0.015	0.011	0.002	0.003
C	0.413	1.67	1.48	0.03	<2	0.002	67	0.003	0.009	<0.002	0.016	0.003	0.015	0.011	0.002	0.003
D	0.288	2.39	1.76	0.02	<2	0.002	59	0.003	0.007	<0.002	0.016	0.003	0.015	0.011	0.003	0.003
E	0.204	1.72	1.43	0.02	22	0.029	54	0.003	0.006	<0.002	0.039	0.004	0.016	0.011	0.003	0.003
F	0.195	3.50	1.54	0.01	<2	0.002	62	0.006	0.006	0.001	0.036	0.004	0.080	0.018	0.018	0.003
G	0.200	1.65	1.53	0.02	<2	0.006	51	0.011	0.004	0.001	0.026	0.002	0.024	0.014	0.005	0.006
H	0.582	1.54	1.42	0.03	<2	0.002	84	0.004	0.010	<0.002	0.017	0.005	0.016	0.012	0.001	0.003
I	0.204	1.72	1.43	0.02	<2	0.002	54	0.003	0.006	0.029	0.039	0.004	0.016	0.011	0.003	0.003
J	0.228	1.68	0.61	1.42	<2	0.001	55	0.004	0.007	<0.002	0.019	0.009	0.015	0.008	0.002	0.003

*(black cells not according to invention)

Table 2 – Processing parameters and mechanical properties

	Cycle	Ac3 (°C)	Ms (°C)	T1 (°C)	T2 (°C)	T3 (°C)	t2 (s)	YS (MPa)	UTS (MPa)	YS/UTS (-)	e_ratio (-)	Comparative / Inventive	Comment
A	1	1000	420	1080	240	400	60	766	952	0.80	0.6	C	Ductility low
A	2	1000	420	1080	320	400	60	809	1004	0.81	0.6	C	Ductility low
A	3	1000	420	1080	430	430	60	434	573	0.76	0.7	C	Ductility & Strength low
A	4	1000	420	1080	20	NA	NA	925	1335	0.69	1.1	C	Ductility low
B	5	855	330	880	200	350	60	734	1324	0.55	1.9	C	YS/UTS Low
B	6	855	330	880	220	450	60	849	1137	0.75	1.9	I	Inventive Example
B	7	855	330	920	370	370	480	543	1018	0.53	2.2	C	YS/UTS Low
B	8	855	300	820	370	370	600	588	926	0.63	2.1	C	t ₂ too long
B	9	855	330	920	20	450	300	1317	1363	0.97	1.1	C	Ductility low
C	10	825	300	860	180	320	60	879	1344	0.65	0.2	C	Ductility low
C	11	825	260	800	200	400	60	606	1098	0.55	0.3	C	Ductility low
C	12	825	300	860	340	340	900	1012	1414	0.72	2.4	I	t ₂ too long
C	13	825	260	800	350	350	900	975	1305	0.75	3.5	I	t ₂ too long
C	14	825	300	860	20	NA	NA	833	954	0.87	0.0	C	Ductility Low
D	15	855	325	900	200	350	60	1196	1576	0.76	1.8	I	Inventive Example

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	Cycle	Ac3 (°C)	Ms (°C)	T1 (°C)	T2 (°C)	T3 (°C)	t2 (s)	YS (MPa)	UTS (MPa)	YS/UTS (-)	e_ratio (-)	Comparative / Inventive	Comment
D	16	855	300	810	370	370	900	860	1273	0.68	2.5	I	t ₂ too long
D	17	855	325	880	360	360	900	770	1179	0.65	1.9	I	t ₂ too long
D	18	855	325	900	20	450	300	1306	1420	0.92	1.2	C	Ductility low
E	19	895	370	930	20	450	300	1153	1193	0.97	1.0	C	Ductility low
E	20	895	370	930	410	410	180	809	1169	0.69	1.5	C	Ductility low
F	21	840	315	870	270	450	20	780	1426	0.55	1.4	C	Ductility low
F	22	840	315	870	270	350	20	1057	1565	0.68	1.8	I	Inventive Example
F	23	840	315	870	360	360	900	790	1500	0.53	1.7	C	Ductility low
F	24	840	315	870	20	NA	NA	1220	1709	0.71	1.6	C	Ductility low
G	25	895	380	900	20	350	60	987	1269	0.78	1.1	C	Ductility low
G	26	895	380	900	250	250	60	934	1461	0.64	1.5	C	Ductility low
G	27	895	380	900	350	350	60	929	1335	0.70	0.7	C	Ductility low
G	28	895	380	900	240	350	60	1078	1298	0.83	1.0	C	Ductility low
G	29	895	380	900	240	450	60	1063	1176	0.90	1.2	C	Ductility low
G	30	895	380	900	280	400	60	1105	1246	0.89	1.0	C	Ductility low
G	31	895	350	850	240	350	60	431	1020	0.42	2.2	C	YS/UTS Low
G	32	895	350	850	280	400	60	421	876	0.48	2.2	C	YS/UTS Low
G	33	895	380	900	420	420	60	754	986	0.76	1.4	C	Ductility low
H	34	795	230	830	100	300	60	1095	1208	0.91	0.1	C	Ductility low
H	35	795	230	780	120	300	60	763	854	0.89	0.0	C	Ductility low
H	36	795	230	830	320	320	1800	1126	1524	0.74	2.9	I	t ₂ too long
H	37	795	230	780	320	330	1800	1145	1437	0.80	3.4	I	t ₂ too long
H	38	795	230	830	20	450	300	1591	1645	0.97	1.4	C	Ductility low

*(black cells not according to invention)

CLAIMS

1. A method for producing a quenched and partitioned steel by providing a cold rolled and annealed steel strip containing (in weight %):

- 0.18 - 0.4% C
- 1.5 - 4.0% Mn
- 0.5 - 2.0% Si
- 0 - 1.5% Al
- 0 - 0.5% Mo
- 0 - 0.5% Ti
- 0 - 0.4% V
- 0 - 0.010% Nb
- 0 - 0.005% B
- 0 - 0.015% N
- 0 - 0.08% P
- 0 - 0.01% S
- 0 - 0.06 % Sb
- 0 - 0.05 % Ca
- 0 - 1.0%Cr
- 0 - 1.0% Ni
- the remainder being iron and unavoidable impurities.

wherein the annealing process comprises the following steps:

- (i) reheating the cold rolled strip to an annealing temperature T_1 of between $A_{C3}-40$ and $A_{C3}+80$;
- (ii) holding the strip at T_1 for an annealing time t_1 of between 10 and 200 seconds;
- (iii) cooling the annealed strip at a cooling rate CR1 to a quench temperature T_2 for producing a microstructure in the strip comprising a martensite fraction and a retained austenite fraction;
- (iv) repartitioning annealing the cooled strip at a temperature T_3 for enriching the austenite in carbon by repartitioning the carbon from the martensite to the austenite fraction for a repartitioning time t_2 of between 20 and 500 seconds wherein the martensite fraction is between 50 and 90% of the microstructure at the start of the repartitioning annealing;
- (v) cooling the strip at a cooling rate CR2 to ambient temperature.

2. A method according to claim 1 wherein the martensite fraction is at least 60 and/or at most 85% of the microstructure at the start of the repartitioning annealing.

3. A method according to any one of the preceding claims wherein

$(Ac_3-40) < T_1 \leq (Ac_3+50)$ for carbon between 0.18 - 0.3% C or

$(Ac_3-20) < T_1 \leq (Ac_3+30)$ for carbon between 0.3 to 0.4% C

4. A method according to any one of the preceding claims wherein T_1 is above Ac_3 .
5. A method according to any one of the preceding claims wherein the cold rolled strip contains at most 0.5% Al.
6. A method according to any one of the preceding claims wherein the cold rolled strip contains at least 0.25% C and at least 0.01% Al.
7. A method according to any one of the preceding claims wherein T_2 equals T_3 .
8. A method according to any one of claims 1 to 6 wherein T_2 is lower than T_3 .
9. A method according to any one of the preceding claims wherein $(Ms-70) < T_3 \leq (Ms+150)$, preferably wherein $(Ms-70) < T_3 \leq (Ms+50)$
10. A method according to any one of the preceding claims wherein the cold-rolled and annealed strip is coated with one or more metallic layer(s), preferably wherein the step of providing the metallic coating is by hot-dip galvanising or electro-galvanising.
11. Steel strip produced by any one of the claims 1 to 10 wherein the steel has a microstructure containing at least 5% austenite and the e-ratio of at least 1.8 and a tensile strength (UTS) of at least 900 MPa.
12. Steel strip according to claim 11 wherein the microstructure further contains one or more of bainite, martensite, tempered martensite, ferrite, fine carbides.
13. Steel strip according to claim 11 or 12 wherein the microstructure does not contain ferrite and/or coarse cementite.
14. Steel strip according to any one of claims 11 to 13 wherein the yield strength to tensile strength ratio is at least 0.6.

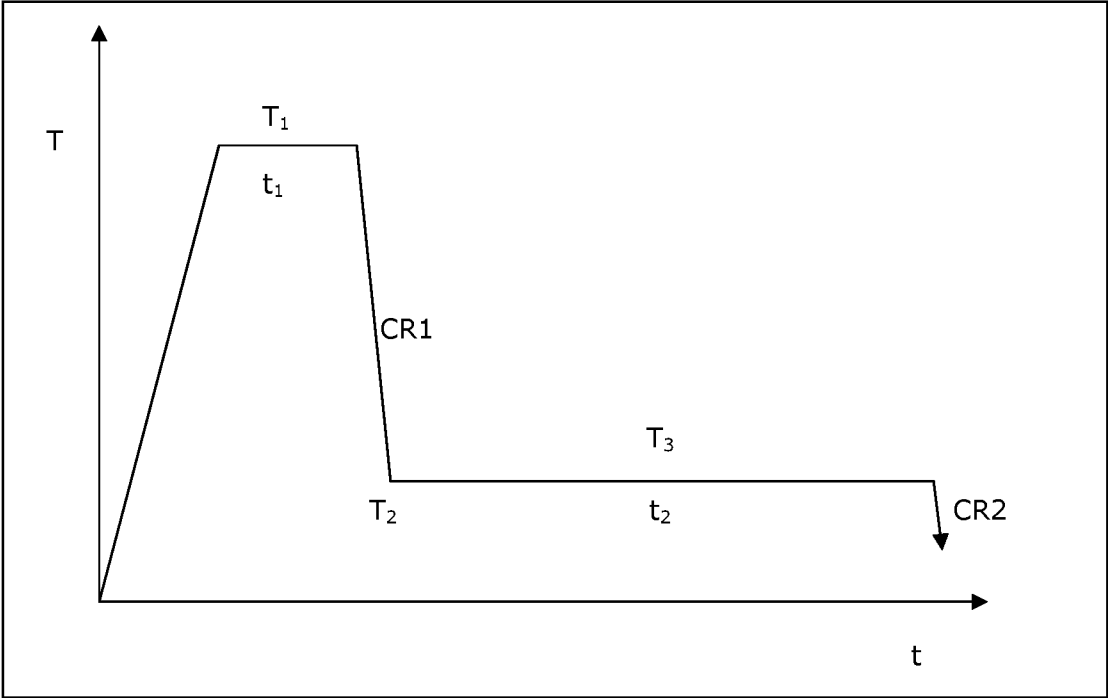


Figure 1

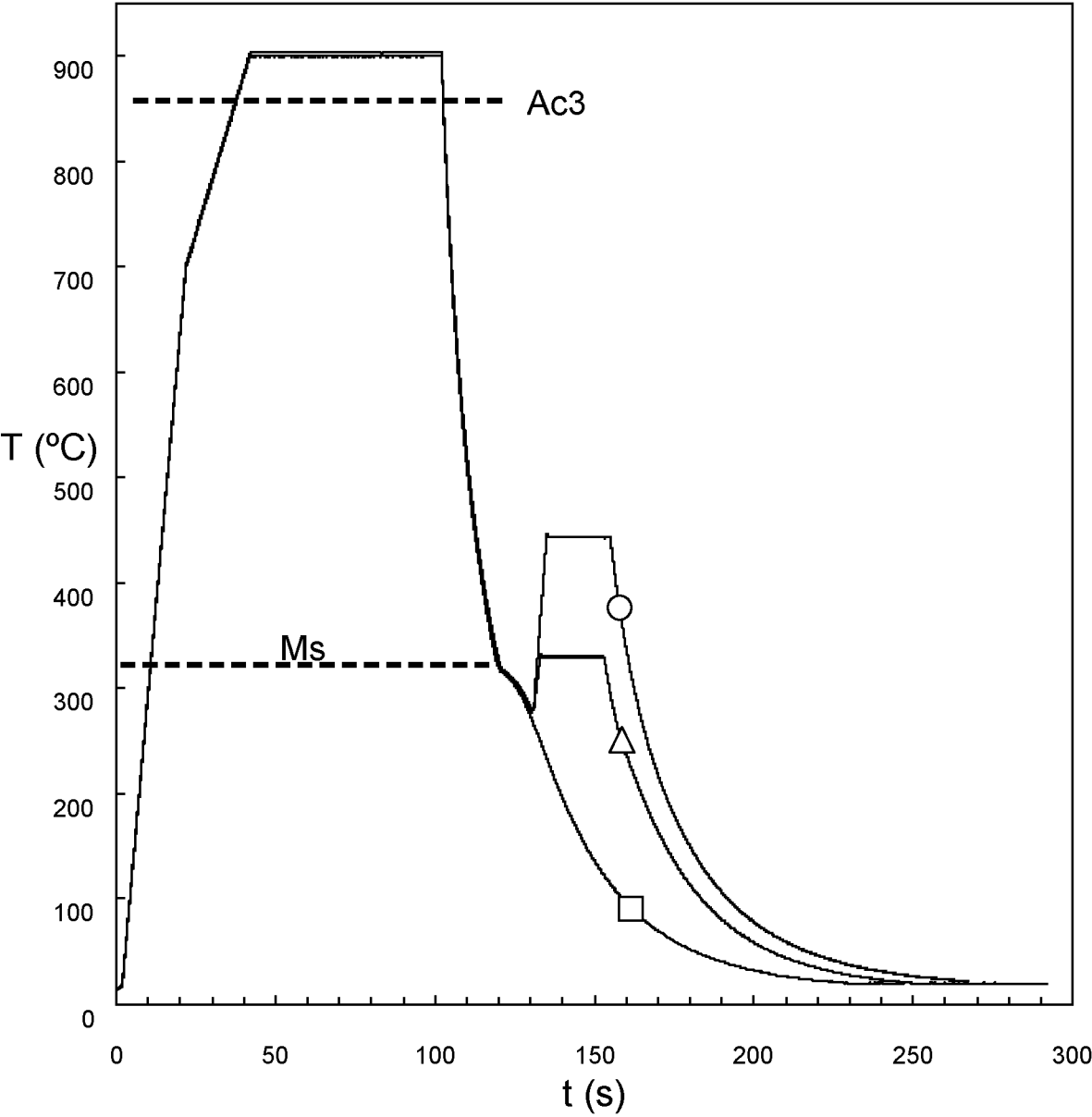


Figure 2(a)

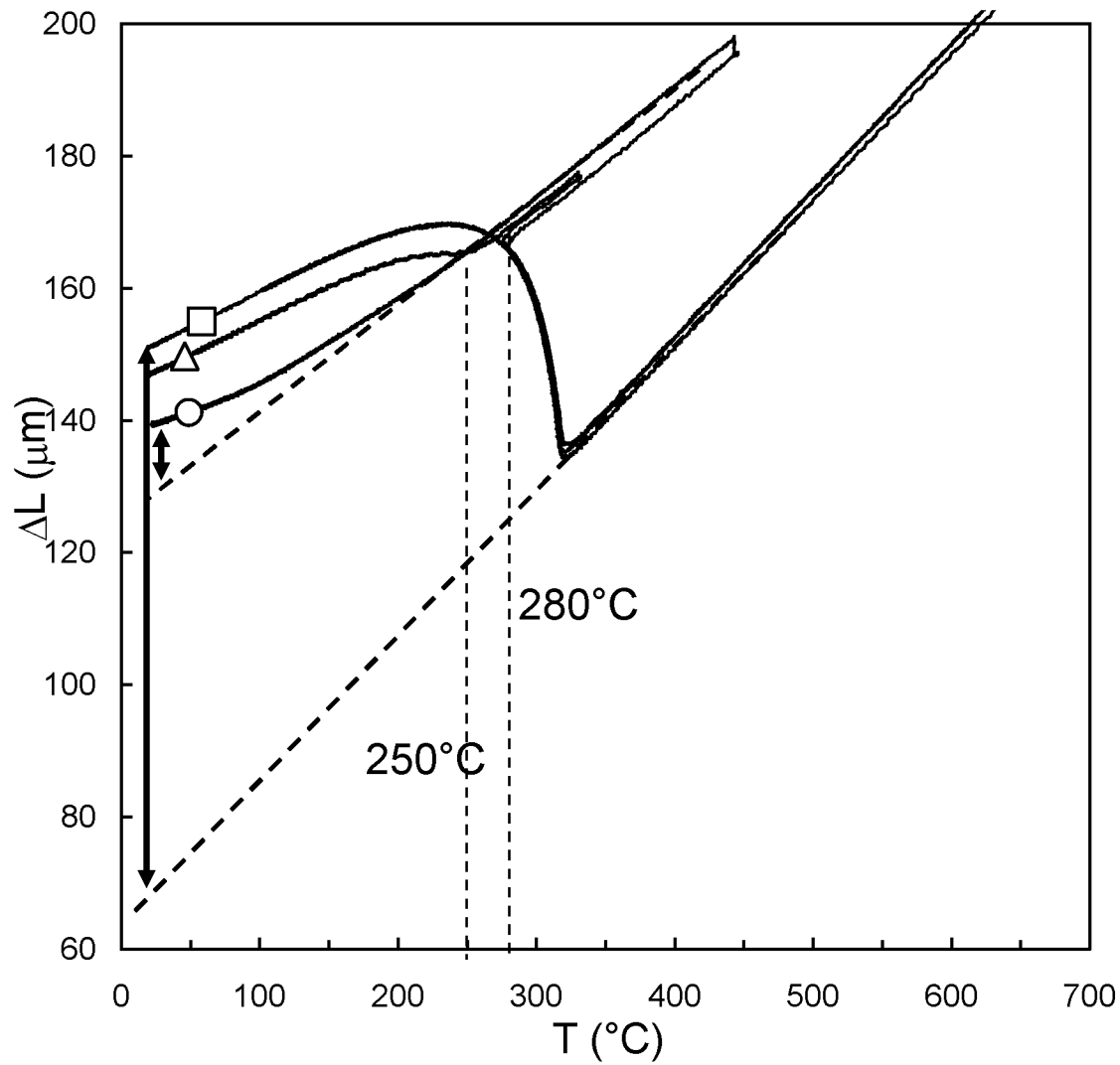


Figure 2(b)

INTERNATIONAL SEARCH REPORT

International application No
PCT/EP2012/053856

A. CLASSIFICATION OF SUBJECT MATTER

INV. C21D8/02 C21D9/46 C22C38/04 C23C2/06 C23C2/28
C23C2/40

ADD.

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)

C21D C22C C23C

Documentation searched other than minimum documentation to the extent that such documents are included in the fields searched

Electronic data base consulted during the international search (name of data base and, where practical, search terms used)

EPO-Internal, WPI Data, CHEM ABS Data, INSPEC

C. DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
X	WO 2010/029983 A1 (JFE STEEL CORP [JP]; MATSUDA HIROSHI [JP]; FUNAKAWA YOSHIMASA [JP]; TA) 18 March 2010 (2010-03-18) abstract -----	1-14
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Further documents are listed in the continuation of Box C.



See patent family annex.

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"O" document referring to an oral disclosure, use, exhibition or other means

"P" document published prior to the international filing date but later than the priority date claimed

"T" later document published after the international filing date or priority date and not in conflict with the application but cited to understand the principle or theory underlying the invention

"X" document of particular relevance; the claimed invention cannot be considered novel or cannot be considered to involve an inventive step when the document is taken alone

"Y" document of particular relevance; the claimed invention cannot be considered to involve an inventive step when the document is combined with one or more other such documents, such combination being obvious to a person skilled in the art.

"&" document member of the same patent family

Date of the actual completion of the international search

26 March 2012

Date of mailing of the international search report

02/04/2012

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INTERNATIONAL SEARCH REPORT

International application No

PCT/EP2012/053856

C(Continuation). DOCUMENTS CONSIDERED TO BE RELEVANT

Category*	Citation of document, with indication, where appropriate, of the relevant passages	Relevant to claim No.
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