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(54) METHOD FOR HOT-FORMING A STEEL BLANK

(57) The invention relates to a method for hot forming a steel blank into an article having enhanced mechanical properties, such as an automotive part with improved ductility and impact toughness, and to a hot formed article obtained by said method.

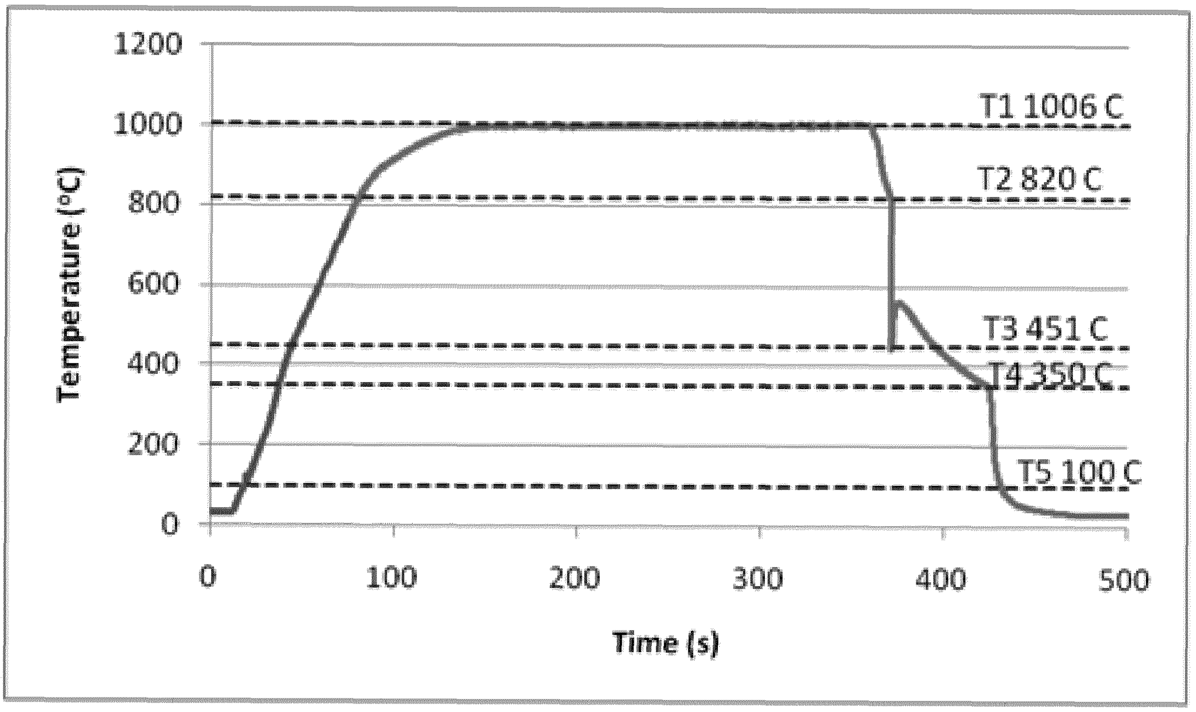

FIG 1

EP 3 327 152 A1

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Description

[0001] The invention relates to a method for hot forming a steel blank into an article having enhanced mechanical properties, such as an automotive part with improved ductility and impact toughness, and to a hot formed article obtained by said method.

[0002] Recent progress in developing advanced high strength steels (AHSS) and ultra high strength steels (UHSS) enables automotive manufacturers to increase crashworthiness of automotive body parts and to reduce the weight through down gauging. However, cold formed AHSS and UHSS are still limited by formability, while cold and hot formed AHSS and UHSS are still limited by ductility and impact toughness.

[0003] In attempts to solve the formability problem, hot forming (also known as hot stamping, press-hardening and die-quenching) has been developed. The basics of the hot forming technique and steel compositions adapted to be used were for the first time described in GB1490535.

[0004] In a typical hot forming process using a boron steel composition the blank is furnace-heated and austenised at 900-950 °C, transferred from furnace to forming tool, and stamped into the desired part geometry. The blank has an ultimate tensile strength R_m<200 MPa and a total elongation A>50%. The formed blank is finally die-quenched to 100-200 °C with a mean cooling rate of >30 °C/s and a homogenous martensitic microstructure is obtained, while it is constrained in the forming tool. The final martensitic part typically exhibits a proof strength $Rp_{0.2}$>1100 MPa, an ultimate tensile strength R_m>1500 MPa and a total elongation A<8%. Hot stamped boron steel parts typically include anti-intrusive structural body parts such as roof pillar-, door beam- and bumper beam-reinforcements that constitute the 'safety-cell'. The relatively soft and ductile high-temperature austenitic microstructure during forming permits down-gauging and lightweighting while not compromising forming limits and moreover, permits part-consolidation and in turn, increased structural strength and increased process efficiency compared to cold forming as joining/welding is reduced. Forming above the ferrite recrystallization temperature, the martensitic transformation, which releases forming stresses and geometric constraint on the part during quenching, eliminates springback giving rise to geometric accuracy. The ultra high strength martensitic microstructure resulting in the final part permits down-gauging (lightweighting) simultaneously while improving anti-intrusive crashworthiness compared to the use of cold formed parts. The supremacy of boron steel in hot stamping applications owes to the (quench) hardenability (owing to the boron addition) and in turn ultra high strength martensitic microstructure that can be obtained from the relatively lean chemical composition. However, while hot stamped martensitic boron steel provides excellent mechanical properties for anti-intrusive crashworthiness, the limited ductility and in turn limited toughness give rise to very poor impact-energy absorptive crashworthiness.

[0005] Therefore, hot stamped boron steel and the advantages of the hot stamping process cannot be exploited with impact-energy absorptive structural body parts, such as fender and longitudinal beams that constitute 'crumple zones'. Such parts are typically cold formed from ferritic or multiphase steels, such as Transformation Induced Plasticity (TRIP) steels.

[0006] In order to meet the requirements of the automotive industry for lighter yet safer vehicles, there is an ongoing demand for steel articles having improved mechanical properties, such as improved tensile strength, ductility and impact toughness for impact-energy absorptive crashworthiness, while exhibiting excellent formability to enable down-gauging and moreover, while exhibiting excellent dimensional accuracy (minimal or no springback).

[0007] The present invention aims to find the right steel composition and an improved method for hot-forming steel blanks into a complex shaped article having ultra high tensile strength combined with high ductility and in turn, excellent impact-energy absorptive crashworthiness and which does not have the drawbacks mentioned above for conventional hot formed boron steels and cold formed multiphase steels.

[0008] It is therefore an objective of the present invention to provide a method for hot-forming a steel blank into an article, as well as steel compositions suitable for use in said method, wherein the article has excellent mechanical properties for anti-intrusive crashworthiness, enhanced ductility and in turn improved toughness, and thus improved impact-energy absorptive crashworthiness compared to conventional hot formed boron steels and cold formed multiphase steels.

[0009] It is a further object of the invention to provide a method for hot-forming a steel blank into an article for impact-energy absorptive structural body parts that exhibits minimum or no springback and thus improved dimensional accuracy compared to conventional cold formed multiphase steels.

[0010] It is a further object of the invention to provide a method for hot-forming a steel blank into an article having good coatability.

[0011] It is a further object of the invention to provide a method for hot-forming a steel blank into an article enabling down-gauging and lightweighting simultaneously while improving impact-energy absorptive crashworthiness.

[0012] A method for hot-forming a steel blank into an article according to the present invention comprises the steps of:

(c) forming a heated steel blank in a forming tool into an article and simultaneously quenching it from a temperature T2 to a temperature T3 in a time period t3, wherein the forming tool applies pressure on the steel blank;

(d) cooling the article in the forming tool from temperature T3 to a temperature T4 in a time period t4, wherein the forming tool does not apply any pressure on the steel blank,

(e) quenching the article in the forming tool from temperature T4 to a temperature T5 in a time period t5, wherein the forming tool applies pressure on the steel blank.

[0013] The inventors have found that through forming the heated blank into an article in the sequence as described above, complex shaped articles with enhanced properties can be obtained. In particular the articles exhibit excellent impact-energy absorptive crashworthiness and in turn, down-gauging and lightweighting opportunities based on impact-energy absorptive crashworthiness compared to the use of conventional hot-formed boron steels and cold-formed multiphase steels.

[0014] In the common hot stamping process, the steel blank is simply formed into the desired geometry at and then die-quenched to near-ambient temperature in which the 'complete' transformation to martensite takes place. Thus, the final part exhibits a fully or almost fully martensitic microstructure.

[0015] The inventors have surprisingly found that interrupting the die-quench at a temperature range between Ar1 and Ms by removing the tool pressure hot formed steel articles having a distinctive microstructure and thus improved mechanical properties can be obtained. Said articles show mechanical properties, such as enhanced tensile strength, ductility and impact toughness for impact-energy absorptive crashworthiness, while exhibiting excellent formability to enable down-gauging and moreover exhibiting excellent dimensional accuracy (minimal springback or no springback). In the context of this invention removing the tool pressure means that the punch of the hot-forming tool is removed/open, but the hot-formed article still remains in the die. Once tool pressure has been removed, the part remains in the forming tool, but without tool pressure applied and so the cooling rate from the temperature range between Ar1 and Ms down to 200-400 °C is relatively slow. During the slow air cooling down to 200-400 °C, bainitic ferrite and retained austenite are produced in the microstructure. The tool pressure is resumed to continue die-quenching to 100-200 °C and then the part is removed from the tool. In the context of this invention resuming the tool pressure means that after the opening of the punch and slow air cooling, the punch of the hot-forming tool is closed and pressure is applied again to the hot-formed article in the hot-forming tool. During the process, rather than complete martensite formation, we achieve a significant volume fraction of retained austenite to provide the TRIP steel like microstructure in the final part.

[0016] In step (c) the heated blank is formed in the forming tool into the desired part geometry having a partially or a fully homogenous austenitic microstructure and simultaneously quenched from a temperature T2 to a temperature T3 in a time period t3 under the application of tool pressure.

[0017] Temperature T3 is in the chemistry specific Ar1 and Ms temperature range of the steel composition. The inventors have found that when T3 is in the range of 400-600 °C good results have been achieved. When T3 is below 400 °C, atomic diffusion rates are lower and thus the carbon partitioning rate from bainitic ferrite to austenite is reduced. When T3 is above 400 °C, atomic diffusion rates are higher and permit an appreciable degree of carbon partitioning from bainitic ferrite to austenite in order to stabilise the austenite at ambient temperature (and thus retain austenite in the final microstructure). When T3 is above 600 °C, pearlite formation may occur rather than bainitic ferrite and retained austenite. Better results have been achieved when T3 is in the range of 400-580 °C. Advantageously T3 is in the range of 450 - 550 °C.

[0018] Time t3 is dictated by the quenching rate achieved between temperature T2 and temperature T3. The inventors have found that when t3 is equal or less than 10 s and/or a quenching rate between T2 and T3 of 50 °C/s or more is applied good results have been achieved. Preferably the quenching rate is equal or more than 100, more preferably equal or more than 150, most preferably equal or more than 200 °C/s. Below 50 °C/s austenite may transform to pearlite and proeutectoid ferrite on cooling, preventing martensite, bainite and retained austenite formation in the final microstructure. There is no specific upper limit on the cooling rate. Preferably t3 is equal or less than 8, more preferably equal or less than 5, most preferably equal or less than 4s. Advantageously t3 is equal or less than 2 s.

[0019] Quenching the article in the forming tool to temperature T3 takes place by the action of conductive heat transfer from the heated blank constrained in the forming tool under tool pressure, to the forming tool itself. The forming tool may or may not exhibit integrated cooling channels to enhance the conductive heat transfer rate.

[0020] In step (d) slow cooling of the article from temperature T3 to temperature T4 takes place with the formed part residing in the forming tool in a time period t4, but without applying forming tool pressure i.e. the punch is open/removed. While and during the punch is removed/open the residual heat collected by the forming tool during the forming and quenching stage from temperature T2 to T3 and moreover, latent heat of phase transformation, is able to circulate the formed part by convective heat transfer allowing slow cooling of the hot-formed article. Slow cooling from T3 to T4 takes place purely through air-cooling, with the cooling rate minimised by the residual heat in the forming tool and the latent heat of phase transformation circulating the formed part by convective heat transfer. Thus, the currently use of forced or additional heating following temperature T3 via implementation of an additional heating device is not necessary. During the natural air-cooling step, carbide-free bainitic ferrite is formed and austenite is retained. In the context of this invention natural air-cooling means that the hot-formed part is cooled only though the air of the hot-forming working environment.

[0021] Temperature T4 may be above the chemistry specific Ms temperature, between the chemistry specific Ms and Mf temperatures or below the chemistry specific Mf temperature of the steel. The three preferred temperature ranges mentioned above dictate the volume fraction of martensite that forms in the final quench.

[0022] The inventors have found that when T4 is in the range of 200-400 °C good results have also been achieved. Slow cooling to below 200 °C may result in austenite decomposing into bainite. The rapid quench from temperatures higher than 200 °C ensures that austenite is retained in the final microstructure. There is no specific upper limit on T4. In practice, it will be challenging to maintain T4 above 400 °C as natural air cooling will drop the temperature below 400 °C before we have formed an appreciable volume fraction of bainitic ferrite. Preferably T4 is 300 - 380 °C, more preferably 320 - 370 °C.

[0023] Time t4 is dictated by the residual heat accumulation and management in the forming tool. During time t4, the anisothermal bainitic transformation takes place with slow cooling from temperature T3 to temperature T4, marked by austenite transforming to bainitic ferrite, carbon partitioning from the bainitic ferrite to remaining austenite, solute carbon concentration increase of the remaining austenite and stabilisation of the remaining austenite at ambient temperature. The inventors have further found that when t4 is equal or more than 60 s, good results have been achieved. When t4 is less than 60 s it is likely that insufficient retained austenite will be formed. If t4 is longer than 180 s it would be also not beneficial, because forced heating to retard cooling must be applied. Preferably t4 is in the range of 60-180 s, preferably is in the range of 100-180 s, more preferably is in the range of 120-180 s. The inventors have further found that when the cooling rate between T3 and T4 is in the range of 2-10 °C/s good results have been achieved. Preferably the cooling rate is in the range of 2-8 °C/s. Advantageously the cooling rate is in the range of 2-5 °C/s.

[0024] In the step (e) the formed blank is quenched from temperature T4 to temperature T5, wherein austenite may be completely retained during the final quench to temperature T5, or a significant volume fraction may transform to martensite. There is no specific lower or upper limit on the retained austenite volume fraction. However, it is preferably to retain not less than 5%, more preferably not less than 10 vol. % austenite in the final part following the final quench. If no retained austenite is present there will be no stress/strain inducted transformation during a crash event. The inventors have found that when 5 % or more retained austenite are present a significant stress/strain induced transformation during a crash event can be achieved. Preferably the retained austenite in the hot-formed part is of at most 20 vol.%. Quenching takes place by the action of conductive heat transfer from said blank constrained in the forming tool under tool pressure, to the forming tool itself. The forming tool may or may not exhibit integrated cooling channels to enhance the conductive heat transfer rate.

[0025] The inventors have found that when T5 is in the range of 20 to 200 °C good results have been obtained. When T5 is above 200 °C, austenite may decompose into bainite rather than retained at ambient temperature. Moreover, any martensite formed may temper above 200 °C, degrading tensile strength. Advantageously T5 is in the range of 20-100 °C.

[0026] Time t5 is dictated by the quenching rate achieved between T4 and T5. The inventors have found that when t5 is equal or less than 8 s good results have been obtained. Better results have been achieved when t5 is equal or less than 4s. Advantageously t5 is equal or less than 2 s. The inventors have found that when the cooling rate is at least 50 °C/s good results have been achieved. Better results have been achieved when the cooling rate is at least 100 °C/s. Advantageously the cooling rate is at least 150 °C/s.

[0027] The sequence of the steps c) to e) as laid out above are essential for the invention disclosed.

[0028] In a preferred embodiment for obtaining a partially or homogenous austenitic microstructure the method according to the invention comprises - prior to the hot forming step (c), the steps of:

(a) heating a steel blank to a temperature T1 and holding the heated blank at T1 for a time period t1;
(b) optionally transferring the heated blank from the heating apparatus to a forming tool in a time t2 during which the temperature of the heated blank decreases from temperature T1 to a temperature T2;

[0029] In the method according to the present invention a steel strip or sheet is provided as an intermediate for the subsequent steps. The steel strip or sheet can be obtained by standard casting processes.

[0030] The steel strip or sheet is cut to a steel blank and then heated to a temperature T1 for a time period t1. A preformed steel blank may also be used. The preformed blank may be partially or entirely formed into the desired geometry, preferably at ambient temperature.

[0031] The heating apparatus may be an electric or gas powered furnace, electrical resistance heating device, infra-red induction heating device or any other heating device.

[0032] Temperature T1 is above the chemistry specific Ac1 temperature of the steel to form a ferritic austenitic microstructure.

[0033] In a more preferred embodiment T1 is above the chemistry specific Ac3 temperature to produce a fully or almost fully homogenous austenitic microstructure with uniform distribution of carbon. When the microstructure is a homogenous austenitic microstructure the formability is enhanced.

[0034] The inventors have found that when T1 is at least 710 °C good results have been achieved. Preferably T1 is in the range of 850 - 1150, more preferably 980 - 1100, most preferably 1000 -1050 °C.

[0035] The inventors have further found that the holding time t1 may be chosen in combination with the temperature T1 and the blank thickness in order to control austenitic grain growth and associated quench hardenability of said steel. This means that a greater blank thickness will require longer time t1 and/or higher temperature T1 in order to achieve the desired (a given) austenitic grain size compared to lower blank thickness. The austenite grain size at the end of time t1 will influence quench hardenability, where larger austenite grain size (corresponding to smaller austenite grain boundary surface area) reduces the number of nucleation sites for proeutectoid ferrite formation on cooling and thereby, increases quench hardenability. The degree of quench hardenability will then influence the microstructural evolution namely, proeutectoid ferrite formation during cooling. Therefore t1 is equal or less than 15 min. In a preferred embodiment the inventors have found that when time t1 is equal or less than 10 min. good results have been achieved. Advantageously t1 is in the range of 3-5 minutes in order to increase time and energy efficiency of the process.

[0036] In a preferred embodiment the heating of the steel blank to temperature T1 in step (a) is performed with the steel blank residing in the forming tool. This technique of 'in press heating' is applied with electrodes that contact the steel blank in the forming tool providing electrical resistance heating in the forming tool. This adaption indicates that immediate forming and quenching follows from temperature T1 and time t1, and cooling from temperature T1 to temperature T2 may not be practiced (i.e. time (t2)=0 s). This adaption permits greater quench hardenability and avoidance of proeutectoid ferrite formation for a leaner steel composition. Thus, a leaner chemical composition may be used without forming proeutectoid ferrite. This will enhance economy, formability, weldability and coatability of the steel blank. Moreover a shorter time period t1 and a lower temperature T1 may be practiced giving rise to a finer austenitic grain size and increased rate of carbon partitioning from bainitic ferrite to austenite without any sacrifice to quench hardenability. A finer microstructure gives rise to improved toughness and moreover, increased process efficiency and economy.

[0037] Optionally the heated blank is transferred from the heating apparatus to a forming tool in a time period t2 (step (b)). During the transfer the blank may cool from temperature T1 to temperature T2 by the act of natural air-cooling and/or any other available cooling method.

[0038] The heated blank may be transferred from the heating apparatus to the forming tool by an automated robotic system or any other transfer method.

[0039] Temperature T2 is preferably above the chemistry specific Ar1 temperature of the steel to exhibit an austenitic-ferritic microstructure. In a more preferred embodiment T2 is above the chemistry specific Ar3 temperature to exhibit a homogenous austenitic microstructure. As mentioned above, a fully or almost fully austenitic microstructure at the commencement of forming and quenching is preferable for optimal formability.

[0040] In a preferred embodiment the inventors have found that a delayed time t2 or a temperature T2 lower than Ar3 permits the formation of proeutectoid ferrite, which is desirable in view of the mechanical properties and product performance of the final article, giving rise to a lower proof strength, higher proof strength to ultimate tensile strength ratio and improved impact-energy absorptive crashworthiness.

[0041] The inventors have further found that when T2 is above the Ar3 temperature good results have also been achieved. In particular the article has excellent formability associated with the fully or almost fully austenitic microstructure so that very complex shapes can be formed in a single stroke.

[0042] Time t2 is dictated by the method of transfer. As already mentioned, in-press resistance heating may eliminate time t2 (t2=0 s). A more conventional robotic transfer system from external heating device to forming tool typically takes 12 s. However shorter transport times are beneficial in view of corrosion, process efficiency and hardenability.

[0043] Time t2 may be also chosen in combination with T1, t1 and T2 in order to control the microstructural evolution of steel at the commencement of forming and quenching. For example, higher temperature T1 and/or longer time t1 will give rise to greater austenitic grain growth and delay proeutectoid ferrite formation during time t2. Thus, a shorter time t2 may be used for a given degree of proeutectoid ferrite formation during time t2. Moreover, if a lower temperature T2 is required (to allow proeutectoid ferrite formation), necessarily a longer time t2 will be required to reach temperature T2. As mentioned above, a fully or almost fully austenitic microstructure at the commencement of forming and quenching is preferable for optimal formability. The inventors have further found that when t2 is equal or less than 10 s good results have been also achieved. Preferably t2 is equal or less than 8s, more preferably equal or less than 6s.

[0044] It is a further object of the present invention to provide a steel composition suitable for any one of the methods described herein.

[0045] According to the present invention the steel blank comprises, in weight % (wt%) following elements:

C: 0.05-0.50, preferably, 0.15-0.45, more preferably 0.15-0.35, most preferably 0.15-0.25.

Mn: 0.05-3.00, preferably 1.00-2.50, more preferably 1.00-1.50.

Si: ≤2.0, preferably ≤1.0, more preferably 0.25-0.75, most preferably 0.25-0.50.

Al: ≤2.0, preferably ≤1.0, more preferably 0.25-0.75,

most preferably 0.25-0.50.
P: ≤0.17, preferably ≤0.1, more preferably 0.02-0.07, most preferably 0.02-0.05
S: ≤ 0.002, preferably ≤ 0.001.
B: ≤0.01, preferably 0.001-0.005, more preferably 0.003-0.005.
N: ≤0.01, preferably 0.001-0.005, preferably 0.001 - 0.003.
Nb: ≤0.1, preferably 0.02-0.04, more preferably 0.02-0.03.
Ti: ≤0.1, preferably 0.02-0.04, more preferably 0.02-0.03.
Cr: ≤1.0, preferably 0.2-0.4, more preferably 0.2-0.3.

and optionally one or more of the elements selected from V, Mo, Co, W, and other microalloy rare earth elements (REM), in a total amount of equal or less than 1.0 wt%, the remainder being Fe and unavoidable impurities.

[0046] According to a preferred embodiment the steel blank consists, in weight % (wt%) of the elements mentioned above.

[0047] The reason for the amounts of the main constituting elements is as follows (in wt%).

[0048] C: 0.05-0.50. The carbon content is at least 0.05 wt% to provide adequate interstitial solid solution strengthening, adequate quench hardenability and adequate stabilisation of austenite at ambient temperature while maintaining sufficiently low carbon equivalent for automotive resistance spot-welding techniques. The range of preferable carbon contents will provide a range of products exhibiting a range of strength-ductility values according to the invention. Preferably the carbon content is in the range of 0.15-0.45. In a more preferred embodiment the carbon is in the range of 0.15-0.35. Advantageously, carbon is in the range of 0.15-0.25.

[0049] Mn: 0.05-3.00 The manganese content is at least 0.05 to provide adequate substitutional solid solution strengthening, adequate quench hardenability and adequate stabilisation of austenite at ambient temperature, while minimising segregation of Mn during casting and while maintaining sufficiently low carbon equivalent for automotive resistance spot-welding techniques. The inventors have found that the range of manganese contents in combination with the range of carbon contents provides a product exhibiting a range of strength-ductility values. In a preferred embodiment the Mn content is equal to or less than 2.50. Advantageously, manganese is in the range of 1.00-1.50.

[0050] The silicon, aluminium and phosphorus contents prevent carbide precipitation in the carbon enriched austenite, thus enabling the Ms temperature of the remaining austenite to be depressed below ambient and retention of the austenite at ambient temperature. The silicon, aluminium and phosphorus contents are chosen so to provide the optimal balance of carbide precipitation retardation, kinetics of the anisothermal bainitic transformation, weldability, coatability, economy and manufacture-ability with conventional steelmaking and sheet / strip processing infrastructure.

[0051] Si ≤2.0: The silicon content is chosen as the dominant carbide precipitation retardant, but is limited to a maximum of 2.0 in order to minimise formation of surface-bound silicates that impede hot-rolling, cold-rolling and coating of strip steels. Preferably silicon is equal or less than 1.0. More preferably silicon is in the range of 0.25-0.75. In a most preferred embodiment silicon is in the range of 0.25-0.50.

[0052] Al ≤2.0: Aluminium is limited to a maximum of 2.0 in order to preserve weldability and minimise 'nozzle blockage' during steelmaking and casting. The aluminium content is preferably at least 0.25 to increase the kinetics of the anisothermal bainitic transformation so that appreciable volume fractions of austenite can be retained at ambient temperature after limited time t4. Preferably Al is limited to a maximum of 1.0, more preferably aluminium is in the range of 0.25-0.75 wt%. Advantageously aluminium is in the range of 0.25 - 0.50.

[0053] P ≤ 0.17: The phosphorus content is limited to a maximum of 0.17 and is present to limit silicon and aluminium contents, while still providing satisfactory carbide precipitation retardation. In a preferred embodiment phosphorus is in the range of 0.02-0.07. Advantageously phosphorus is in the range of 0.02-0.05 in order to provide acceptable weldability by automotive standards.

[0054] Nb ≤0.1: The niobium content is limited to a maximum of 0.1 to form niobium carbide precipitates and in turn provide austenitic grain size refinement which increases the kinetics of the anisothermal bainitic transformation. Niobium carbides may also provide precipitation strengthening. Preferably niobium is in the range of 0.02 - 0.04. Advantageously niobium is in the range of 0.02 - 0.03.

[0055] Ti ≤ 0.1: The titanium content is limited to a maximum of 0.1 in order to bond all nitrogen content and form titanium-nitride precipitates at a titanium:nitrogen ratio of 3.42:1 and in turn to enable the boron content to remain un-bonded in solid solution (in order for hardenability) and moreover, titanium is present to bond all residual sulphur content and form titanium-sulphide precipitates at a ratio of titanium:sulphur 1.5:1. The titanium 'excess' remaining after all nitrogen and sulphur has been bonded will bond with carbon to provide titanium-carbide precipitates and in turn provide precipitation strengthening. In a preferred embodiment titanium is in the range of 0.02 - 0.04. Advantageously titanium is in the range of 0.02-0.03.

[0056] Cr ≤ 1.0: The chromium content is limited to at most 1.0 in order to aide hot workability and corrosion resistance. Preferably chromium is in the range of 0.2-0.4. Advantageously chromium is in the range of 0.2-0.3.

[0057] B ≤0.01: The boron content is limited to a maximum of 0.01 in order to enhance quench hardenability and in turn to avoid proeutectoid ferrite formation during cooling from temperature T1 to temperature T2 and during quenching to temperature T3. Preferably boron is in the range of 0.001-0.005. Advantageously boron is in the range of 0.003-0.005.

[0058] N ≤0.01: The nitrogen content is limited to a maximum of 0.01 to prevent boron nitride precipitation and minimise the unwanted TiN formation, where TiN has no benefit to mechanical properties. N has no significant effect on the mechanical properties. Preferably nitrogen is in the range of 0.001-0.005, more preferably in the range of 0.001-0.003.

[0059] S ≤ 0.002. Sulphur is an impurity and needs to be minimised for minimisation of harmful non-metallic inclusions. Therefore the sulphur content is limited to a maximum of 0.002, preferably to a maximum of 0.001.

[0060] The steel strip composition may optionally comprise one or more alloying elements in small amounts such as V, Mo, Co, W, or rare earth elements.

[0061] V, Mo, Co, W, and the other microalloy rare earth elements can be added in a total amount not exceeding 1.00 wt%. in order to benefit from the known property-improving effects of these elements.

[0062] In a preferred embodiment the steel microstructural composition is a multiphase steel, preferably a TRIP or complex phase (CP) like steel that exhibits the capacity for stress / strain induced transformation effect. More particularly the steel is an advanced high strength TBF (TRIP-aided bainitic ferrite) steel product suitable for hot press forming as a boron steel alternative for automotive applications.

[0063] The inventors have surprisingly found that applying the present hot-forming method to a TRIP steel like composition, results in hot-formed articles combining the capacity for the stress/strain induced transformation effect with the principles of TRIP steel microstructural evolution (i.e. carbide-free bainitic ferrite formation and austenite retention). The obtained hot-formed TRIP steel articles, exhibit significant retained austenite volume fractions in the range of 5 to 20 vol. % and the capacity for stress/strain induced transformation. Compared to the traditional cold formed TRIP steels, the optimal austenite volume fraction and capacity for stress/strain induced transformation is established after the forming process and therefore is present in the article, without the austenite volume fraction and capacity for stress/strain induced transformation being used up in the forming process. This enables down-gauging and lightweighting simultaneously while improving impact-energy absorptive crashworthiness.

[0064] In addition to the above, and compared to the traditional hot stamped boron steel, the articles of the present invention show vastly improved mechanical properties, which are comparable to those of the traditional TRIP steel prior to cold forming, and impact-energy absorptive crashworthiness can be achieved.

[0065] The inventors have further found that the hot-formed article obtained by the present invention has ultimate tensile strength (R_m) of at least 750 MPa, and/or a total elongation of at least 16 % and/or a R_mxA product of at least 20000 MPa% and exhibits improved impact-energy absorptive crashworthiness in the final part and in turn, down-gauging and lightweighting potential based on crashworthiness. Preferably R_m is in the range of 750-1500, more preferably in the range of 900-1300, most preferably in the range of 900-1150 MPa. Preferably the total elongation is in the range of 16-35, more preferably in the range of 25-35, and most preferably in the range of 29-35%. Preferably the R_mxA product is in the range of 20000-30000 MPa%, more preferably in the range of 24000-30000 MPa%, most preferably in the range of 26000-28000 MPa%.

[0066] According to a preferred embodiment the steel strip, sheet, blank, preformed blank, or article is provided with a coating. The addition of a coating to the steel strip, sheet, blank or the preformed blank may be performed prior to the hot-forming process or after the hot-forming process and has the purpose of minimising oxidation of the steel when exposed to an oxidising atmosphere at high temperature and/or providing cathodic corrosion protection of the final hot formed part.

[0067] In a preferred embodiment the steel strip, sheet, blank, preformed blank, or article is coated with a zinc based coating, aluminium-silicon based coating or an organic based coating.

[0068] The zinc based coating is a galvanized or galvannealed coating. Although the coating can be applied in various ways, hot dip galvanising is preferred using a standard GI coating bath. Other Zn coatings may also be applied. An example comprises a Zn alloy coating according to WO 2008102009, in particular a zinc alloy coating layer consisting of 0.3 - 4.0 wt% Mg and 0.05% - 6.0 wt% Al and optionally at most 0.2 wt% of one or more additional elements along with unavoidable impurities and the remainder being zinc. An additional element typically added in a small amount of less than 0.2 wt%, could be selected from the group comprising Pb or Sb, Ti, Ca, Mn, Sn, La, Ce, Cr, Ni, Zr or Bi. Pb, Sn, Bi and Sb are usually added to form spangles. Preferably, the total amount of additional elements in the zinc alloy is at most 0.2%. These small amounts of an additional element do not alter the properties of the coating nor the bath to any significant extent for the usual applications. Preferably, when one or more additional elements are present in the zinc alloy coating, each is present in an amount ≤0.02 wt%, preferably each is present in an amount ≤0.01 wt%. Additional elements are usually only added to prevent dross forming in the bath with molten zinc alloy for the hot dip galvanising, or to form spangles in the coating layer.

[0069] The articles according to the present invention exhibit good adhesion to a coating layer, have good surface appearance and superior corrosion resistance after coating.

[0070] It is a further object of the present invention to provide with an article obtained by any one of the methods described herein.

[0071] According to a preferred embodiment the steel sheet or strip is a hot-rolled or a cold-rolled steel strip or sheet.

[0072] It is a further object of the present invention is to provide a multiphase microstructured steel resulting in simultaneously improved strength and ductility. The multiphase microstructure according to the invention is obtained by proving a steel composition as described above and by applying careful control of the thermomechanical treatment method as described above.

[0073] The inventors have found that the final microstructure of the steel comprises bainitic ferrite, (retained) austenite, optionally martensite and/or proeutectoid ferrite. In a preferred embodiment the microstructure may further comprise proeutectoid ferrite. According to the present invention the optimal austenite volume fraction and capacity for stress/strain induced transformation is established after the forming process and therefore is present in the article. The austenite volume fraction is optimal for stress/strain induced transformation to martensite during a crash event, giving rise to optimal impact-energy absorptive crashworthiness. The desired microstructural evolution will not be achieved if the steps are arranged in any other sequence than that disclosed herein.

[0074] In a preferred embodiment the final microstructure comprises (sum added up to 100) by volume fraction (vol.%) of:

- 0 - 75 proeutectoid ferrite, preferably at most 30, most preferably 0.
- 10 - 30 bainitic ferrite, preferably 20-30.
- 5 - 20 retained austenite, preferably 10-20.
- 0 - 75 martensite, preferably 25-75, most preferably 25-50.

[0075] In a more preferred embodiment the final microstructure consist of the above-mentioned volume fractions (vol.%).

[0076] Proeutectoid ferrite: 0-75. Proeutectoid ferrite may be present. However, in those cases where an elevated yield strength (yield strength to ultimate tensile strength ratio) is aimed for, the fraction of proeutectoid ferrite must be limited. Surprisingly, with the correct balance of other microstructural constituents (most notably retained austenite) high work hardening and tensile ductility can still be attained. Above this limit, the final microstructure will not contain enough bainitic ferrite and/or martensite, and thus ultimate tensile strength will be too low. Large fractions of proeutectoid ferrite will also lead to low initial yield strength (low yield strength to ultimate tensile strength ratio). Preferably the proeutectoid ferrite is present in an amount of at most 30 vol.%. Advantageous the proeutectoid ferrite is 0 vol.%. This provides optimal volume fractions of martensite (for ultimate tensile strength) and retained austenite (for stress/strain induced transformation effect) and thus, the maximum R_mxA product. Moreover, the absence of proeutectoid ferrite and in turn, a microstructure characterised by microconstituents exhibiting more similar mechanical properties is envisaged to be beneficial for edge ductility / hole-expansion capacity (HEC).

[0077] Bainitic ferrite: 10 - 30. Bainitic ferrite not only provides strength, but the formation thereof is also a prerequisite for retaining austenite. Beyond the upper limit, insufficient martensite will be present, and thus ultimate tensile strength will be too low and moreover, it is unlikely that more than 30 vol.% bainitic ferrite can be produced within the prescribed heat treatment. The formation of bainitic ferrite in the presence of silicon, aluminium and/or phosphorus drives the partition of carbon to the austenite phase, enabling levels of carbon enrichment in the austenite phase allowing formation of a (meta)stable phase at ambient temperature. Bainitic ferrite has also the advantage over martensite as a strengthening phase that it causes less micro-scale localisation of strain to softer phases and consequently improves resistance to fracture with respect to dual phase steels. Preferably the bainitic ferrite is in the range of 20-30 in order to achieve a more sufficient austenite stabilization.

[0078] Martensite: 0-75. Martensite may be formed during the final rapid quench of the hot forming process. An optimal balance of ultimate tensile strength and ductility is obtained, when the martensite volume fraction is equal or less than 75 vol.%. The complete absence of martensite is advantageous for maximum ductility, where the microstructure is composed of entirely proeutectoid ferrite, bainitic ferrite and/or retained austenite. The complete absence of martensite can be achieved by sufficient carbon enrichment of austenite during the bainitic ferrite/retained austenite formation, so that the martensite start temperature of the remaining/retained austenite is lowered below ambient temperature and thus, during the final rapid quench of the hot forming process, no austenite transforms to martensite. Preferably the martensite is 25-75. Advantageously martensite is in the range 25-50.

[0079] Retained austenite: 5-20. The metastable retained austenite fraction ensures the balanced combination of strength and ductility properties. Retained austenite enhances ductility partly through the stress / strain induced transformation effect, which manifests itself in an observed increase in uniform elongation and total elongation as the work hardening exponent increases to higher plastic strains. Below 5 vol.% the desired level of ductility and/or uniform elongation according to the present invention will not be achieved. The upper limit is set by the composition. Preferably the retained austenite is in the range of 10-20.

[0080] Part or the totality of the process according to the present invention may be conducted in a controlled inert atmosphere of hydrogen, nitrogen, argon or any other inert gas in order to prevent oxidation and/or decarburisation of said steel.

[0081] Forming the blank into the desired part geometry according to the present invention preferably takes place through a stamping/pressing operation, wherein the forming tool is represented by a mating punch and die with said steel positioned between the mating punch and die before punch and die close/mate so to stamp/press said steel into the desired part geometry. However the current process is also applicable to other forming methods such as roll forming.

[0082] In addition to the advantages described above an article produced by the present hot-forming method are characterized by ultra high tensile strength, complex shapes, exhibit minimum or no springback, and achieve a high increment in yield strength, especially for painting. Based on these advantages, excellent impact properties of the steel article are attained.

[0083] In terms of using the final products from the invention, namely the hot stamped parts exhibiting retained austenite and capacity for stress / strain induced transformation owing to the air-cooled anisothermal bainitic transformation, could be used as parts of the automotive body structure, termed Body in White (BiW), for impact-energy absorptive characteristics. Besides automotive BIW applications, the product from the invention could also be used in other applications requiring impact-energy absorptive characteristics.

Figure 1 illustrates the time-temperature cycle of a TRIP800 CR steel strip with a gauge of 1.2 mm applying the hot-forming method according to the present invention.

Figure 2 is the enlarged part of the time-temperature cycle of figure 1, illustrating the process cycle between the opening the furnace at T1 and the opening of the forming tool at temperature T5.

Figure 3 is the enlarged part of the time-temperature cycle of figure 1, illustrating the process cycle between temperature T4 and the opening of the forming tool at temperature T5.

Figures 4a-c show the scanning electron micrographs of final hot-formed article of the hot-formed TRIP800 CR steel article according to the present invention.

Ac1:Temperature at which, during heating, austenite starts to form.

Ac3: Temperature at which, during heating, transformation of the ferrite into austenite ends.

Ar3: Temperature at which, during cooling, transformation of the austenite into ferrite starts.

Ar1: Temperature at which, during cooling, transformation of the austenite into ferrite + cementite ends.

Ms: Temperature at which, during cooling, transformation of the austenite into martensite starts

Mf: Temperature at which, during cooling, transformation of the austenite into martensite ends.

[0084] Critical phase transformation temperatures noted above may be determined by dilatometry experiments.

[0085] The invention will be elucidated by means of the following, non-limiting example making reference to the accompanying figures.

[0086] A cold rolled TRIP800 steel strip with gauge of 1.2 mm is provided by the conventional and known processes. The steel strip contains in wt%: C:0.186; Mn:1.330; Si:1.670; P:0.008; Al:0.131; N:0.004; Nb:0.001; Ti:0.014; Cr:0.026, the balance being Fe and inevitable impurities. The steel strip is heated at 1006 °C (T1), hold for 194.4 seconds (t1) at T1, and transferred from the furnace to the hot-forming tool within 11.3 seconds (t2). During the transfer the steel strip is cooled to 820 °C (T2) due to exposure to air cooling. The steel strip is placed into the hot-forming tool having temperature T2, and simultaneously hot-stamped and die-quenched to 451 °C (T3) within 0.2 seconds (t3). At temperature T3 the punch of the hot-forming tool is opened, i.e. the forming tool applied no pressure and cooled slowly within 53.2 seconds (t4) to 350 °C (T4). The inventors believe that a temperature rise at temperature T3, as shown in figure 3 may be attributed to the upset of the thermocouple and/or to the latent heat of phase transformation (martensite and/or bainitic ferrite formation). At 350 °C (T4) the forming tools is closed again, and the hot-formed article is quenched within 6.8 seconds (t5) to 100 °C (T5). At temperature T5 the forming tool is opened. Figure 4 shows that closing the tool for a second time at 350 °C a delay to the cooling rate (the inflection in the cooling curve) is observed. The inventors believe that the delay may be due to upset of the thermocouple and/or to the latent heat of phase transformation as martensite forms.

[0087] Table 1 gives the temperature values T1, T2, T3, T4 and T5 in °C and table 2 the time periods t1, t2, t3, t4 and t3 in seconds during the process according to the invention.

Table 1

T1	T2	T3	T4	T5
1006	820	451	350	100

Table 2

t1	t2	t3	t4	t5
194.2	11.3	0.2	53.6	6.8

[0088] Figures 4a-c illustrate scanning electron microscopy (SEM) images obtained with the JEOL JSM 6100 Scanning Electron Microscope. Samples were prepared by standard metallographic preparation procedures including etching in 2 % nital solution. Figure 4a illustrates a representation of the global microstructure. Figures 4b and 4c illustrate the finer details of the microstructure at higher magnification. Both figures 4b and 4c illustrate the proeutectoid ferrite matrix and the martensitic dispersions (as annotated in the figures). Figure 4b also illustrates what is considered to be a retained austenite dispersion, while figure 4c also illustrates what is considered to be a bainitic ferrite dispersion. The SEM images confirm a multiphase TRIP steel-like microstructure.

Claims

1. A method of hot-forming a steel blank into an article comprising the steps of:

(c) forming a heated steel blank in a hot forming tool into an article and simultaneously quenching it from a temperature T2 to a temperature T3 in a time period t3, wherein the hot forming tool applies pressure on the steel blank,
(d) cooling the article in the hot forming tool from temperature T3 to a temperature T4 in a time period t4, wherein the hot forming tool does not apply any pressure on the steel blank,
(e) quenching the article in the hot forming tool from temperature T4 to a temperature T5 in a time period t5, wherein the hot forming tool applies pressure on the steel blank.

2. Method according to claim 1, wherein T4 is in the range of 200-400 °C, preferably 300-380, more preferably 320-370 °C.

3. Method according to claim 1 or 2, wherein the time period t4 is equal or more than 60 s. preferably in the range of 60-180 s, more preferably 100-180 s, most preferably 120-180 s, and/or wherein the cooling rate between temperature T3 and T4 is between in the range of 2-10 °C/s, preferably 2-8 °C/s, more preferably 2 -5 °C/s.

4. Method according to any one of the claims 1-3, wherein T5 is in the range of 20-200, preferably 20-100 °C.

5. Method according to any one of the claims 1-4, wherein the time period t5 is equal or less than 8 s, preferably equal or less than 4 s, more preferably equal or less than 2 s, and/or wherein the quenching rate between temperature T4 and T5 is equal or more than 50 °C/s, preferably equal or more than 100 °C/s, more preferably equal or more than 150 °C/s.

6. Method according to any one of the claims 1-5, further comprising - prior to the hot forming step c), the steps of

(a) heating a steel blank to a temperature T1 and holding the heated blank at T1 for a time period t1;
(b) optionally transferring the heated blank to a forming tool in a time t2 during which the temperature of the heated blank decreases from temperature T1 to temperature T2.

7. Method according to claim 6, wherein T1 is at least 730 °C, preferably is in the range of 850-1150, more preferably 980-1100, most preferably 1000 -1050 °C and/or wherein time period t1 is equal or less than 15 min., preferably 3-5 min.

8. Method according to any one of the claims 1-7, wherein the steel blank is partially or entirely preformed.

9. Method according to any one of the claims 1-8, wherein the blank, preformed blank, or article is provided with a zinc based coating or an aluminium-silicon based coating or an organic based coating or any other coating designed to reduce oxidation and/or decarburisation during the hot forming process.

10. Method according to claim 9, wherein the zinc based coating is a coating containing 0.5 - 3.8 wt% Al, 0.5-3.0 wt% Mg, optionally at most 0.2 wt% of one or more additional elements, unavoidable impurities, the balance being zinc.

11. Method according to any one of the claims 1-10, wherein the steel blank comprises, in weight % (wt%) following elements:

C: 0.05-0.50, preferably 0.15-0.45, more preferably 0.15-0.35, most preferably 0.15-0.25;
Mn: 0.05-3.00, preferably 1.00-2.50, more preferably 1.00-1.50;
Si: ≤2.0, preferably ≤1.0, more preferably 0.25-0.75, most preferably 0.25-0.50;
Al: ≤2.0, preferably ≤1.0, more preferably 0.25-0.75, most preferably 0.25-0.50;
P: ≤0.17, preferably ≤0.1, more preferably 0.02-0.07, most preferably 0.02-0.05;
S: ≤ 0.002, preferably ≤ 0.001;
B: ≤0.01, preferably 0.001 - 0.005, more preferably 0.003-0.005;
N: ≤0.01, preferably 0.001 - 0.005, preferably 0.001 - 0.003;
Nb: ≤0.1, preferably 0.02 - 0.04, more preferably 0.02-0.03;
Ti: ≤0.1, preferably 0.02 - 0.04, more preferably

0.02-0.03;
Cr: ≤1.0, preferably 0.2-0.4, more preferably 0.2-0.3;

and optionally one or more of the elements selected from V, Mo, Co, W, and other microalloy rare earth elements (REM), in a total amount of equal or less than 1.0 wt%, the remainder being Fe and unavoidable impurities.

12. Steel article obtained by the method according to any one of the claims 1-11, wherein the steel article has an ultimate tensile strength R_m of at least 750 MPa, preferably 750-1500, more preferably 900-1300, most preferably 900-1150 MPa.

13. Steel article according to claim 12, wherein the steel article has a total elongation in the range of 16-35%, more preferably 25-35, and most preferably 29-35% and/or wherein the R_mxA product is in the range of 20000-30000 MPa%, more preferably 24000-30000 MPa%, most preferably 26000-28000 MPa%.

14. Steel article according to claims 12 or 13, wherein the microstructure of the steel article comprises at least 5 vol.% retained austenite.

15. Steel article according to any one of the claims 12-14, wherein the microstructure of the steel article comprises (sum added up to 100) by volume fraction (vol.%) of:

- 0-75 proeutectoid ferrite, preferably at most 30, most preferably 0,
- 10 - 30 bainitic ferrite, preferably 20-30,
- 5 - 20 austenite, preferably 10-20,
- 0 - 75 martensite, preferably 25-75, most preferably 25-50.

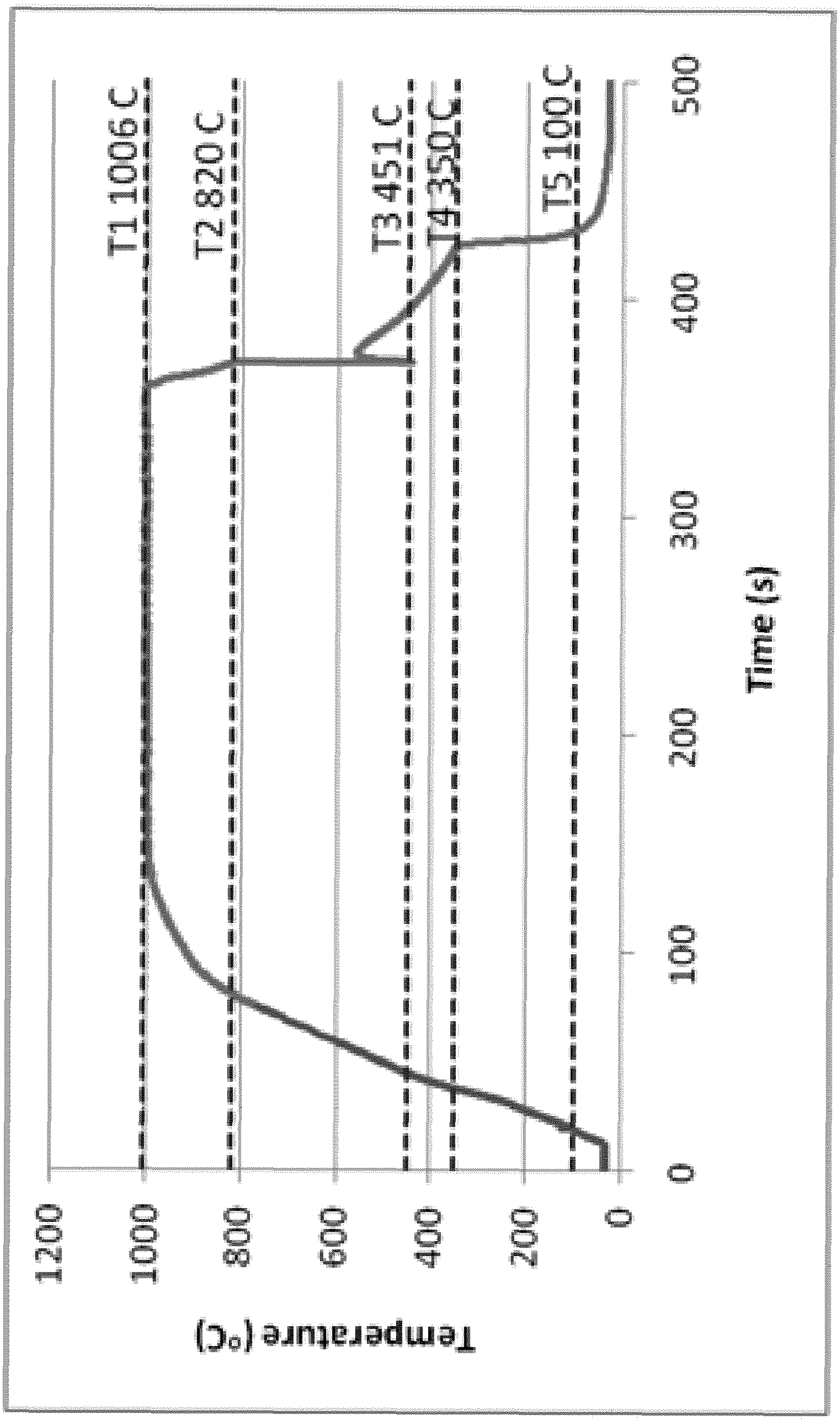

T1 1006 C
T2 820 C
T3 451 C
T4 350 C
T5 100 C
Temperature (°C)
1200
1000
800
600
400
200
0
0
100
200
300
400
500
Time (s)

FIG 1

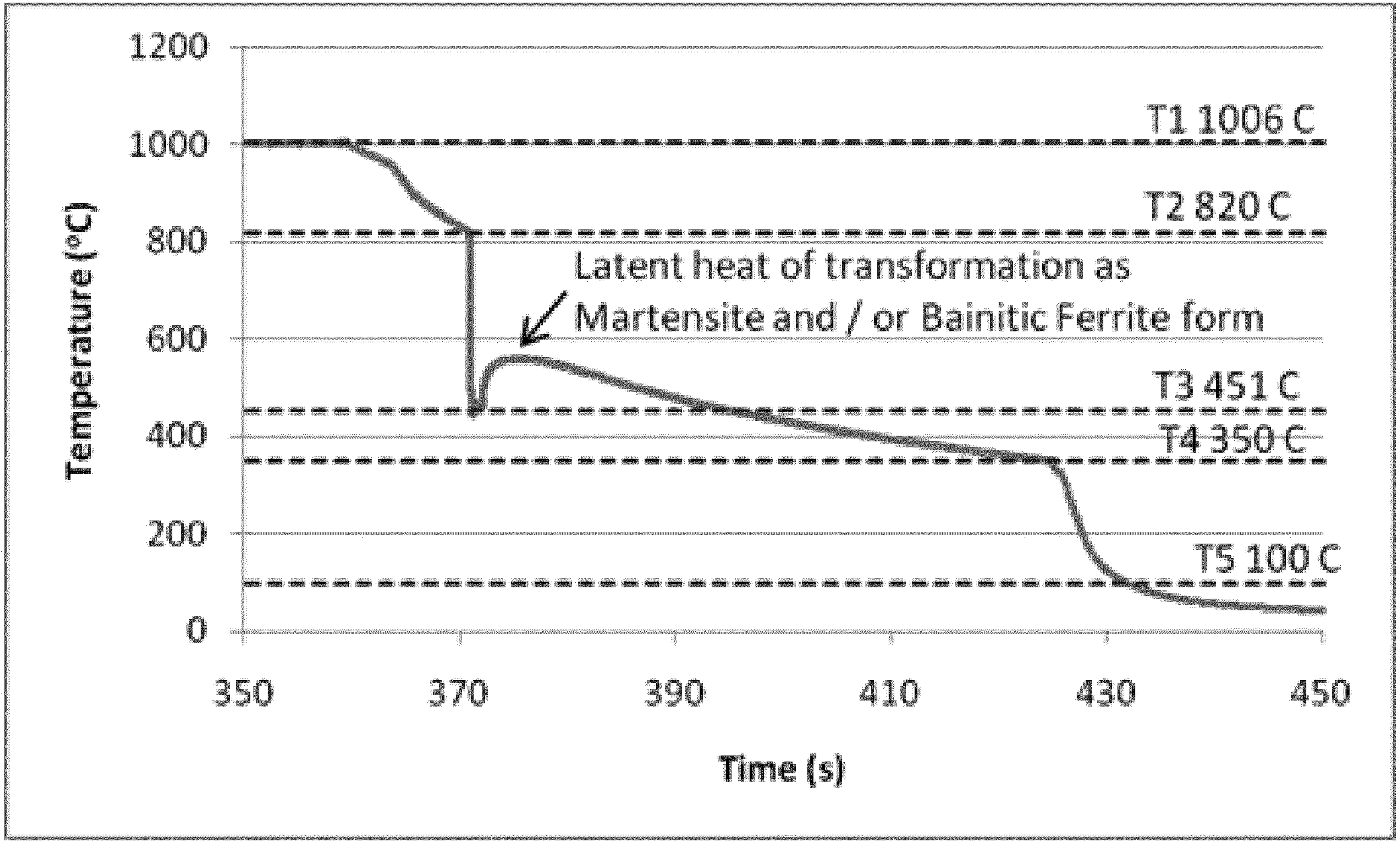

1200
1000
800
600
400
200
0
Temperature (°C)
350
370
390
410
430
450
Time (s)
T1 1006 C
T2 820 C
Latent heat of transformation as
Martensite and / or Bainitic Ferrite form
T3 451 C
T4 350 C
T5 100 C

FIG 2

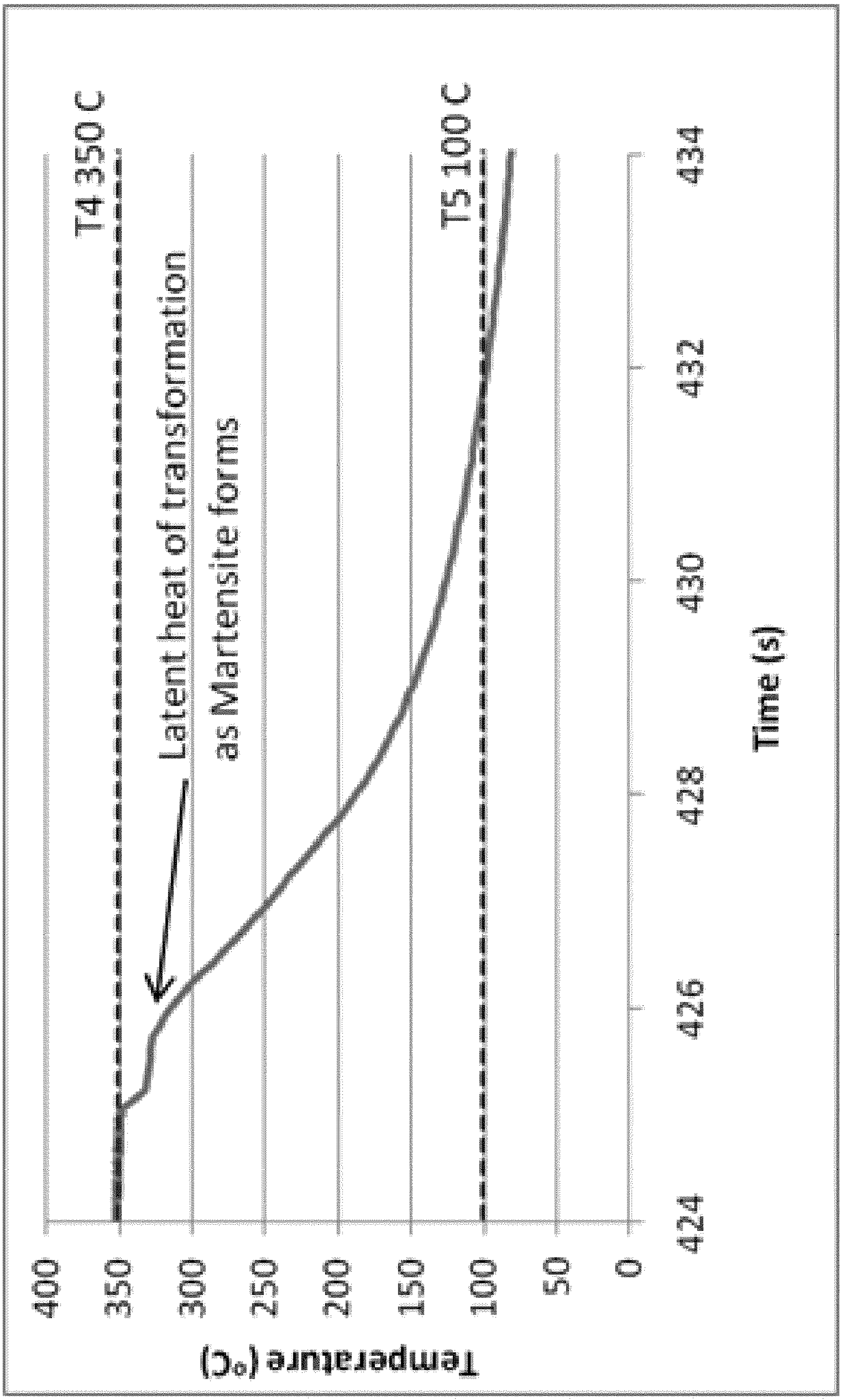

Temperature (°C)
400
350
300
250
200
150
100
50
0
Time (s)
424
426
428
430
432
434
T4 350 C
T5 100 C
Latent heat of transformation as Martensite forms

FIG 3

FIG 4a

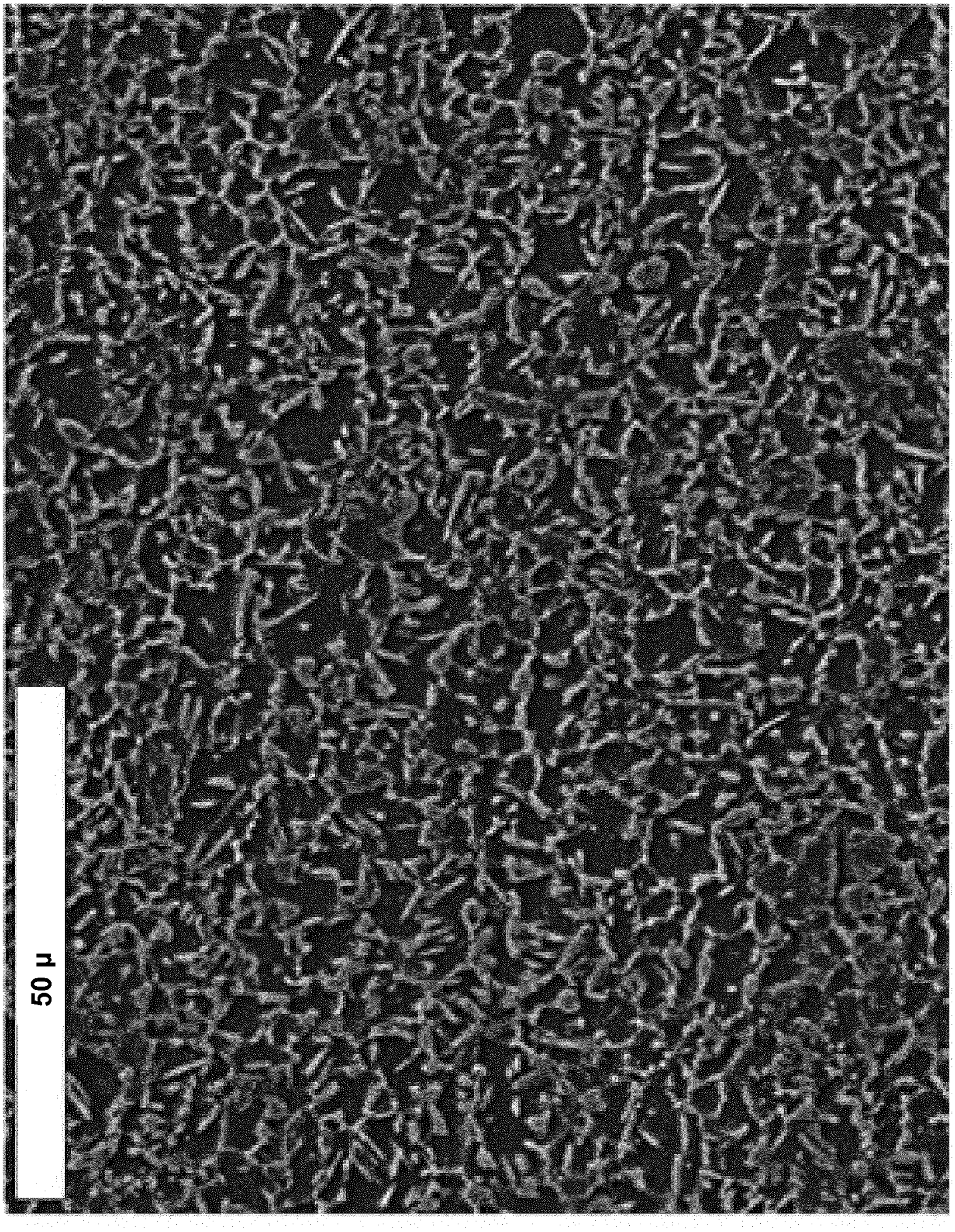


FIG 4b

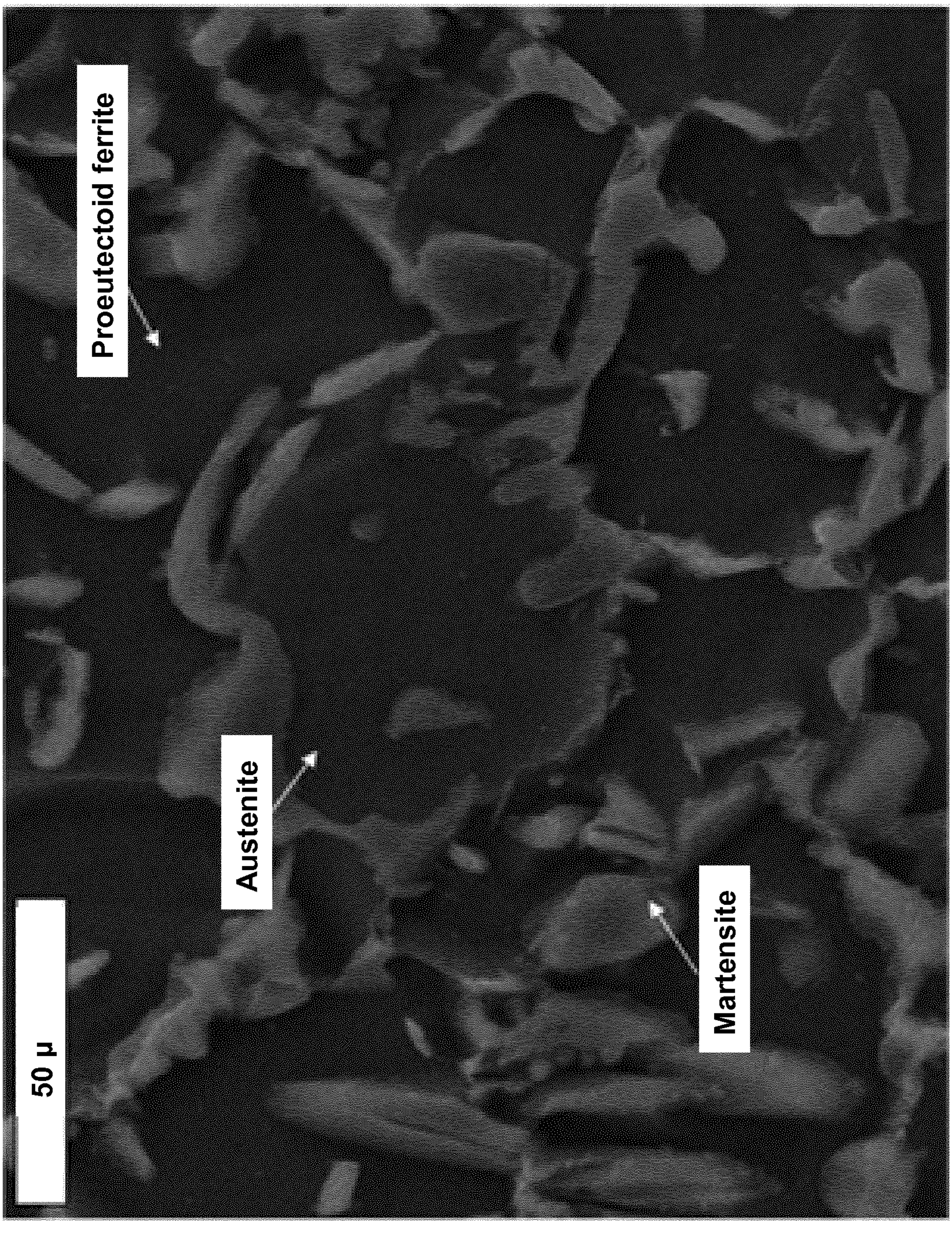
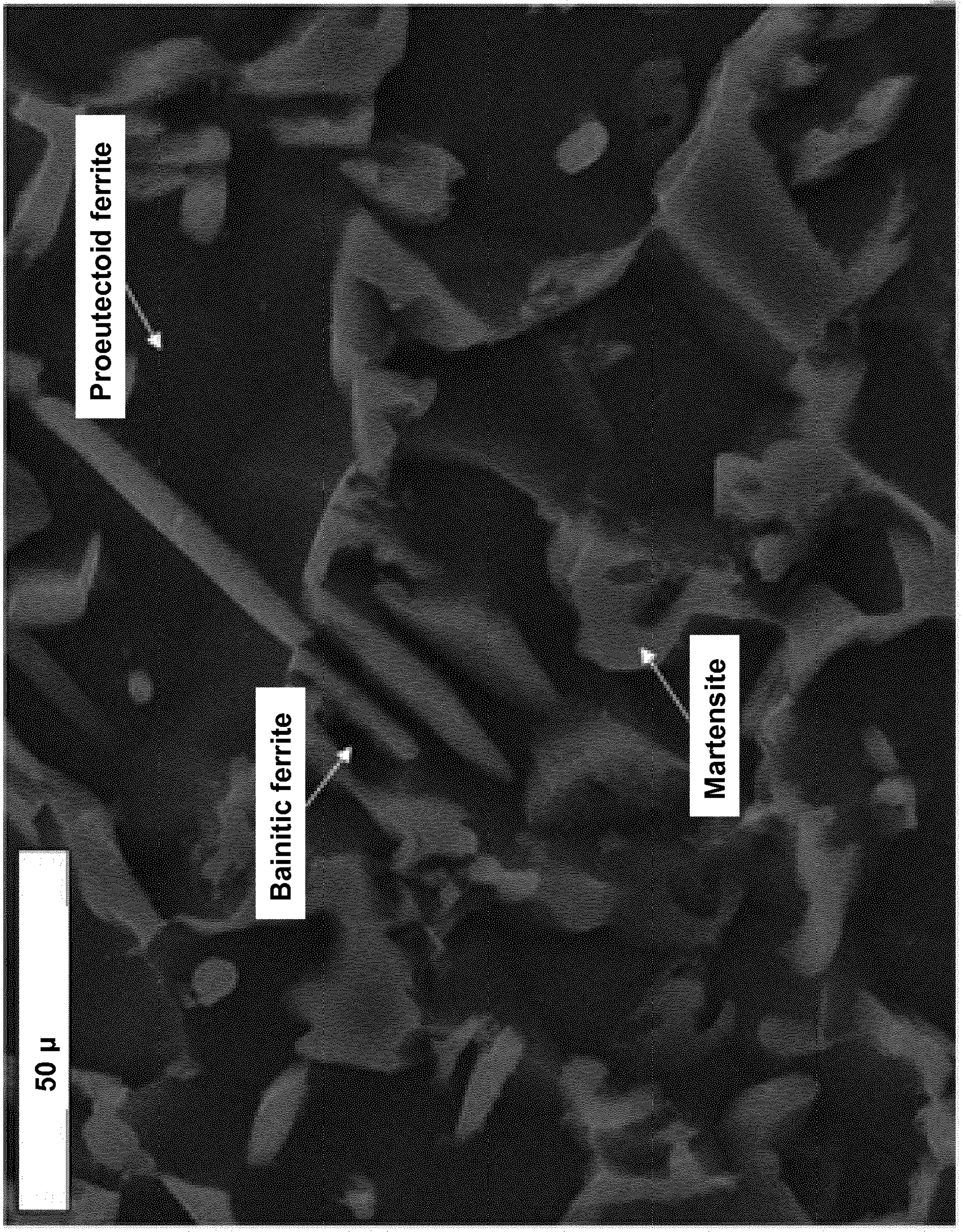


FIG 4c



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EUROPEAN SEARCH REPORT

Application Number
EP 16 20 1185

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Category	Citation of document with indication, where appropriate, of relevant passages	Relevant to claim	CLASSIFICATION OF THE APPLICATION (IPC)
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			C21D B21D

The present search report has been drawn up for all claims

1

Place of search	Date of completion of the search	Examiner
Munich	19 May 2017	Abrasonis, Gintautas

CATEGORY OF CITED DOCUMENTS

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ANNEX TO THE EUROPEAN SEARCH REPORT ON EUROPEAN PATENT APPLICATION NO.

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