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(54) ALUMINIUM ALLOY AND METHOD FOR PRODUCING THE ALLOY

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(71) Applicant: NANO ALLOYS TECHNOLOGY, Alphen aan den Rijn (NL)

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(72) Inventors: Tomasz TOKARSKI, Krakow (PL); Rafal PASTUSZEK, Bielsko-Biala (PL)

(57)

ABSTRACT

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A rapidly-solidified and plastically consolidated aluminium alloy comprises between 3.00 and 10.00 wt. % of magnesium and between 1.00 and 6.00 wt. % of manganese and dispersoid forming transition elements that are selected from the group consisting of chromium, vanadium, titanium, zirconium, molybdenum, cobalt and niobium. The total amount of these transition elements is at least 0.50 wt. %. The maximum amounts of these elements is 1.50 wt. % for Cr, 1.50 wt. % for V, 1.00 wt. % for Ti, 1.00 wt. % for Zr, 1.50 wt. % for Mo, 1.50 wt. % for Co and 1.00 wt. % for Nb. The aluminium alloy is corrosion resistant and allows, due to its fine grained stabilized structure, to combine a high yield strength at room temperature with a good thermal resistance and with relatively low forming forces at high temperatures. The liquidus temperature of the alloy remains relatively low so that it can be produced easily on an industrial scale.

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✖	HV 20.00 kV	HFW 207 µm	curr 16 nA	det CBS	mag 2 000 x	WD 9.9 mm	50 µm
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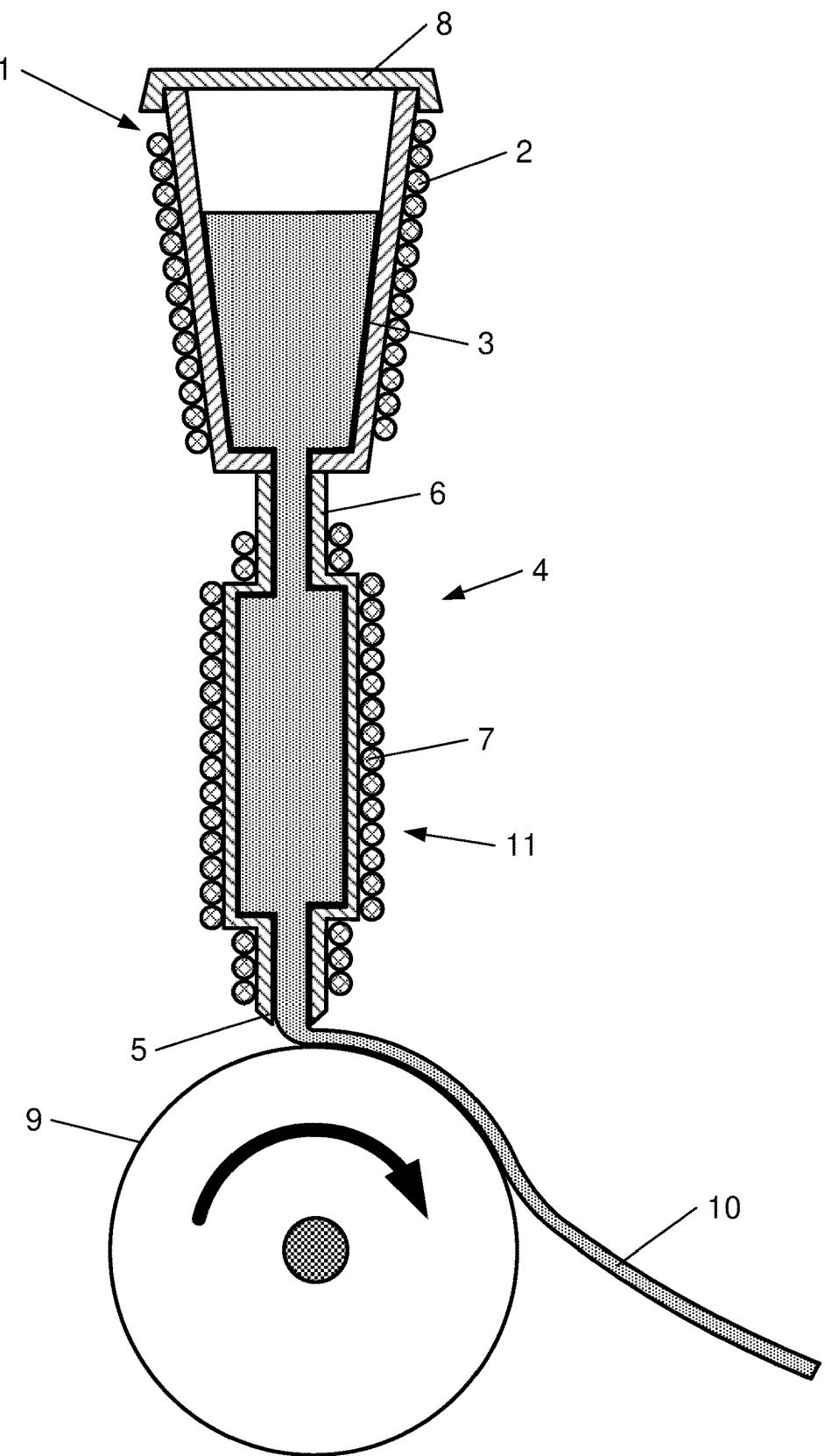


Fig. 1



Fig. 2



Fig. 3

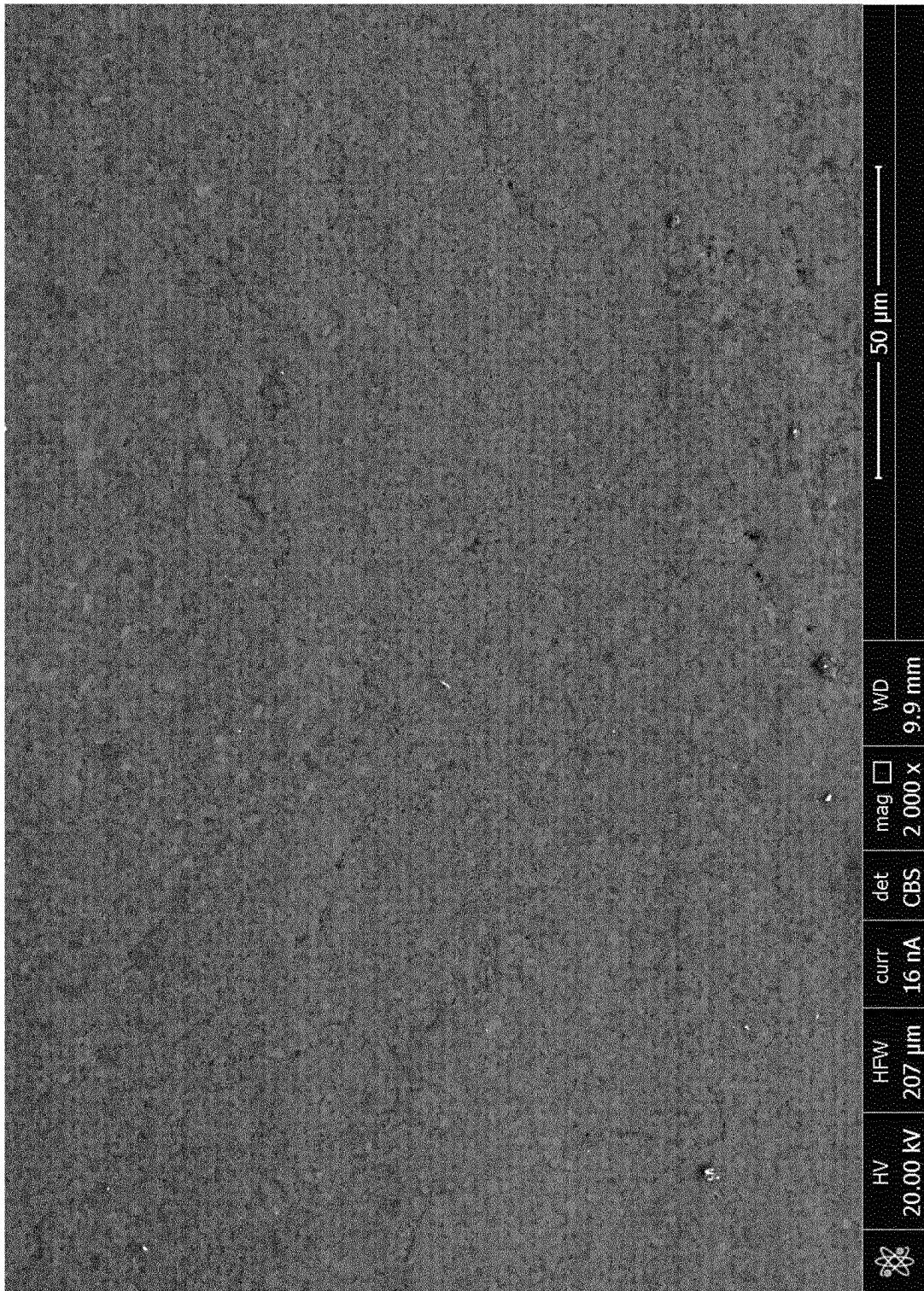


Fig. 4

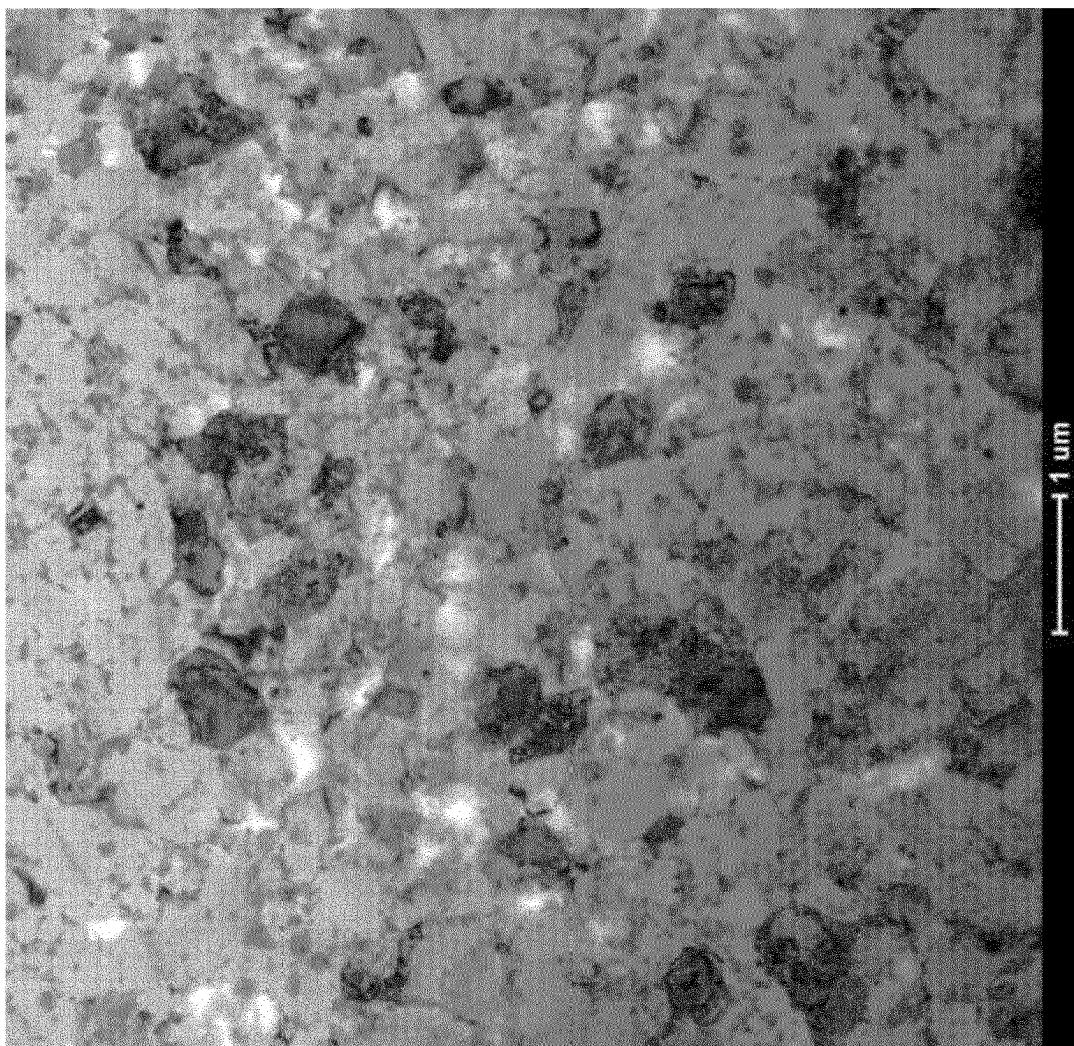


Fig. 5

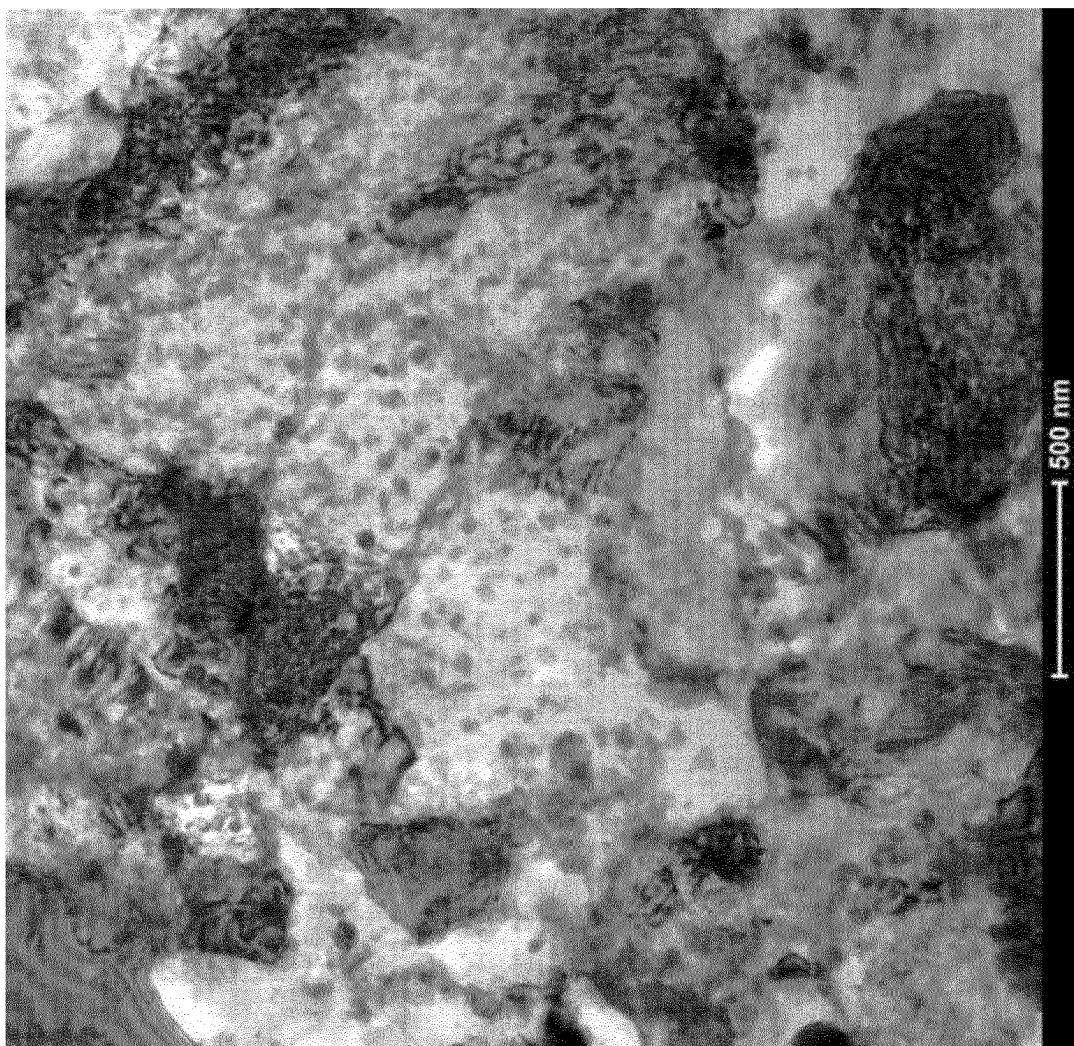


Fig. 6

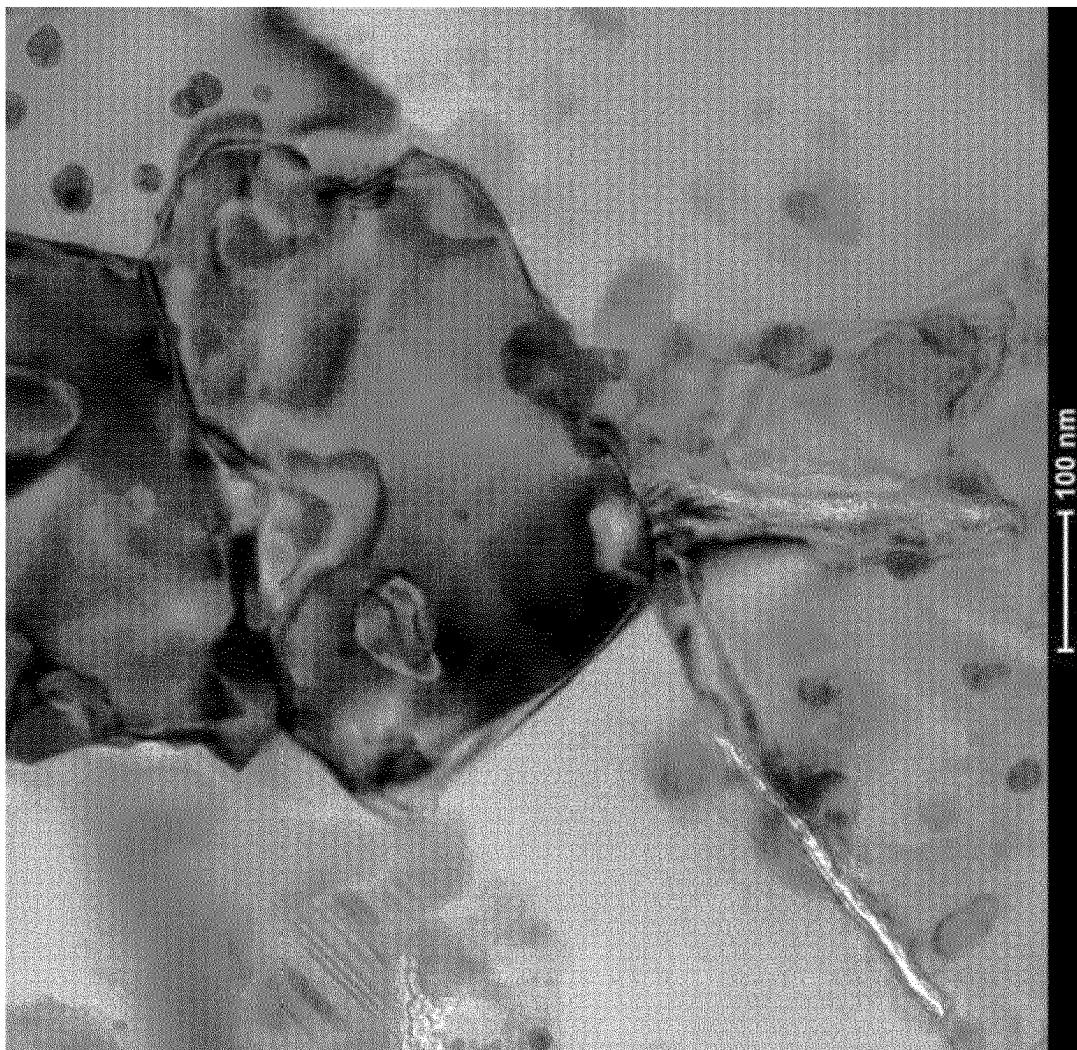


Fig. 7

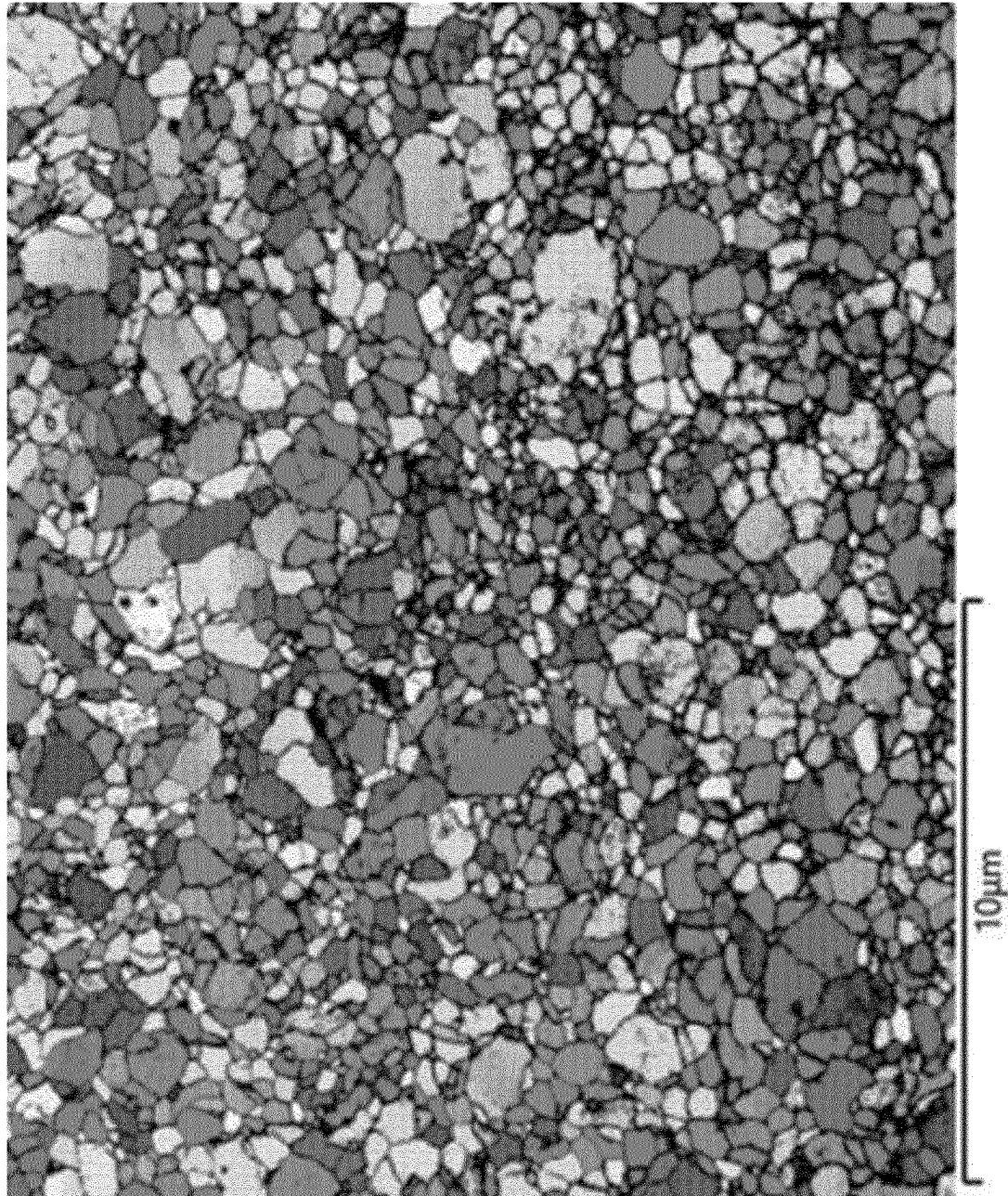


Fig. 8

ALUMINIUM ALLOY AND METHOD FOR PRODUCING THE ALLOY

CROSS-REFERENCE TO RELATED APPLICATION(S)

[0001] This present application is a national stage application of International Patent Application No. PCT/EP2023/059579, filed Apr. 12, 2023, which claims priority to International Patent Application No. PCT/EP2022/059808, filed Apr. 12, 2022, the disclosures of which are hereby incorporated by reference in their entireties.

TECHNICAL FIELD

[0002] The present disclosure relates to an aluminium alloy that is rapidly-solidified and plastically consolidated in particular by plastic deformation under pressure at a temperature higher than 300° C. The aluminium alloy is more particularly a wrought aluminium alloy suited for being plastically deformed, in particular by hot forging such as die forging. The disclosure also relates to a method for producing the rapidly-solidified, plastically consolidated aluminium alloy.

BACKGROUND

[0003] Current industrial wrought aluminium alloys are not produced by a rapid solidification process but are produced by casting the molten alloy and solidifying it by a relatively slow cooling process. The industrial wrought aluminium alloys are divided in several groups that are called series. They can be additionally grouped following the type of the major mechanism responsible for the strength of the alloy. The most popular constructional materials are defined in the 6xxx, the 2xxx and the 7xxx series and are precipitation strengthened. The 7xxx series is known to contain the strongest aluminium alloys whilst the 6xxx series comprises the moderately strong aluminium alloys. For example 6082 alloys have a yield strength (YS) of approximately 350 MPa while 7075 alloys have a yield strength of approximately 500 MPa. A major drawback of the high strength alloys is that the higher their strength, the more demanding the techniques used to process the alloy, requiring for example high forming forces and complicated thermal treatments. Additionally corrosion resistance may be an issue, in particularly in the high strength 2xxx and 7xxx series.

[0004] Another group of wrought alloys comprises the 3xxx and the 5xxx series, wherein the major strengthening mechanism is solid solution strengthening. These alloys are corrosion resistant but their strength is significantly lower than the strength of the alloys of the 6xxx series. For example a 5083 alloy, being one of the strongest of the 5xxx series, has a yield strength of only about 200 MPa.

[0005] Most of these industrial alloys are made in bulk, by a slow cooling process, which involves several disadvantages. During the slow cooling process, alloying elements segregate within dendrites and to the inter-dendritic regions and intermetallic phases are formed that grow during the cooling process. These phenomena pose significant constraints within the field of standard industrial crystallisation processes (utilizing a low cooling rate). Moreover, the compositions of the industrial alloys are confined to certain limits.

[0006] Chemical segregations may produce volumes (phases) in the material that have a lower melting point. It can cause incipient (local) melting during hot forming so that the forming temperature needs to be reduced. Industry is mitigating those effects to some extent by a proper homogenisation treatment at a high temperature.

[0007] Structure inhomogeneities and large and brittle inter-metallic particles are making plastic forming processes very hard, if not impossible. The presence of inter-metallic phases, which are often big and rectangular or needle shaped, can lead to the initiation of cracks inside the material reducing the ductility thereof. This problem arises especially when higher concentrations of alloying elements are applied: chemical segregations are becoming more pronounced, inter-metallic phases are becoming larger to the extent that casting processes are even no longer possible. During slow crystallisation inhomogeneities can cause even such a high concentration of stresses across the solidified billet that the material may crack during cooling.

SUMMARY

[0008] The purpose of producing the aluminium alloy by a rapid solidification process is to prevent the formation and growth of inter-metallic phases during crystallisation as well as to reduce or suppress segregation of the alloying components within dendrites and to the inter-dendritic regions. A rapid solidification process, involving a fast/rapid cooling of the liquid alloy, allows therefore to introduce higher amounts of alloying elements without the negative effects described above.

[0009] The cooling rate is however physically limited by heat transfer phenomena and by the heat capacity of the material. This means that the dimensions of the material that is subjected to the cooling process need to be limited. In fact, achieving a homogenous chemical composition requires very high cooling rates that can only be obtained by producing fragmented pieces of the material. In practice two technologies are mainly used: powder atomisation and melt spinning. Both are characterised by a high surface to volume ratio that makes it possible to extract the heat very fast. In particular, by using the efficient heat transfer between a molten alloy and a cold drum surface, melt spinning is very efficient in terms of heat extraction. As a result of these rapid solidification techniques a crystallised material is produced in the form of a highly fragmented material, having in particular at least one dimension that is smaller than 100 µm.

[0010] The rapidly solidified material has to be consolidated to produce larger pieces of the alloy. Typically this is achieved by a plastic consolidation process, such as for example by a hot extrusion process. In this process, the aluminium alloy is consolidated by being plastically deformed under pressure at an elevated temperature. During the consolidation process, the particles (i.e. the fragmented pieces) of the rapidly solidified alloy are subjected both to compressive stresses and to shear stresses. The individual rapidly solidified elements/particles have indeed to be brought in very close contact enabling the formation of new interatomic bonds. Plastic deformation is required to create fresh surfaces that are free of oxides and that can thus be bonded to one another. Due to the shear stresses produced by the plastic deformation, the surface to volume ratio of the particles is increased so that new, oxide free, surfaces are exposed. At the same time, smaller grains are produced.

[0011] A major advantage of the smaller grains is that the aluminium alloy becomes stronger. The grain boundaries are indeed barriers for dislocation motions. The relationship between the yield stress and the grain size can be described mathematically by the Hall-Petch equation. A further advantage of the small grains is that grain boundary sliding may be activated at higher temperatures so that the alloy can be formed with reduced stresses/forces at these higher temperatures.

[0012] A drawback of smaller grains is, however, that the material has a higher amount of internal energy that is stored in the form of a large surface area of grain boundaries. This means that when heating up such a material, for example to improve its formability, recovery mechanisms become very active leading to grain growth and to the loss of the positive characteristics of the alloy.

[0013] In practice, there exist already a number of rapidly solidified aluminium alloys. Based on their application, they can be divided in two groups. The first group comprises high strength alloys that are designed directly as a replacement of the 7xxx series alloys whilst the second group comprises aluminium alloys intended for application at elevated temperatures.

[0014] An example of alloys that belong to the first, high strength group are alloys designed by Alcoa and presented in the paper "Rapidly solidified aluminium alloys: market assessment", E. Lavernia et al., Journal of Metals, 1985. Alloys designated as the X7090 and X7091 are shown as alternatives for the standard 7075 alloy. Depending on their composition and the production route, the rapidly solidified alloys showed a 10 to 15% strength increase in comparison to the industrial counterpart. The observed improvement is minor as precipitation strengthening remains the primary strengthening mechanism governing mechanical properties, while the effect of the grain refinement remains small.

[0015] Higher strengthening effects are obtained by rapid solidification of solid solution strengthened alloys. A rapidly solidified and plastically consolidated fine grained aluminium alloy, which contains a combination of magnesium, manganese and a small amount of a number of transition elements, is disclosed in the article of T. Tokarski et al. "Microstructure and plasticity of hot deformed 5083 aluminium alloy produced by rapid solidification and hot extrusion" in Archives of Metallurgy and Materials, Volume 57, Issue 4, 2012 (DOI: 10.2478/v10172-012-0140-2). As described in this article, an aluminium alloy of the 5083 series was produced both in the form of a conventional alloy by a casting process and in the form of an ultra-fine grained alloy by a rapid solidification/plastic consolidation process. The exact composition of the alloy is not disclosed in the article but the aluminium alloy of the 5083 series is described to contain 4.00-4.90 wt. % of Mg, 0.40-1.00 wt. % of Mn, max. 0.40 wt. % of Si, max. 0.25 wt. % of Zn, max. 0.40 wt. % of Fe, max. 0.25 wt. % of Cr and max. 0.15 wt. % of Ti.

[0016] The molten alloy composition was rapidly solidified by melt spinning in the form of ribbons having a thickness in the range of 50 to 100 μm . The rapidly solidified material was extruded at a temperature of 400° C. with a cross section reduction area 2 equal to 25. The same composition was also conventionally cast, homogenized for 12 hours at 530° C. and hot extruded in the same way as the rapidly solidified material. Due to the fine grained structure, the rapidly solidified and extruded material had a higher

yield strength than the conventionally cast alloy, namely a yield strength of 320 MPa compared to a yield strength of only 180 MPa for the conventionally cast alloy, which is an improvement of about 70%. A further advantage of the rapidly solidified material was that it had a lower flow stress during hot deformation. The rapidly solidified material was thus in other words easier to form under hot conditions. At a constant strain rate of 10^{-2} s^{-1} and a temperature of 450° C. the rapidly solidified material had a maximum flow stress of 19 MPa whilst the conventionally cast material had a maximum flow stress of 48 MPa. The reduced flow stress would be due to grain boundary sliding, in presence of higher density of grain boundary regions, thus being more fine-grained.

[0017] The strength of the rapidly solidified and extruded alloy was partially due to the solid solution strengthening effect of Mn and especially of Mg in the aluminium matrix. The grain refinement, to an average grain size of 700 nm, was however the major strengthening effect. Fine scale intermetallic phases, having a size of up to 50 nm and containing Mn, Fe and Cr, were also formed. These particles were dispersed both inside the grains and along the grain boundaries. The particles present on the grain boundaries provided for a thermal stabilization of the alloy structure during hot deformation by the so-called grain boundary pinning or Zener pinning. As presented in the paper "The effect of solutes on grain boundary mobility during recrystallization and grain growth in some single-phase aluminium alloys", Y. Huang, F. J. Humphreys, Materials Chemistry and Physics 132 (2012) 166-174, an additional grain growth suppressing factor could be connected to the segregation of Mg to the grain boundaries.

[0018] Notwithstanding the thermal stabilization by the dispersed nanoparticles and the grain boundary element segregation, the rapidly solidified material lost about 10% of its hardness during the hot compression test at 450° C. whilst the conventionally cast material lost only about 1% of its hardness during this test. The initial hardness of the rapidly solidified and extruded alloy dropped more particularly from 111 HV0.5 (Vickers Hardness) to 98.5 HV0.5. The loss of alloy hardness was attributed to coarsening of the grains or of the intermetallic phases. Due to the loss of alloy hardness, it was advised to avoid alloy temperatures above 450° C.

[0019] As described in the article of T. Tokarski et al. "Thermo-mechanical processing of rapidly solidified 5083 aluminium alloy-structure and mechanical properties" in Archives of Metallurgy and Materials, Volume 60, Issue 1, 2015 (DOI: 10.1515/am-2015-0028) the yield strength of the alloy described in the previous article could be increased by a one- and by a two-step thermo-mechanical treatment. The thermo-mechanical treatment consisted of a cold rolling step followed by an isothermal annealing step at 300° C. for 3 hours, and was repeated twice, in a further experiment, resulting in a further refinement of the microstructure of the alloy.

[0020] The increase of the yield strength was due to a refinement of the alloy structure from the initial average grain size of 710 nm to an average grain size of 360 nm and 270 nm. The increase in yield strength was in accordance with the Hall-Petch relationship. The yield strength increased from 320 MPa for the as-extruded material to 373 MPa for the one-step thermo-mechanical treatment and to 419 MPa for the two-step thermomechanical treatment. Even with such a complex thermo-mechanical treatment it

was not possible to achieve a high-strength aluminium alloy having in particular a yield strength of about 500 MPa or higher. Moreover, due to the thermo-mechanical treatment the ductility/elongation of the material dropped from 20% to 12%. The thermal stability of the rapidly solidified material as extruded was also examined at a temperature of 500° C. After being annealed for 3 hours at this temperature, the yield strength of the alloy dropped from 320 MPa to 130 MPa and the average grain size increased enormously, namely from 710 nm to 100 µm. Notwithstanding the formation of the highly dispersed intermetallic compounds the rapidly solidified and extruded alloy could clearly not be stabilized sufficiently to be able to be exposed to temperatures higher than 450° C. Due to the increased internal energy as a result of the increased number of grain boundaries, the stronger, thermo-mechanically treated alloys would even have a lower thermal stability.

[0021] As described in the first article of T. Tokarski et al., it was known that the thermal stability of a material having an ultra-fine grained structure can be enhanced by the addition of elements such as Zr, Sc, Cr and Mn that form highly dispersed intermetallic compounds that provide thermal stability during hot deformation by grain boundary pinning. In the tested alloy, the thermal stability was increased by dispersoids formed by Mn, Fe and Cr rich phases. However, the aluminium alloy still had only a limited thermal stability that did not allow the alloy to be further deformed at elevated temperatures higher than 450° C. Moreover, although the alloy was strengthened by the Hall-Petch effect, the yield strength thereof was still limited, namely to about 320 MPa. Finally, although the flow stress during hot deformation at higher temperatures was reduced due to the grain boundary sliding effect, it was still quite high at the maximum temperature of the alloy, namely 19 MPa at a temperature of 450° C. and a strain rate of 10^{-2} s^{-1} .

[0022] Another way to strengthen an aluminium alloy like AlMg4.5Mn (AA5083) that is rapidly solidified is disclosed in the manuscript "ScalmalloyR=A unique high strength AlMgSc type material concept processed by innovative technologies for aerospace applications", Frank Palm et al., in Proceedings of the World Powder Metallurgy Congress and Exhibition, World PM 2010. An aluminium alloy was made containing 3.6 wt. % Mg, 1.4 wt. % Sc, 0.2 wt. % Zr and 0.3 wt. % Mn. In order to be able to dissolve the alloying elements, the alloy had to be heated to a temperature of 900° C. The alloy was heated under an argon atmosphere to prevent oxidation and the formation of hydrogen. The molten alloy was rapidly solidified by melt spinning to avoid the formation of intermetallic phases during the rapid solidification. The produced flakes were not subjected to a plastic consolidation process to produce a fine grained structure. Instead, the rapidly solidified flakes were subjected to a heat treatment to produce nano-scale $\text{Al}_3(\text{Sc}_{1-x}\text{Zr}_x)$ precipitations in order to increase the strength of the alloy (to a hardness of up to 240 HV) by precipitation hardening. A drawback of such a Scalmalloy® is that high forming forces are still required. Moreover, the high cost of scandium makes the alloy very expensive and thus only suitable for high end applications in particular for aerospace applications.

[0023] Examples of rapidly solidified aluminium alloys that belong to the second group, i.e. to the group of alloys that remain sufficiently strong also at higher temperatures, can be found for example in U.S. Pat. No. 10,232,442. The

rapidly solidified aluminium alloy compositions disclosed in this US patent are suited for applications wherein the alloy has to keep its mechanical properties also at somewhat higher temperatures, such as for example in case of a turbocharger compressor wheel. Notwithstanding the fact that those alloys are dispersion strengthened, they are still not stable at the higher temperatures that are required to reduce the pressures for hot extruding or hot forming the alloy. Indeed, in order to keep the mechanical properties of the alloy, U.S. Pat. No. 10,232,442 teaches to extrude the rapidly solidified alloy under temperature-limited forming conditions of 275° C. or less. The extruded alloy should subsequently also be formed, for example die forged, at such relatively low temperatures. A drawback of such low temperatures is, however, that the extrusion process and the subsequent forming processes require quite high pressures/forces. Both the extrusion and the forming steps thus require heavy and expensive tools and involve also high operation costs.

[0024] The alloy compositions disclosed in U.S. Pat. No. 10,232,442 all contain high amounts of dispersoid forming elements. Although those alloys are suited for applications at relatively elevated temperatures, it appears that they are still not stable at higher forming temperatures. In general, all the aluminium alloys that are designed for being applied under elevated temperature conditions are alloyed with high amounts of transition elements. Of these transition elements, Ni and Fe are used in the highest quantities as dispersoid forming elements because they can be easily alloyed into the molten aluminium without significant overheating of the melt. However, even quite high amounts of these transition elements do not enable to achieve a stable alloy that can also be formed at high temperatures reducing the flow stress, for example during a hot forging step.

[0025] Another example of a rapidly solidified aluminium alloy designed to be applied at relatively elevated temperatures, and wherein not only dispersoids are formed but wherein additionally Mg is employed as a solid solution strengthening element, is described in the article of A. Kula et al. "Structural and mechanical features of rapidly solidified Al-2Fe-2Ni-5Mg alloy". It contains much more alloying elements than the 5083 alloy presented in the paper of Tokarski et al., namely, in addition to the 4.56 wt. % Mg that is mainly present in solid solution, 2.73 wt. % Fe and 2.37 wt. % Ni. Notwithstanding the presence of such a high amount of additional dispersoid forming alloying elements, the alloy still had a yield strength of only about 300 MPa at room temperature. This yield strength was attributed to solid solution hardening achieved by the Mg addition and to dispersion strengthening achieved by the refining of intermetallic compounds. On the scanning transmission electron microscopy (STEM) images, it can be seen that the alloy indeed contained a lot of dispersoids that were however quite large. Moreover, also the grain size of the alloy was quite large so that only a limited Hall-Petch strengthening effect was obtained. Not only the mechanical strength of the alloy was quite limited but apparently also its thermal stability. After deformation tests at increased temperatures, the microhardness of the alloy decreased considerably, in particular with more than 30% after the deformation test at 450° C.

[0026] U.S. Pat. No. 4,726,843 discloses a rapidly solidified aluminium alloy that is described to have not only a good high-temperature strength but also a high thermal

stability coupled with good ductility and toughness of the material. The alloy does not contain any Mg and, instead of Fe and Ni, it contains, as dispersoid forming alloying elements, 2 to 5.5 wt. % of Cr and 2 to 5.5 wt. % of V. The molten alloy was rapidly solidified at a very high cooling rate and the rapidly solidified powder was subsequently consolidated by an extrusion process, followed by a heat treatment. Finely divided dispersoids were produced consisting of $\text{Al}_{13}\text{Cr}_2$ and Al_{10}V intermetallic phases having a size of less than 0.1 μm . In the Examples, a room temperature tensile strength of 520 MPa was obtained for the as extruded alloy. This as extruded alloy had a relatively small ductility, namely only a limited elongation at break of 10%. The Vickers hardness was considerably increased by precipitation hardening during the subsequent heat treatment for one or two hours at 400° C.

[0027] Binary phase diagrams of Al—Fe and Al—Ni show that the liquidus temperatures of these alloys is gently increasing with an increase of the Fe or Ni content. As was mentioned earlier, it enables rich chemical compositions for such alloys. The transition elements Cr and V as presented in U.S. Pat. No. 4,726,843 are on the contrary characterised by a very steep liquidus temperature increase as a function of the elements additions. As a result a high alloying temperature of 1150° C. was needed in U.S. Pat. No. 4,726,843 to dissolve the high amount of alloying elements. At such high temperature, vacuum or inert gas conditions needed to be applied so that this alloy is not suited for industrial production. The large volume of relatively large dispersoids formed by the high Cr and V additions did allow to achieve a high strength. However, these large dispersoids make the alloy more brittle and increase especially also the forces required to form the alloy at higher temperatures.

[0028] Also EP 0 577 944 B1 discloses a rapidly solidified aluminium alloy that contains a large amount of alloying elements including, in addition to 5 to 10 atomic % of Ni, elements such as V, Cr, Mn, Co, Mo, Nb, Ta, Mg, Cu and Zn. The aluminium alloy was produced by a rapid solidification process followed by an extrusion process at a temperature of 200 to 550° C. It had a fine grained structure with an average grain size of 40 to 2000 nm and it contained intermetallic compounds having an average grain size of 10 to 1000 nm. The obtained materials had a high tensile strength and a high hardness. Due to the brittle nature of the alloy, the flow stress of the alloy is relatively high, even at higher temperatures, so that costly, heavy forging presses have to be used. As a result of the high nickel content, the alloy is moreover not thermally stable so that the forming temperature cannot be increased. Another drawback of the presence of large amounts of Ni is that it reduces the corrosion resistance of the alloy. Finally, EP 0 577 944 B1 teaches that the amount of the matrix strengthening elements Mg, Cu and Zn should be kept below 4 atomic % since otherwise the strength of the alloy would be reduced.

[0029] U.S. Pat. No. 6,334,911 discloses an aluminium alloy that comprises Cu in combination with Cr and/or Mn, and optionally one or more of the transition elements V, Fe, Co and Ni. The alloy was rapidly solidified to obtain fine grains and was subsequently either heat-treated or consolidated and compacted. In the examples, the rapidly solidified aluminium alloy powder was extruded at a temperature of 320 to 500° C. At room temperature, the extruded materials had a relatively high tensile strength ranging from 500 to about 700 MPa. The strength of these materials was due to

solid solution strengthening, especially by the Cu, and due to the formation of quasi-crystals that were finely dispersed in the aluminium matrix and that were formed as a result of the presence of Mn and/or Cr. Moreover, when present, the transition elements produced intermetallic compound phases that were homogeneously and finely dispersed in the alloy structure. An important disadvantage of this rapidly solidified alloy is the presence of copper that negates the corrosion resistance of the alloy. Also iron and nickel have a negative effect on the corrosion resistance and also on the thermal stability of the alloy. Moreover, in order to achieve a high strength, Cr and some of the transition elements are used in the examples in relatively high amounts resulting in an alloy having a relatively high liquidus temperature requiring a high alloying temperature.

[0030] WO 2015/169232 discloses an aluminium alloy that is not produced by a rapid solidification process but that is produced by a die casting process. The alloy is strengthened by solid solution strengthening with 1 to 4 wt. % of Mn, forming Al_6Mn phases, and with 0.1 to 5 wt. % of Mg. The alloy moreover contains at least La as rare earth element to improve the metallurgical property and fluidity of the aluminium alloy and to ensure also a better anodic oxidation performance. The alloy may additionally contain limited amounts of Co and Ti that act as grain refining elements. Notwithstanding the presence of these alloying elements, the alloy compositions disclosed in the examples all had a limited ultimate tensile strength, the highest value being 300 MPa, and a limited HV hardness, the highest hardness being HV 100. The yield strength of the alloys has not been disclosed. In the examples of WO 2015/169232, the alloy with the formula $\text{Mn}2.5\text{Mg}3\text{La}0.2$ had a tensile strength of only 275 MPa, which could be increased to 300 MPa by increasing the amount of Mg to 5 wt. %. A further increase of the amount of Mg to 6 wt. % resulted however in a decrease of the tensile strength to 280 MPa so that the range of the amount of Mg was limited to 5 wt. %.

[0031] The present disclosure is directed to rapidly solidified and plastically consolidated aluminium alloys that have, as plastically consolidated, a considerably higher strength than these die cast alloys, in particular a yield strength higher than 300 MPa or preferably even higher than 340 MPa at room temperature. An object of the present disclosure is to provide an improved rapidly solidified and plastically consolidated aluminium alloy that is preferably corrosion resistant and that allows to combine a high yield strength at room temperature with a good thermal resistance and with relatively low forming forces at high temperatures. Moreover, the liquidus temperature of the alloy should not be too high in order to make alloying and solidifying the alloy easier. The liquidus temperature of the alloy should preferably be lower than 950° C. or more preferably even lower than 900° C.

[0032] The aluminium alloy according to the present disclosure is rapidly-solidified and plastically consolidated and comprises:

[0033] between 3.00 and 10.00 wt. % of magnesium (Mg);

[0034] more than 1.00 but less than 6.00 wt. % of manganese (Mn); and

[0035] one or more transition elements that are selected from the group consisting of chromium (Cr), vanadium (V), titanium (Ti), zirconium (Zr), molybdenum (Mo), cobalt (Co) and niobium (Nb), the total amount of the

one or more transition elements being at least equal to 0.50 wt. % and the maximum amounts of the transition elements being 1.50 wt. % for Cr, 1.50 wt. % for V, 1.00 wt. % for Ti, 1.00 wt. % for Zr, 1.50 wt. % for Mo, 1.50 wt. % for Co and 1.00 wt. % for Nb, wherein the aluminium alloy comprises:

[0036] no or less than 2.00 wt. %, preferably less than 1.00 wt. % of zinc (Zn), no or less than 2.00 wt. %, preferably less than 1.00 wt. % of silicon (Si), no or less than 1.00 wt. % of scandium (Sc) and no or less than 1.00 wt. % of tungsten (W);

[0037] no other elements different from Al, Mg, Mn, Cr, V, Ti, Zr, Mo, Co, Nb, Zn, Si, Sc and W or in total less than 4.00 wt. %, preferably less than 2.00 wt. % and more preferably less than 1.00 wt. % of the other elements; and aluminium as balance.

[0038] In case the alloy comprises Sc, the Sc content is preferably lower than 0.80 wt. % and more preferably lower than 0.60 wt. %.

[0039] The alloy according to the disclosure is first of all rapidly solidified. In this way the formation of primary intermetallic particles (phases) is either prevented or at least considerably reduced. The maximum amount of the different alloying elements is also limited by the respective upper limits in order to keep the required casting temperature sufficiently low. Higher amounts of these transition elements, especially of those different from Cr, increase indeed the liquidus temperature thus making alloying of the composition more difficult and less suitable for industrial processes. Higher amounts of the transition elements, and especially of Cr, may moreover result in the formation of brittle intermetallic phases that may make plastic forming processes more difficult, if not impossible.

[0040] The rapidly solidified alloy is then plastically consolidated. During the plastic consolidation process the particles (pieces) of the rapidly solidified alloy are subjected both to compressive stresses and to shear stresses. This is done at an elevated temperature, in particular at a temperature higher than 300° C. In this way, the alloy is consolidated and the grain structure thereof is refined so that, after being plastically consolidated, the alloy has a fine grained structure. By the higher temperature during the plastic consolidation, the formation of dislocations in the grains is avoided or at least reduced. Moreover, secondary intermetallic particles are produced, i.e. the so-called dispersoids. The plastic consolidation process is preferably controlled so that these dispersoids are refined, i.e. so that they have in particular a volume average size that is preferably smaller than 50 nm.

[0041] As to the chemical composition, the alloy according to the disclosure is first of all solid solution strengthened by the amount of Mg contained in the aluminium matrix. This magnesium amount is limited by an upper limit in order to avoid a too low solidus temperature of the alloy. The alloy is also solution strengthened by Mn. In contrast to Mg, Mn increases the strength of the alloy without lowering its solidus temperature. The solubility of Mn in Al is about 1.5 wt. % at 660° C. and is still 0.5 wt. % at about 535° C. At room temperature, the rapidly solidified alloy is thus supersaturated with Mn. During the plastic consolidation process, the Mn also contributes to the production of dispersoids in the rapidly solidified alloy, i.e. small particles of intermetallic compounds containing Mn, Al and other alloying elements. The formation of small dispersoids is essential for stabilizing the fine-grained alloy structure by Zener pinning.

In the amounts wherein the Mn is present in the alloy according to the present disclosure, it was found to be very effective in forming alloy strengthening/stabilizing dispersoids. In a binary Al—Mn system these dispersoids may be formed in particular by secondary Al_6Mn , which is formed above 510° C., and especially by secondary Al_{12}Mn , which is formed below 510° C., i.e. at the temperatures at which the plastic consolidation is performed.

[0042] The formation of coarser primary Al_4Mn and/or Al_6Mn is avoided or at least considerably reduced by the rapid solidification process. When formed within the grains, especially the Al_{12}Mn intermetallic phases can also provide for a strengthening effect since it is semi-coherent with the surrounding aluminium matrix. An advantage of both Mg and Mn is that they can easily be dissolved in larger amounts in the alloy composition, i.e. without requiring high alloying temperatures. The liquidus temperature of the alloy indeed decreases even when adding more magnesium whilst the liquidus temperature of a binary Al—Mn system that contains 6 wt. % of Mn, is only about 720° C.

[0043] Compared to Ni and Fe, which are used in some prior art alloys to strengthen the alloy by the formation of dispersoids, Mn has a lower diffusion coefficient so that the Mn containing dispersoids are thermally more stable. Moreover, in the alloy according to the present disclosure, the dispersoids have a greater tendency to be formed at the grain boundaries instead of within the grains, especially during the plastic consolidation process. The alloy according to the disclosure further comprises one or more of the transition elements Cr, V, Ti, Zr, Mo, Co and Nb, which have diffusion coefficients that are even lower than the diffusion coefficient of Mn. These transition elements can stabilize the Mn containing dispersoids, in particular the Al_{12}Mn phases, by partial replacement of Mn by one or more of these transition elements.

[0044] A relatively high total amount of the transition elements is required in the alloy according to the present disclosure. An important finding of the present disclosure is indeed that by the selection of the transition elements and by the relatively large total amount thereof, a larger number of smaller dispersoids is produced and the grain size of the alloy as plastically consolidated can be reduced. The larger number of smaller dispersoids produced by the presence of the transition elements was found to be able to thermally stabilize the fine-grained alloy structure notwithstanding the higher amount of interface boundary energy resulting from the increased number of grain boundaries, i.e. from the more refined grain structure. The fine-grained alloy structure contributes to a high yield strength by the Hall-Petch effect. Moreover, the fine-grained alloy structure is obtained by the plastic consolidation process, which is preferably carried out at an elevated temperature so that no or less dislocations are produced within the grains, which would reduce the ductility of the alloy.

[0045] The alloy according to the present disclosure was found to have a low flow stress, or even a superplastic behaviour, during hot deformation. Due to its high thermal stability, higher forming temperatures may be applied to even further reduce the flow stress. The low flow stress may be explained by the pronounced boundary sliding effects due to the high number of grain boundaries, by the small size of the dispersoids and by the fact that the dispersoids are predominantly present at the grain boundaries instead of within the grains.

[0046] The list from which the transition elements used in the alloy according to the present disclosure are selected comprises only transition elements that have a low diffusion coefficient in aluminium. The use of such transition elements was found to enable to produce smaller dispersoids and more refined grains during the plastic consolidation and to achieve an alloy that has a good thermal stability. The selected transition elements are also readily available in amounts and at a price that allows to produce the alloy on an industrial scale. The selected transition elements have a much lower diffusion coefficient than Ni and Fe and they also do not reduce the corrosion resistance of the alloy, in contrast to for example Ni, Fe and Cu. Surprisingly, in combination with Mg and Mn, they enable to achieve a high yield strength at room temperature combined with a low flow stress during hot deformation and an assured preservation of the fine-grained structure. They also enable to achieve a somewhat lower but still high yield strength in the as plastically consolidated state, in particular a true yield strength (YS, R_{0.2}) higher than 300 MPa, or even higher than 340 MPa, in combination with a high ductility, in particular an elongation at fracture of at least 15% or at least 18% or even higher at room temperature.

[0047] The amount of each of the transition elements is limited by a maximum value. The maximum amounts are determined by the solubility of the respective transition element in the aluminium alloy and also by the effect of an increasing amount of the transition element on the yield strength and on the hot forming stress of the alloy. Chromium can be dissolved in higher amounts than the maximum amount as claimed. However, it has been found that such higher chromium amount results in a more brittle alloy. Higher amounts of the different transition elements were not always found to increase the yield strength of the alloy but may sometimes even reduce the yield strength thereof by forming for example phases with Mg reducing the concentration of Mg in solid solution. Moreover, higher amounts of alloying elements were found to have a negative effect on the flow stress during hot deformation. The alloy according to the disclosure has to contain, in total, a minimum total amount of the one or more transition elements, more particularly a total minimum amount of at least 0.50 wt. % in order to be able to produce the minimum amount of dispersoids that are required for the minimal thermal stability and the minimal mechanical properties of the alloy.

[0048] As explained hereabove, the aluminium alloy according to the present disclosure comprises at least Mg, Mn and the one or more transition elements selected amongst Cr, V, Ti, Zr, Mo, Co and Nb as alloying elements. Preferably, it does not contain other elements. Other elements may however be contained in limited amounts in the alloy without negatively affecting its advantageous properties. In particular Zn and Si may be present in a maximum amount of 2.0 wt. % whilst Sc and W may each be present in a maximum amount of 1.0 wt. %. Zn may be added as a further solid solution strengthening element but the strengthening effect was found to be only minimal. Moreover, adding more than 1.0 wt. % of Zn was found to reduce the strength of the alloy so that preferably no more than 1.0 wt. % of Zn is added. This effect of zinc could be attributed to the formation of various phases containing Mg and Zn that deplete the solid solution from Mg. Sc is much more expensive than the other transition elements. When it is used, it is preferably used in combination with Zr in order to

stabilize the intermetallic phases containing Al, Sc and Zr. Zn and Si are relatively inert so that they may be present in a limited amount of up to 2.0 wt. %, especially when they are introduced as impurities in the alloy as a result of the use of secondary aluminium as a raw material.

[0049] In addition to the above described elements, the alloy may contain still other elements/impurities but, in the same way as in standard industrial alloys, the amount thereof should be sufficiently low, in particular in total lower than 4.0 wt. %, preferably lower than 2.0 wt. % and more preferably lower than 1.0 wt. %. The main reason is the lower corrosion resistance in their presence. Especially Cu, Fe, Ni, Bi, Sn and Pb are lowering the corrosion resistance of the alloy. In view of the advantageous effect on the mechanical properties of the alloy, the alloy may contain more of Fe and of Cu compared to the other elements, notwithstanding their negative effect on the corrosion resistance of the alloy. The alloy should preferably contain no or less than 3.00 wt. % of Fe, preferably no or less than 2.00 wt. % of Fe, and more preferably no or less than 1.00% of Fe, and no or less than 2.00 wt. % of Cu, and preferably no or less than 1.00 wt. % of Cu. The other elements, i.e. Ni, Bi, Sn and Pb, should preferably either be absent in the alloy or present in an amount of at most 0.40 wt. % or preferably at most 0.30 wt. %. Moreover, too high amounts of other alloying elements may have a negative effect on the mechanical properties of the alloy as they may disturb the solid solution strengthening effect of the Mg and the Mn and the formation of fine intermetallic dispersoids by Mn in combination with the other transition elements. Any other alloying element should therefore preferably either not be present in the alloy or each in an amount of up to at most 0.4 wt. %.

[0050] In an embodiment of the alloy according to the present disclosure, the alloy comprises at most 1.20 wt. %, preferably at most 1.00 wt. % and more preferably at most 0.75 wt. % of V.

[0051] The liquidus line of the aluminium-vanadium phase diagram is very steep so that lowering the vanadium content considerably lowers the liquidus temperature of the alloy. The present inventors have moreover found that the strength increasing effect of vanadium can be maintained by replacing part of the vanadium by one or more of the other transition elements.

[0052] In an embodiment of the alloy according to the present disclosure, or according to the preceding embodiment, the alloy comprises at most 1.40 wt. %, preferably at most 1.30 wt. % and more preferably at most 1.20 wt. % of Cr.

[0053] Compared to vanadium, much more chromium can be dissolved at a certain temperature, for example at 750° C. However, the present inventors have found that the maximum amount of chromium is preferably kept at a lower level since higher amounts of chromium may make the alloy more brittle. Moreover, the present inventors have also found that the strength increasing effect of chromium can be maintained by replacing part of the chromium by one or more of the other transition elements. In case the alloy should have a high ductility, in particular a high elongation at fracture, such as an elongation at fracture of at least 15%, it is preferably free of chromium or it should contain preferably at most 0.3 wt. %, preferably at most 0.2 wt. % of chromium.

[0054] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding

embodiments, the alloy comprises at least 0.10 wt. %, preferably at least 0.20 wt. %, more preferably at least 0.30 wt. % and most preferably at least 0.40 wt. % of Cr and/or at least 0.10 wt. %, preferably at least 0.20 wt. %, more preferably at least 0.30 wt. % and most preferably at least 0.40 wt. % of V. Preferably the alloy comprises both chromium and vanadium. It preferably comprises Cr and V in a total amount of at least 0.50 wt. %, preferably of at least 0.70 wt. % and more preferably of at least 0.90 wt. %.

[0055] Both chromium and vanadium, and especially the combination of these two transition elements, were found to be very effective in achieving the above-described advantages as to the ease of alloying, the high tensile strength and the thermal stability and hot formability. Only when the ductility of the alloy should be increased, the amount of chromium is preferably limited.

[0056] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy comprises two or more, preferably three or more of the transition elements.

[0057] A problem of the transition elements is that in the phase diagram of a binary system with aluminium they have a steep liquidus line so that the temperature required to alloy the composition quickly rises as the amount of the transition element increases. In the alloy of the present disclosure, higher liquidus temperatures can be avoided by using, for larger total amounts of transition elements, a combination of two or more of the transition elements. It has indeed been found that by replacing one transition element by a combination of two or more of the transition elements a higher total amount of transition elements can be provided in the alloy without increasing the liquidus temperature of the alloy whilst still enabling to achieve the desired thermal stability and mechanical properties. Moreover, synergistic effects on the yield strength of the alloy have been observed when using a same amount of more different alloying elements.

[0058] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy comprises Cr and/or V, preferably Cr and V, and additionally at least 0.15 wt. %, preferably at least 0.20 wt. % of Ti, at least 0.15 wt. %, preferably at least 0.20 wt. % of Zr, at least 0.15 wt. %, preferably at least 0.20 wt. % of Mo, at least 0.15 wt. %, preferably at least 0.20 wt. % of Co and/or at least 0.15 wt. %, preferably at least 0.20 wt. % of Nb.

[0059] The addition of these alloying elements enables to reduce the amounts of Cr and/or V so that the liquidus temperature of the alloy, or in other words the temperature at which the melt that is rapidly solidified has to be prepared, can be further reduced. The negative effects of Cr on the brittleness of the alloy can also be avoided. Moreover, synergistic effects on the yield strength of the alloy were observed when replacing part of the Cr and/or V by a same amount of one or more of these other transition elements.

[0060] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the total amount of the one or more transition elements is larger than 0.70 wt. %, preferably larger than 0.80 wt. %, more preferably larger than 0.90 wt. % and most preferably larger than 1.00 wt. %.

[0061] It was found that the higher the total amount of the transition elements, the higher the yield strength and the thermal stability of the alloy, at least when the alloy is

prepared by the same rapid solidification and plastic consolidation process. A higher yield strength, but a lower thermal stability, can indeed be achieved by performing the plastic consolidation at a lower temperature whilst a lower yield strength, but a higher thermal stability, can be achieved by performing the plastic consolidation at a higher temperature.

[0062] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the total amount of the one or more transition elements is smaller than 3.00 wt. %, preferably smaller than 2.75 wt. %, more preferably smaller than 2.50 wt. % and most preferably smaller than 2.25 wt. %.

[0063] It has been found that higher amounts of the transition elements produce no or only a relatively small increase of the yield strength/thermal stability. Higher amounts may even result in the formation of brittle intermetallic phases that may make plastic forming processes more difficult, if not impossible.

[0064] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy comprises at least 1.10 wt. %, preferably at least 1.20 wt. %, more preferably at least 1.30 wt. % and most preferably at least 1.40 wt. % of Mn.

[0065] Such higher amounts of Mn were found to increase the yield strength of the alloy. Since a considerable portion of the Mn will stay in solid solution, increasing the amount of Mn has a relatively high effect on the formation of dispersoids. These dispersoids are indeed only formed by the excess of Mn in the supersaturated solid solution as obtained after the rapid solidification of the alloy. Mn was found to have a high strength increasing effect.

[0066] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy comprises less than 5.00 wt. %, preferably less than 4.50 wt. % and more preferably less than 4.00 wt. % of Mn.

[0067] Below these upper limits, a substantial increase of the yield strength could be obtained by adding more Mn. However, higher amounts of Mn were found to provide no or only a small increase of the strength of the alloy. Moreover, higher amounts of Mn may result in the formation of brittle intermetallic phases that may make plastic forming processes more difficult, if not impossible.

[0068] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy comprises at least 4.00 wt. %, preferably at least 4.50 wt. %, more preferably at least 5.00 wt. %, most preferably at least 5.50 wt. % and even more preferably at least 6.00 wt. % of Mg.

[0069] Higher amounts of Mg were found to increase the strength of the alloy. Increasing the amount of Mg had no or only a minimum effect on the flow stress during hot deformation and on the thermal stability. The solubility of Mg in aluminium is indeed quite high so that the aluminium matrix is only supersaturated with Mg at relatively low temperatures, i.e. at temperatures at which the diffusion coefficient of Mg is also low.

[0070] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy comprises less than 9.00 wt. %, preferably less than 8.00 wt. % and more preferably less than 7.00 wt. % of Mg.

[0071] The Mg content of the alloy is preferably below these upper limits since the higher the Mg content, the lower the solidus temperature of the alloy. According to the binary Al—Mg phase diagram, the solidus temperature of an Al—Mg alloy containing 7 wt. % Mg would indeed be equal to about 550° C. This means that the alloy cannot be heated to higher temperatures since otherwise it would start losing its structure and thus its advantageous mechanical properties. Moreover, coarse primary brittle phases would be formed that reduce the ductility and also the strength of the alloy and that make plastic forming (hot forging) difficult if not impossible.

[0072] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy as plastically consolidated and preferably even the alloy as hot forged has a true yield strength ($YS, R_{0.2}$), measured at 20° C. in accordance with ASTM E8/E8M-13a, of at least 400 MPa, preferably of at least 450 MPa, more preferably of at least 500 MPa and most preferably of at least 550 MPa.

[0073] It has been found surprisingly that the alloy according to the present disclosure enables to achieve such high true yield strengths without a dedicated thermo-mechanical treatment that would induce dislocations and/or precipitation strengthening in the alloy thus decreasing the ductility and hot formability of the alloy and that will result in more elaborative processing of the alloy (mechanical and/or precipitation strengthening by dedicated mechanical and/or heat treatment).

[0074] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy, in particular the alloy as plastically consolidated, has a maximum true stress during plastic deformation in a compression test at a true strain rate of 10^{-2} per second, of less than 25 MPa, preferably less than 21 MPa and more preferably less than 18 MPa at 450° C., and/or of less than 20 MPa, preferably less than 17 MPa and more preferably less than 14 MPa at 500° C., and/or of less than 17 MPa, preferably less than 14 MPa and more preferably less than 8 MPa at 550° C.

[0075] Such low maximum true stresses at these temperatures enable to hot forge the alloy with very low forming forces. Advantages hereof are the possibility to use smaller forming presses and smaller and cheaper tooling and the prolonged life of the tooling.

[0076] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy, in particular the alloy as consolidated, has a true yield strength ($YS, R_{0.2}$), measured at 20° C. in accordance with ASTM E8/E8M-13a, and a maximum true stress during plastic deformation in a compression test to 50% of the initial height at a temperature of 525° C. and at a true strain rate of 10^{-2} per second, which maximum true stress, expressed in MPa, being smaller than 3.0%, preferably smaller than 2.0% and more preferably smaller than 1.5% of the true yield strength expressed in MPa.

[0077] It has been found surprisingly that the alloy according to the present disclosure enables to achieve high true yield strengths in combination with maximum true stress during plastic deformation at 525° C. that is only a very small fraction of the true yield strength of the alloy.

[0078] In an embodiment of the alloy according to the present disclosure, the alloy comprises between 1.50 and 3.50 wt. %, preferably between 2.00 and 3.00 wt. % of Mn

and between 3.00 and 5.00 wt. %, preferably between 3.00 and 4.50 wt. % of Mg, with the total amount of the one or more transition elements being less than 1.00 wt. %, and with Cr being preferably either absent or present in an amount of at most 0.3 wt. % or preferably of at most 0.2 wt. %. Preferably, the alloy has a percent elongation at fracture of at least 15%, preferably of at least 18%, and a true yield strength ($YS, R_{0.2}$), of at least 320 MPa, preferably of at least 340 MPa, both measured at 20° C. in accordance with ASTM E8/E8M-13a. The true yield strength is further preferably defined by the following equation:

$$YS > 525 - 10 \cdot el; \quad (\text{Equation 1})$$

or more preferably by the following equation:

$$YS > 530 - 10 \cdot el$$

wherein:

[0079] el is the percent elongation at fracture; and

[0080] YS is the true yield strength in MPa.

[0081] The alloy according to the present disclosure enables to combine a high tensile strength with a high elongation thus obtaining an alloy having tensile properties that are outside the area of the existing standard wrought aluminium alloys. FIG. 65 of the publication “Making sustainable aluminium by recycling scrap: The science of “dirty” alloys” by D. Raabe et al. in *Progress in Material Science* 128 (2022) displays the tensile properties covered by the existing standard wrought aluminium alloys. The yield strength of these alloys may go up to about 750 MPa whilst the elongation may go up to about 40%. However, there are no standard wrought aluminium alloys that combine a relatively high tensile strength of at least 320 MPa with an elongation of at least 15.

[0082] Such a combination of tensile properties is however quite interesting in applications such as automotive applications. The body or other parts of an automotive vehicle have to be for example sufficiently stiff/strong to provide the required intrusion resistance of the vehicle upon collision and at the same time they have to be ductile to absorb the collision energy. The ductile standard alloys have a smaller strength so that more material is needed making the vehicle heavier and the collision energy higher. In practice, patchwork can be welded to the vehicle parts to increase their stiffness or strength and partial softening can be applied to provide ductility. The alloy according to the present disclosure provides however new possibilities to make the different vehicle parts lighter, thereby also lowering the weight of the vehicle and the collision energy absorption.

[0083] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy has a liquidus temperature lower than 950° C., preferably lower than 900° C. and more preferably lower than 850° C. The liquidus temperature of the alloy is preferably higher than 600° C.

[0084] Such a low liquidus temperature can be achieved by limiting the amount of transition elements and/or by using combinations thereof. A lower liquidus temperature is

advantageous in that a lower temperature is required to melt to alloy composition so that oxidation of the molten aluminium can also be avoided more easily, especially in an industrial production process.

[0085] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy is plastically consolidated by plastic deformation under pressure at a temperature of at least 300° C. Preferably, the temperature at which the alloy has been consolidated is lower than the solidus temperature of the alloy, preferably at least 10° C. lower than the solidus temperature, but higher than 350° C., preferably higher than 375° C. and more preferably higher than 400° C., the temperature being most preferably comprised between 400° C. and 550° C.

[0086] By consolidating the alloy at a temperature below its solidus temperature, the advantages of the rapidly solidified alloy structure can be maintained. The higher temperatures during the plastic consolidation enhance the dynamic recrystallization of the alloy. The formation of dislocations is thus avoided or further reduced and the thermal stability of the alloy is enhanced. However, in most of the cases a somewhat larger grain size will be achieved when consolidating the alloy at higher temperatures and thus a somewhat smaller increase in strength. Usually a stronger alloy, having a higher (very high) yield strength, can be obtained by plastically consolidating the alloy at lower temperatures. This alloy has then, however, a somewhat lower thermal stability. Within the preferred temperature range, high strengths can be achieved and a thermal stability that enables the alloy to be hot formed at temperatures at which the alloy can easily be formed, i.e. with a considerably reduced forging pressure.

[0087] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy has, as plastically consolidated, an average grain size, measured in accordance with standard ASTM E2627-13 (2019), of less than 2000 nm, preferably of less than 1000 nm, more preferably of less than 800 nm and most preferably of less than 600 nm.

[0088] Such a fine grained alloy has the advantage that its strength is increased due to the small grain size, i.e. due to the Hall-Petch effect. Moreover, in the alloy according to the present disclosure, the smaller grain size in combination with the large amount of nano-scale dispersoids was also found to reduce the flow stress during hot formation of the alloy to very low values by activation of the grain boundary sliding process within its rich grain boundary network.

[0089] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the plastically consolidated alloy is additionally hot forged at an elevated temperature, in particular die forged, and has an average grain size, measured in accordance with standard ASTM E2627-13 (2019), of less than 4000 nm, preferably of less than 2000 nm, more preferably of less than 1500 nm and most preferably of less than 1200 nm.

[0090] Hot forging of the alloy is done at an elevated temperature in order to reduce the flow stress of the alloy and in order to be able to deform it to a larger extent than at room temperature. During the hot forging, the alloy has in particular a temperature higher than 400° C., preferably higher than 450° C., more preferably higher than 475° C. or most preferably even higher than 500° C. Heating the

consolidated alloy sufficiently uniformly to such a high temperature, and performing the forging step, requires some time so that the alloy is subjected for some time to an elevated temperature. Notwithstanding the fact that the alloy is thermally stabilized, especially the larger grains of the alloy will become coarser so that the average grain size of the alloy increases. In this embodiment, coarsening of the grains is limited thanks to the presence of multiple nanometric dispersoids within grain boundary regions, so that the final grain size still provides the required strength increasing effect.

[0091] In an embodiment of the alloy according to the present disclosure, or according to any one of the preceding embodiments, the alloy is substantially free of primary intermetallic phases and preferably also of dendrites.

[0092] This can be achieved by the rapid solidification process, in particular by rapidly solidifying the alloy at least as from its liquidus temperature to its solidus temperature.

[0093] The present disclosure also relates to a method for producing the aluminium alloy according to the present disclosure. The alloy has a liquidus and a solidus temperature and a predetermined difference between the liquidus and the solidus temperature. Both temperatures are to be determined on an equilibrium phase diagram of the alloy.

[0094] In a first step, a molten aluminium alloy composition is made that has the composition of the alloy according to the present disclosure. The molten aluminium alloy composition is rapidly solidified in the form of pieces of rapidly solidified material, and the pieces of the rapidly solidified material are plastically consolidated to produce the plastically consolidated aluminium alloy.

[0095] By rapidly solidifying the molten alloy composition, the formation of primary, usually brittle phases is avoided. Also the formation of dendrites is avoided. All of these phases/structures have negative effects on the strength and/or on the thermal stability of the alloy and also on the forces required during hot forging of the alloy. Rapid solidification of the alloy produces more uniform phases and structures that have a high amount of internal energy. During the subsequent plastic consolidation step, the pieces of the rapidly solidified alloy are not only consolidated but a fine-grained alloy structure is produced and dispersoids are formed stabilizing this fine-grained alloy structure, in particular at a lower energy level than the initial energy level of the rapidly solidified material.

[0096] After the plastic consolidation, the alloy is still stabilized at a relatively high energy level. As a result thereof, the alloy has an increased strength. Moreover, the grain boundary sliding process is more easily activated and is more pronounced thus providing for lower hot forming forces.

[0097] In an embodiment of the method according to the present disclosure, the molten aluminium alloy composition that is rapidly solidified has a liquidus temperature lower than 950° C., preferably lower than 900° C., more preferably lower than 850° C. and most preferably lower than 800° C. The liquidus temperature of the alloy is preferably higher than 600° C.

[0098] The aluminium alloy can thus be molten at relatively low temperatures, in particular at temperatures at which oxidation of the aluminium and formation of hydrogen gas can be avoided or reduced more easily without having to use a vacuum or an inert gas. In other words, due

to such low liquidus temperatures, the alloy can be produced more easily on an industrial scale.

[0099] In an embodiment of the method according to the present disclosure, or according to any one of the preceding embodiments, the molten aluminium alloy composition is ejected from at least one nozzle to be rapidly solidified, the molten aluminium alloy composition exiting the at least one nozzle having a predetermined temperature upon exiting the at least one nozzle and is rapidly solidified by being cooled down within a predetermined period of time to the solidus temperature of the aluminium alloy at an average cooling rate that is determined as the ratio of the difference between the predetermined temperature and the solidus temperature over the predetermined period of time and that is higher than 10 000° C./sec and preferably higher than 100 000° C./sec.

[0100] When applying such high cooling rates, the molten aluminium alloy is rapidly solidified as from being ejected from the nozzle.

[0101] Preferably, the predetermined temperature is at least 75% of the liquidus/solidus temperature difference higher than the solidus temperature, the predetermined temperature being preferably equal to or higher than the liquidus temperature, and being more preferably at least 10° C. higher than the liquidus temperature.

[0102] Primary intermetallic phases and dendrites can only be formed when the molten aluminium alloy has a temperature lower than its liquidus temperature. In this embodiment, the formation of primary intermetallic phases and dendrites can thus be minimized or even avoided.

[0103] In an embodiment of the method according to the preceding embodiment, the molten aluminium alloy composition is supplied through a piping to the at least one nozzle and is additionally heated in the piping before being ejected out of the nozzle, the piping is preferably substantially completely filled with the aluminium alloy composition.

[0104] When water vapour is present, aluminium oxidizes already rather quickly at relatively low temperatures. In the furnace wherein the aluminium alloy or components thereof are prepared and wherein air is present, only a relatively low temperature is preferably applied to melt the aluminium alloy or the components thereof. This temperature may be lower than the liquidus temperature so that the aluminium alloy is not yet completely molten. Once the aluminium alloy is in the piping leading to the nozzle, it is further heated to a higher temperature in this piping so that the alloy is substantially completely molten before being ejected out of the nozzle to be rapidly solidified. Since the piping is substantially completely filled with the alloy composition, the aluminium does not oxidise once it is in this piping and can thus be heated to a higher temperature without being oxidised. Oxidation can thus be avoided easily, in an industrial process, without having to apply vacuum conditions or without having to work under an inert atmosphere.

[0105] In an embodiment of the method according to the present disclosure, or according to the preceding embodiment, the pieces of rapidly solidified material are plastically consolidated by plastic deformation under pressure at a temperature of at least 300° C., preferably of at least 400° C. and more preferably of at least 450° C.

[0106] During the plastic consolidation step the pieces of the rapidly solidified material are subjected both to compressive stresses and to shear stresses. The individual rapidly solidified pieces are brought in very close contact enabling

the formation of new interatomic bonds. Plastic deformation is required to create fresh surfaces that are free of oxides and that can thus be bonded to one another. Due to the shear stresses produced by the plastic deformation, the surface to volume ratio of the particles is increased so that new, oxide free, surfaces are exposed. At the same time, smaller grains are produced. During the plastic consolidation, the rapidly solidified material preferably has a temperature higher than 300° C. In this way, smaller forces need to be exerted onto the material. Moreover, due to the dynamic recrystallization of the alloy as a result of such higher temperature during the plastic consolidation process the formation of dislocations is avoided or at least considerably reduced and the thermal stability of the alloy is enhanced.

[0107] In an embodiment of the method according to the present disclosure, or according to any one of the preceding embodiments, the temperature at which the pieces of the rapidly solidified material are consolidated is lower than the solidus temperature of the alloy, preferably at least 10° C. lower than the solidus temperature, but higher than 350° C., preferably higher than 375° C. and more preferably higher than 400° C., the temperature being most preferably comprised between 400° C. and 550° C.

[0108] Such high plastic consolidation temperatures are promoting consolidation effectiveness. Alloy flow stress is lower at higher temperatures facilitating the deformation of the pieces of rapidly solidified material that is required to achieve close contact. Additionally, higher temperatures are increasing local diffusion processes improving the effectiveness of the inter-particles bonding.

[0109] In an embodiment of the method according to the present disclosure, or according to the preceding embodiment, the pieces of rapidly solidified material are consolidated by plastic deformation to reduce the average grain size of the consolidated aluminium alloy, measured in accordance with standard ASTM E2627-13 (2019) to a value of less than 2000 nm, preferably less than 1000 nm, more preferably less than 800 nm and most preferably less than 600 nm.

[0110] Such a fine grained alloy has the advantage that its strength is increased due to the small grain size, i.e. due to the Hall-Petch effect. Moreover, with the alloy composition according to the present disclosure, the smaller grain size was also found to reduce the flow stress during hot formation of the alloy by activation of the grain boundary sliding process.

[0111] In an embodiment of the method according to the present disclosure, or according to any one of the preceding embodiments, the pieces of the rapidly solidified material are consolidated by extruding them with a cross section reduction **2** of at least 3, preferably of at least 6, more preferably of at least 8 and most preferably of at least 10.

[0112] For the consolidation process not only a compressive stress is to be exerted onto the pieces of rapidly solidified material but also shear stress. The higher the cross section reduction, the better the consolidation of the material. For lower values of the cross section reduction, a good consolidation may require one or more additional plastic processing steps, which may be provided during a subsequent hot forging step.

[0113] In an embodiment of the method according to the present disclosure, or according to any one of the preceding embodiments, the consolidated aluminium alloy is hot forged, in particular die forged.

[0114] In some applications, the alloy according to the present disclosure can be plastically consolidated in or approaching the required shape. In other applications, the shape of the part that is to be produced can be achieved or at least approached by hot forging the alloy so that no or less machining is required to achieve the final shape. The production process of the part thus requires less material, is less time and energy consuming and produces less material, in the form of chips, which needs to be recycled.

[0115] Preferably, the consolidated aluminium alloy is hot forged at a temperature higher than 400° C., preferably higher than 450° C. and more preferably higher than 475° C. or higher than 500° C., but lower than the solidus temperature of the alloy, preferably at least 10° C. lower than the solidus temperature.

[0116] The higher the forging temperature, the lower the pressure that has to be exerted onto the material to forge it. In the method according to the present disclosure such higher forging temperatures can be applied due to the thermal stability of the alloy. Moreover, the high thermal stability of the alloy allows processing the alloy in the isothermal forming regime. This means that during plastic deformation (hot forging) the temperature is kept substantially constant over the entire work piece. In particular, the temperature of the die is kept to the temperature of, or slightly below the temperature of the starting work piece. It allows better temperature control and improved material flow so that net shape forming (requiring no machining) can be achieved or approached more closely.

BRIEF DESCRIPTION OF THE DRAWINGS

[0117] Other advantages and particularities of the present disclosure will become apparent from the following description of some particular embodiments of the aluminium alloy and the method for producing it according to the disclosure. This description is only given by way of example and is not intended to limit the scope of the disclosure. The reference numerals used in the description relate to the annexed drawings wherein:

[0118] FIG. 1 shows schematically a possible embodiment of an installation for producing the molten aluminium alloy and for rapidly solidifying this alloy by a melt spinning device;

[0119] FIG. 2 is an SEM image showing the microstructure of the Al-7Mg-1V alloy with coarse primary Al₃V intermetallic phases produced in the RS example wherein the alloy composition was rapidly solidified at a temperature of 650° C.;

[0120] FIGS. 3 and 4 are similar to FIG. 2 but show the microstructure of the Al-7Mg-1V alloy rapidly solidified at a temperature of 750° C. and 850° C. and showing less or no coarse primary Al₃V intermetallic phases;

[0121] FIGS. 5 to 7 are TEM images, with an increasing magnification, of the Al-7Mg-2Mn-1V-1Cr alloy plastically consolidated at a temperature of 400° C.; and

[0122] FIG. 8 is a grain orientation map obtained from a TKD analysis of the Al-7Mg-2Mn-1V-1Cr alloy plastically consolidated at a temperature of 400° C.

DETAILED DESCRIPTION

[0123] The disclosure generally relates to a new aluminium alloy, in particular a wrought aluminium alloy, which is rapidly solidified and plastically consolidated. The

aluminium alloy has a liquidus and a solidus temperature. The liquidus temperature is the lowest temperature at which the alloy, in an equilibrium state, is completely liquid while the solidus temperature is the highest temperature at which the alloy, in an equilibrium state, is completely solid. Both temperatures can be seen on the equilibrium phase diagram of the alloy. The microstructure and the composition of the alloy according to the disclosure enables to achieve a high yield strength at room temperature combined with a low hot deformation stress so that is can be easily formed by a hot forging, in particular by a hot die forging process. A high ductility can also be achieved in combination with a yield strength that is still high, in particular higher than 300 MPa. The rapidly solidified and plastically consolidated alloy has a quite high thermal stability so that its yield strength is not lost, or only to a limited extend, during a hot forging process. The liquidus temperature of the new aluminium alloy can be kept low in order to make alloying and solidifying the alloy easier, especially on an industrial scale.

[0124] The alloy may be a monolithic material and can be used as such to produce aluminium parts. However, it is also possible to include the alloy in a metal matrix composite (MMC) or in a metal matrix nanocomposite (MMNC) material. A metal matrix composite is a composite material with at least two constituent parts, one being the aluminium alloy. The other material may be a different metal or may be another material, such as a ceramic or an organic compound. MMCs are made by dispersing reinforcing material into a metal matrix. For example carbon fibres are commonly used in aluminium matrix. The matrix, i.e. the aluminium alloy, is the monolithic material into which the reinforcement is embedded and is completely continuous. MMNCs can be defined as a metal matrix composite that is reinforced by nano-reinforcements. The reinforcements in MMCs or MMNCs do not form part of the aluminium alloy that forms the matrix. The composition of the aluminium alloy according to the present disclosure is therefore given in percent by weight of the composing alloy elements not including any reinforcements when the aluminium alloy would be part of an MMC or an MMNC. Also the phase diagram defining the liquidus and solidus temperature of the alloy is only based on the aluminium alloy forming the monolithic matrix material.

[0125] The alloy according to the present disclosure comprises aluminium and different alloying elements, including alloying elements that provide for a solid solution strengthening effect and alloying elements that are intended to produce small, nanoscale dispersoids that assist in obtaining a fine grained alloy structure and that stabilize this fine grained structure. The refined grain structure provides for an additional strengthening effect enabling to achieve, in combination with the solid solution strengthening effect, a high strength. In this way, no cold working of the alloy is necessary to increase the strength thereof nor any separate/dedicated precipitation hardening of the aluminium matrix within the grains by additional strengthening heat treatment operations. Cold working and separate precipitation hardening operations are preferably avoided or kept to a minimum since both of these strengthening mechanisms have a negative effect on the plasticity/ductility of the alloy and delivers additional effort (cost) to production routines.

[0126] In general, the aluminium alloy according to the present disclosure comprises between 3.00 and 10.00 wt. % of Mg and between 1.00 and 6.00 wt. % of Mn. For higher

strengths, it preferably comprises at least 4.00 wt. % of Mg. For lower strengths, but higher ductilities, it may comprise less Mg, in particular 3.00 wt. % or more or 3.5 wt. % or more Mg. The alloy moreover comprises one or more transition elements that are selected from the group consisting of Cr in a maximum amount of 1.50 wt. %, V in a maximum amount of 1.50 wt. %, Ti in a maximum amount of 1.00 wt. %, zirconium Zr in a maximum amount of 1.00 wt. %, Mo in a maximum amount of 1.50 wt. %, Co in a maximum of 1.50 wt. %, and Nb in a maximum amount of 1.00 wt. %. The total amount of these transition elements is at least 0.50 wt. %. Other elements may be present in the alloy but the alloy should contain no or less than 2.00 wt. %, preferably less than 1.00 wt. % of Zn, no or less than 2.00 wt. %, preferably less than 1.00 wt. % of Si, no or less than 1.00 wt. % of Sc and no or less than 1.00 wt. % of W. Finally, the alloy should contain no other elements different from Al, Mg, Mn, Cr, V, Ti, Zr, Mo, Co, Nb, Zn, Si, Sc and W or in total less than 4.00 wt. % of such other elements.

[0127] The aluminium alloy of the present disclosure is produced by preparing a molten aluminium alloy composition, by rapidly solidifying the molten aluminium alloy composition in the form of pieces of rapidly solidified material, and by plastically consolidating the pieces of rapidly solidified material. The plastically consolidated alloy is then preferably hot forged and optionally further machined to produce the desired parts.

[0128] Rapid solidification of the molten alloy composition is a mandatory step in the production of the alloy according to the present disclosure. The key element is to prevent the growth of inter-metallic phases during crystallisation of the molten alloy composition as well as segregation of the alloying elements to dendrite regions. In the alloy according to the present disclosure these phenomena are avoided by increasing the crystallisation rate by a fast cooling process. The cooling rate that can be achieved is limited physically by heat transfer phenomena and the heat capacity of the material. The dimensions of the material that is subjected to the cooling process has to be limited. In fact, in order to reach the required chemical composition very high cooling speeds are required that can be obtained by producing highly fragmented pieces of molten material. Two technologies can be used: powder atomisation and melt spinning. Both are characterised by a high surface to volume ratio enabling a fast heat extraction. Especially, a melt spinning process enables a more efficient heat extraction by making use of the efficient heat transfer between the molten alloy composition and a cold drum surface, preferably a copper drum surface. As a result of the rapid solidification technique, a solidified/crystallised material is produced in the form of a highly fragmented material, consisting of pieces having at least one dimension that is smaller than 100 µm. Preferably, the pieces of solidified material have at least one dimension that is even smaller, in particular smaller than 80 µm or even smaller than 60 µm.

[0129] By the expression "rapidly solidified" is meant that when being solidified the molten aluminium alloy is cooled down at an average cooling rate that is higher than 10 000° C./sec (104° C./sec) and preferably higher than 100 000° C./sec (105° C./sec). The average cooling rate is in particular calculated over a time interval starting from the moment the accelerated extraction of heat from the liquid alloy composition has started until the average temperature of the alloy composition has dropped to its solidus temperature. The

accelerated extraction of heat is preferably started as from an average temperature of the alloy composition that is equal to or higher than the liquidus temperature of the alloy composition. The average cooling rate is preferably higher than 10 000° C./sec (104° C./sec) and more preferably higher than 100 000° C./sec (105° C./sec) over the time interval between the moment in time when the alloy composition is at its liquidus temperature and the moment in time when the temperature of the alloy composition has dropped to its solidus temperature.

[0130] In practice, the molten aluminium alloy is rapidly solidified as from the moment it is ejected from one or more nozzles. As explained hereabove, the molten aluminium alloy ejected from the nozzle can be solidified, in a melt spinning process, in the form of ribbons or, in a powder atomisation process, in the form of a powder.

[0131] Rapid solidification starts when the molten aluminium alloy exits the nozzle or nozzles. At the outlet of the nozzle, the molten aluminium alloy has a predetermined temperature that is substantially uniform. If not uniform, this predetermined temperature is the volume weighted average temperature of the molten aluminium alloy upon exiting the nozzle. The average cooling rate has to be calculated over the interval wherein the volume-weighted average temperature of the molten aluminium alloy drops from the predetermined temperature at the outlet of the nozzle to the solidus temperature of the aluminium alloy. The difference between these two temperatures divided by the time it takes to cool the molten aluminium alloy down from the predetermined temperature to the solidus temperature is the average cooling rate that has to be higher than 10 000° C./sec (104° C. per second) to have a rapid solidification process.

[0132] When appropriate measurement devices are available, this average cooling rate can be measured but it can also be calculated based on different parameters. The required cooling rate can also be determined experimentally by testing incrementally increasing cooling rates starting from a cooling rate at which primary intermetallic phases are formed until a cooling rate is reached at which primary intermetallic phases are no longer formed during the rapid solidification step.

[0133] For a rapid solidification process by melt spinning the average cooling rate can be calculated based on calculations as disclosed in the article "Analyses of the melt cooling rate in the melt-spinning process" of B. Karpe et al. in *Journal of Achievements in Materials and Manufacturing Engineering*, Vol. 51, Issue 2, April 2012. The content of this article is incorporated herein by way of reference. In a number of the figures of this articles, the initial quick drop of the temperature from about 700° C. to 660° C., i.e. to the solidus temperature of pure aluminium, can be seen and this for different ribbon thicknesses and different distances from the surface of the chill wheel. When dividing this temperature drop by the solidification time, and averaging these values for the different distances from the surface of the chill wheel, the average cooling rate can be calculated.

[0134] For a rapid solidification process by gas (or liquid) atomisation the average cooling rate can be calculated based on calculations as disclosed in the article "Rapidly Solidified Gas-Atomized Aluminium Alloys Compared with Conventionally Cast Counterparts: Implications for Cold Spray Materials Consolidation" by Bryer C. Sousa et al., *Coatings* 2020, 10, 1035. The content of this article is incorporated

herein by way of reference. The formula enabling to calculate the cooling rate is given in this article. In FIG. 3 of this article it can be seen that the cooling rate is in particular dependent from the particle size of the atomised droplets. Other parameters are the specific heat of the metal droplet and the thermal conductivity of the gaseous species utilized during gas atomisation.

[0135] FIG. 1 shows schematically an installation that can be used to produce the rapidly solidified aluminium alloy by a melt spinning process. This installation comprises a furnace 1 that is provided with a heater 2. As illustrated in FIG. 1 this heater 2 may be an induction heater 2. Other heaters are however also possible, for example gas heaters. The different elements of the aluminium alloy composition are introduced in this furnace, either as pure elements or as mixtures (master alloys) to produce the aluminium alloy composition 3 in the furnace 1. The aluminium may be added as substantially pure aluminium whilst the other alloying elements are preferably added in the form of a master alloy (that also contains a portion of the aluminium).

[0136] The furnace 1 is connected by a piping 4 to a nozzle 5 and can be closed by means of a valve that is not shown in FIG. 1. The piping 4 comprises one pipe 6. This pipe 6 is provided with a further heater 7, which is again an induction heater 7 provided around the pipe 6. The pipe 6 has a widened section forming a chamber 11 that is also heated by a heater 7. The aluminium alloy composition 3 is heated to a first temperature in the furnace 1 and is fed at this temperature to the inlet of the piping 4. The furnace 1 and the piping 4 are arranged vertically one above the other so that the molten alloy composition 3 can flow by gravity from the furnace 1 to the nozzle 5. An extra pressure can be applied onto the molten composition 3 by closing the furnace 1 by means of a lid 8 and by introducing gas under pressure into the furnace 1. Alternatively, a pump (not shown) may be provided in the piping 4. The piping 4, and also the chamber 11 formed by the piping 4, is preferably completely filled with the molten alloy composition so that the molten composition flows by pipe-flow through the piping. In this way, no or substantially no gas, for example air containing water vapour, is present in the piping so that the alloy composition is not subjected to any oxidation processes within the piping 4.

[0137] The molten aluminium alloy that is ejected from the nozzle 5 arrives on a rotatable chill roll 9 that may be made for example of copper, copper-beryllium or stainless steel. The roll 9 is cooled internally, for example by means of water, and is rapidly rotated to achieve the required fast cooling rates. The liquid aluminium alloy is solidified in the form of a ribbon 10, or in the form of several ribbons in case more nozzles 5 are provided at the end of the piping 4 (that may contain more than one or several pipes 6 in parallel). The ribbon thickness can be controlled by the rotational speed of the chill roll 9, the ejection pressure, the nozzle slot size and the gap between the nozzle 5 and the roll 9. Higher cooling rates can be achieved by reducing the ribbon thickness, in particular to the above-described preferred maximum dimensions of 1.00, 0.80 or 0.60 µm. The ribbon may have a width of one to several millimetres, for example a width of between 1 to 10 mm, in particular between 1 and 5 mm.

[0138] Instead of using a chill roll 9 for rapidly solidifying the molten alloy, it can also be solidified rapidly by other existing methods, in particular by a spray forming process

wherein the molten alloy is atomized in the form of droplets out of the nozzle to produce a powder. The droplets are either cooled in the air or with a liquid such as water (in particular in accordance with a granulation technique) in order to further increase the cooling rate. The nozzle may also eject the molten alloy in water, in particular in accordance with the known in-rotating water quenching technique.

[0139] Instead of producing the molten alloy composition in one furnace 1, it is possible to inject one or more of the other alloying elements in the piping 4, or in a mixing chamber provided in the piping 4, which is part of the piping 4 and which is thus also preferably completely filled with the molten alloy composition.

[0140] An important advantage of the additional heating step by means of the further heater 7 is that the aluminium alloy composition contained in the furnace 1 does not need to be molten completely, and has in particular not to be heated to its liquidus temperature. When some of the alloying elements are injected in the piping 4, the liquidus temperature of the fraction of the alloy composition contained in the furnace 1 may moreover be reduced, optionally to such an extent that this fraction of the alloy composition may be heated to its (lower) liquidus temperature whilst still avoiding or minimizing oxidation of the aluminium in the furnace. In this way no vacuum or no inert gas has to be applied in the furnace 1. By the additional heating step in the piping 4, and especially in the chamber 11 comprised therein, the entire alloy composition is then heated to a higher temperature at which the entire alloy composition is completely, or substantially completely, molten before being ejected out of the nozzle 5. The chamber 11 is intended to increase the residence time of the alloy in the piping 4. The increased residence time enables any alloying element that has not yet been completely molten/dissolved in the molten alloy to further dissolve therein. Preferably, the molten aluminium alloy is heated in the piping to a temperature that is equal to or higher than its liquidus temperature, preferably to a temperature that is at least 10° C. or even 20° C. higher than its liquidus temperature.

[0141] The molten aluminium alloy composition that is ejected from the nozzle 5 has a predetermined temperature upon exiting the nozzle and is rapidly solidified by being cooled down within a predetermined period of time to the solidus temperature of the aluminium alloy. For a rapid solidification process, the average cooling rate, which is determined as the ratio of the difference between the predetermined temperature and the solidus temperature over the predetermined period of time, should be higher than 10 000° C./sec (104° C./sec) and preferably even higher than 100 000° C./sec (105° C./sec). A rapidly solidified aluminium alloy is an alloy that is rapidly solidified as from a temperature that is at least 75% of the difference between the liquidus and the solidus temperature of the aluminium alloy higher than its solidus temperature. When being ejected from a nozzle, the molten aluminium alloy exiting the nozzle should thus have such a high predetermined temperature. This predetermined temperature is preferably equal to or higher than the liquidus temperature of the alloy, and more preferably at least 10° C. higher than this liquidus temperature. When rapidly solidifying the aluminium alloy from such a high temperature, the formation of primary interme-

talic phases can be avoided so that the alloy is substantially free of primary intermetallic phases and preferably also of dendrites.

[0142] Since the rapidly solidified material is produced in the form of pieces, in particular of ribbons 10, of rapidly solidified material, those pieces of rapidly solidified material have to be plastically consolidated in a next step to produce solid pieces of the alloy that can be further processed, by hot forging and/or machining, into the final articles. When produced in the form of ribbons, the pieces of rapidly solidified material could be first chopped into smaller pieces. In this way, the particles/pieces of rapidly solidified material can more easily be compacted to remove the gaps/voids between these particles.

[0143] Plastic consolidation is a required step to convert the fragmented rapidly solidified material into the useful consolidated bulk form, i.e. into solid pieces of the alloy material. To achieve this, the separate pieces (particles) of solidified material have to be put together in a very close contact enabling the formation of new interatomic bonds. A very close contact is however not sufficient as such since aluminium and its alloys form continuous and tight layers of oxides on their surfaces (that is often called aluminium passivation). In other words because of these oxidation processes there is no fresh metal surface available for correct bonding. A second requirement that thus has to be fulfilled during plastic consolidation is the creation of oxide free surface portions on the particles. During the plastic consolidation step, the separate particles are preferably compressed together in an enclosed space and subjected to forces forcing them to change their initial shape. By doing so, the two above mentioned requirements are met simultaneously. From a physical point of view the plastic consolidation process is characterised by a stress component having a large hydrostatic component to press the particles closely together and a shear component deforming the particles. When pressing the particles together, they are somewhat deformed to remove the gaps between the particles. As a result of the shear stresses, however, the surface to volume ratio of the particles is always increasing thus creating/exposing new oxide free surface portions.

[0144] The easiest way to obtain the combination of the required pressure and shear stresses on an industrial scale is by an extrusion process. Preferably, the pieces of rapidly solidified material are first compacted to form a billet that is preheated and then extruded through a die opening. An important parameter of the extrusion process is the cross section reduction ratio 2, which is the ratio between the cross sectional area of the billet to the cross sectional area of the extruded part leaving the die opening. The cross section reduction area is preferably at least equal to 10. Smaller cross section reduction areas are however also possible, especially in case the extruded aluminium alloy is subsequently hot forged, in particular die forged. During such hot deformation steps, the produced part is indeed further plastically consolidated. The extrusion process can moreover be followed by one or more further plastic consolidation processes.

[0145] Instead of an extrusion process, or in addition thereto, other consolidation techniques can be applied, for example multistep forging, ECAP (Equal Channel Angular Pressing) or any other techniques characterised by the above described tensor can be used for plastic consolidation purposes. In particular ECAP can be used after the extrusion

process, especially when the extrusion process was done with a small cross section reduction ratio. The ECAP process can be carried out directly onto the extruded parts so that they are still hot. During the ECAP process material is pressed through a die consisting of two intersecting channels with identical cross sections. An advantage of the ECAP process is that the cross section of the part is thus not changed, in particular not reduced. In the ECAP process, the part is subjected to severe plastic deformation that contributes not only for grain refinement but also for additional fracture and refinement of the oxide layers between the aluminium pieces/particles. More than one ECAP step can be carried out successively.

[0146] Before subjecting the rapidly solidified material to the plastic consolidation step it is heated to a predetermined temperature, i.e. to the temperature at which the plastic consolidation process is carried out. The indicated temperatures are the volume-weighted average temperatures of the rapidly solidified material, in particular of the billet of the rapidly solidified material that is being plastically consolidated, in particular extruded. Higher plastic consolidation temperatures are promoting consolidation effectiveness. Alloy flow stress is lower at higher temperatures, facilitating the deformation of the initial particles (pieces of the rapidly solidified material) to achieve close contact and produce new oxide free surface portions. Additionally higher temperatures are increasing local diffusion processes improving the effectiveness of the inter-particles bonding. The ultimate upper limit for the consolidation temperature is connected to the solidus temperature of the alloy. For the alloys according to the present disclosure, the magnesium content is controlling the lowest melting point (solidus temperature) of the alloy, i.e. the temperature at which the alloy starts to melt when being heated up. During all processing steps after the rapid solidification, the material has to be in the solid state. Any local melting will cause close to equilibrium structure formation, characterised by chemical segregations and by the formation of larger inter-metallic phases. In case of the alloy according to the present disclosure, magnesium is controlling the lowest value of the melting temperature range. Particularly, depending on the magnesium concentration the melting point (solidus temperature) of the alloys is in the range of about 550° C. to about 590° C. for magnesium concentrations between 4 and 7 wt. % and is decreasing when the magnesium concentration increases. An additional constraint on the plastic consolidation temperature is imposed by the diffusion processes. A higher temperature means more intense diffusion and faster coarsening of the structure of the alloy. The preferred temperature processing window for the plastic consolidation is between 400° C. and 500° C. Within this preferred temperature range the flow stresses are small enough for effective plastic consolidation in a typical industrial press, and diffusion is not too severe to destroy the highly refined dispersoid based microstructure of the alloy.

[0147] As a result of the plastic consolidation a highly refined microstructure of the consolidated material is achieved. Different processes of structure refinement are acting at the same time. Deformations imposed by the external forces are creating large amounts of dislocations that are the driving forces for recrystallization processes. This is called dynamic recrystallization. The final state of the microstructure is dependent on the intensity of these processes: the imposed deformation tending to decrease the

grain size whilst the higher temperature tending to increase it. The alloy composition according to the present disclosure enables however to shift the balance towards smaller grains by the thermal stabilization of the alloy structure. In the alloy according to the present disclosure small inter-metallic phases/dispersoids are indeed produced as grain boundary stabilising elements. The more of such nano-metric phases that are present in the microstructure, the better the pinning effect that can be achieved. Those phases/dispersoids are also formed in situ during the plastic consolidation process. The aluminium matrix is oversaturated in the alloying components, and where suitable conditions are given (high temperature and diffusion paths) dispersoids will form. The low diffusion coefficient of the alloying elements considerably limits the growth of the dispersoids. The final grain size thus depends upon the processing conditions as well as on the number and size of dispersoids. With the alloy composition according to the present disclosure, grain sizes in the range of 500 nm can easily be obtained. Such small grain sizes significantly boost the mechanical strength of the alloy in cold condition. Moreover, low flow stresses can be achieved at higher temperatures, i.e. during the hot forging steps.

[0148] The plastic consolidation process may result directly into the final component that is in particular in the form of an extruded profile. Usually, an additional hot forging/forming step will however be applied. During hot forming, the same temperature constrains apply as during the plastic consolidation process. The same mechanisms apply, where microstructure is dynamically rebuilt during deformation processes (dynamic recrystallization). The alloy according to the present disclosure enables however to apply higher temperatures during the forming step in order to make the forming step easier. Although some strength will be lost when applying higher forming temperatures, the initial strength of the plastically consolidated alloy can be sufficiently high to allow the use of such higher forming temperatures. By the resulting coarsening of the alloy structure, the thermal stability of the final part is moreover increased to some extent.

[0149] The finer grain structure of the plastically consolidated alloy allows to activate grain boundary sliding processes during the next hot deformation steps. Positive aspects include smaller forming presses, smaller and cheaper tooling and prolonged life of tooling. The finer grain structure also enables to obtain a ductile alloy that still has a high strength. The ductility of the alloy can in particular be increased by reducing the amount of Mg and by optionally increasing the amount of Mn. Additionally the higher thermal stability provided by the finely dispersed alloying elements allows processing in an isothermal forming regime. This means that during the plastic deformation, the temperature can be kept nearly constant. In this way the temperature can be better controlled and an improved material flow can be achieved thus enabling near net shape forming.

[0150] During the forming steps after the plastic consolidation the microstructure of the alloy will normally be coarsened to some extent. During hot deformation diffusion processes are intensified (there are more diffusion paths active, in particular dislocation cores and new grain boundaries) leading to dispersoid and grain coarsening. The final effect is dependent on the processing conditions. After the plastic consolidation, the alloy preferably has an average

grain size of less than 2000 nm, preferably of less than 1000 nm, more preferably of less than 800 nm and most preferably of less than 600 nm. In combination with the small size of the dispersoids, such a fine grain structure enables to activate grain boundary sliding during hot deformation. After the hot forging step, the average grain size of the alloy is preferably still smaller than 4000 nm, preferably smaller than 2000 nm, more preferably smaller than 1500 nm and most preferably smaller than 1200 nm. After the hot forging step, the small grain size is especially important to improve the mechanical strength of the alloy.

[0151] The strength of the alloy according to the disclosure is directly connected to two strengthening mechanisms, namely to solid solution strengthening and grain size strengthening (following the Hall-Petch rule). When a grain size is produced that is in the order of magnitude of 500 nm, a yield strength higher than 600 MPa can easily be achieved with an alloy according to the present disclosure. Since the alloy is produced under hot conditions, the dislocation density is low and the plasticity of the alloy is not deteriorated. For lower yield strengths, higher ductilities can thus also be obtained. This is in contrast to cold forming/working steps during which the dislocation density increases leading to strengthening of the material but at the same time to a loss of plasticity. Also the dispersoids produced during the plastic consolidation process are preferentially produced at the grain boundaries to stabilize the fine grained structure and to enable grain boundary sliding. Precipitation hardening of the aluminium matrix as such is thus avoided or at least minimized. This is advantageous in that precipitation strengthening of the alloy has a negative effect on the flow stress of the alloy during hot deformation whilst grain size strengthening of the alloy reduces the hot flow stress of the alloy and may give the alloy even superplastic properties. Dedicated precipitation hardening steps can be applied but are preferably omitted. Lack of precipitation strengthening routine results in additional benefits as usually performed in standard alloys, requires additional heat treatment steps, increasing cost of the products by additional time, infrastructure and energy use. Moreover, alloys with a higher ductility can be obtained.

[0152] The following examples illustrate some embodiments of alloys of the present disclosure, a production method thereof and some of the advantageous properties that can be provided by the alloys according to the disclosure.

Example of Alloying and Rapid Solidification Technique

[0153] A ternary alloy composition was made consisting of aluminium, 7 wt. % of magnesium and 1 wt. % of vanadium. The alloy composition was made and rapidly solidified in an installation as shown schematically in FIG. 1.

[0154] Based on the binary Al-V phase diagram the alloy would have, in the absence of magnesium, a liquidus temperature of about 820° C. The aluminium, magnesium and the vanadium were applied in the furnace 1 and were heated therein to a temperature of 650° C. This temperature is lower than the solidus temperature of pure aluminium (660° C.) but the solidus temperature of the aluminium/magnesium solid solution is considerably lower (only about 550° C.). The alloy composition was only partially molten and contained, based on the phase diagram, an amount of undissolved Al₃V intermetallic phases.

[0155] The partially molten alloy composition was further heated in the chamber 11 by means of the further heater 7 and was then rapidly solidified by melt spinning onto the water cooled copper chill roll 9. The produced ribbons 10 had a thickness of about 50 µm and a width of about 3 mm. The residence time of the alloy in the chamber 11 was equal to about 30 s.

[0156] FIG. 2 shows the microstructure of the rapidly solidified alloy composition that has not been extra heated by means of the heater 7. A high number of primary Al₃V intermetallic phases with a maximum diameter not higher than 20 µm can be seen.

[0157] FIGS. 3 and 4 show the microstructures of the rapidly solidified alloy compositions that have been extra heated by means of the heater 7 to a temperature of 750° C. and 850° C. respectively. In the structure of FIG. 3 there are a lot less (only one) particles formed by Al₃V intermetallic phases whilst in FIG. 4 there are no longer intermetallic particles larger than 1 µm. As a matter of fact, the alloy was heated to a temperature higher than its liquidus temperature so that the molten alloy did not contain any primary intermetallic phases. The alloy was moreover rapidly solidified

so that no primary intermetallic phases could be formed during the crystallization phase, i.e. when cooling down from its liquidus temperature to its solidus temperature.

[0158] In this last experiment, the distance between the location where the liquid alloy composition was applied onto the roll 9 and the solidification front was determined with a camera. It took about 0.00025 s for the alloy to reach the solidification front. The temperature of the alloy composition dropped in this period of time on the roll 9 from about 850° C. to about 550° C., thus at an average cooling rate that could be estimated at about 1 200 000° C./s.

Examples of Alloy Compositions and Mechanical Properties

[0159] A number of different alloys according to the present disclosure and a number of comparative alloys have been prepared. The composition and the mechanical strength of the alloys are presented in the table 1.

[0160] Tables 1a and b: Composition (in wt. %) and mechanical strength (HV: hardness in Vickers Hardness, YS, R_{0.2}: true yield strength in MPa) of rapidly solidified and plastically consolidated alloys, including alloys according to the present disclosure and comparative alloys.

TABLE 1a

	Alloy No.								
	1 (comp)	2 (comp)	3 (comp)	4	5	6	7 (comp)	8 (comp)	9 (comp)
Al	Bal	Bal	Bal	Bal	Bal	Bal	Bal	Bal	Bal
Mg	5	5	5	5	5	6	7	7	7
Mn				2	2	2			
Zn					2				
V		1		1	1	1			1
Cr			1	1	1	1			1
Zr									
Ti									
Co									
Sum	0	1	1	2	4	2	0	1	1
HV	91	109	126	184	176	192	98	132	144
YS	205	264	310	586	573	604	211	410	415

TABLE 1b

	Alloy No.											
	10	11	12	13	14	15	16	17	18	19	20	21
Al	Bal	Bal	Bal	Bal	Bal	Bal	Bal	Bal	Bal	Bal	Bal	Bal
Mg	7	7	7	7	7	7	7	7	7	7	7	7
Mn	2	2	2	2	2	2	2	2	3.5	2	2	2
Zn												
V	1	1	0.5	1	0.5	0.5	0.25	0.5	1	0.5	0.5	1
Cr		1	0.5	0.5	0.5	1	1	1	1	0.5	0.5	0.5
Zr				0.5	0.5					0.25		
Ti						0.5	0.25		0.2			
Co										0.25	0.25	
Sum	1	2	1	1.5	1.5	2	1.75	1.75	2	1.45	1.25	1.75
HV	183	200	186	191	192	202	195	193	223	193	190	196
YS	552	624	584	581	629	650	636	625	719	594	600	646

[0161] Alloy numbers **20** and **21** could also be produced by replacing all of the Co or a portion thereof with similar amounts of Mo or Nb. This results in alloys with a similar hardness and yield strength as alloys numbers **20** and **21**.

[0162] All compositions are given in weight percent with aluminium balanced to 100%. The row "sum" indicated the total amount of dispersoid forming elements, i.e. of all the elements different from Al, Mg and Mn. The alloys prepared with the indicated composition were heated to a temperature above their liquidus temperature and were rapidly solidified by casting on a copper drum and then extruded at a temperature of 400° C. with an area reduction ration of **16**. The composition was varied to show the influence of the solid solution strengthening elements as well as of the dispersoid forming elements on the mechanical properties expressed in terms of material hardness (determined by compression) and yield strength (measured by an elongation test).

[0163] First of all, the yield strengths of the different alloys were found to show a linear correlation with the hardness values, more particularly according to the equation:

$$YS = 4.0876 HV - 175.65,$$

[0164] with R² being equal to 0.98.

[0165] The Vickers' hardness values are thus a good indication of the tensile strength of the alloys.

[0166] Table 1 shows that the highest strength is achieved by simultaneous addition of solid solution strengthening elements and of dispersoid forming elements. The data in the table allow to analyze the influence of the particular elements on the mechanical properties of the material.

[0167] For example, keeping the amounts of the other elements constant, an increasing Mg content of 5 wt. %, 6 wt. % and 7 wt. % (see alloys Nos. 4, 6, 11), is giving hardness values of 184HV, 192HV and 200HV respectively. Increase of the Mg content is thus giving clear advantages in terms of the material strength. It is also important to mention that an increase of the Mg content was also found to prevent grain growth during hot forming processes, leading to the formation of smaller grain sizes.

[0168] A similar effect can be observed for an increasing content of Mn. For example, adding 2 wt. % Mn to the alloy containing 7 wt. % Mg and 1 wt. % V improves the hardness from 132HV to 183HV (see alloys Nos. 8 and 10) whilst an increase of the Mn content from 2 to 3.5 wt. % increases the hardness from 200HV to 223HV (see alloy Nos. 11 and 18). Apart from its solid solution strengthening effect, Mn also creates small dispersoids leading to an additional increase of the strength.

[0169] Not all elements were found to have a positive effect on the solid solution strengthening. Comparing alloys 4 and 5, it can be seen that addition of a considerable amount of Zn had a negative effect on the final mechanical properties. Such effect can be attributed to the formation of various phases containing Mg and Zn that may provide in other alloys for a solution strengthening and precipitation hardening of the alloy but which in the alloys according to the present disclosure effectively depletes the solid solution from Mg content. Other elements that are typically added to

aluminium alloys, such as for example copper, have not been tested since they have a negative impact on the corrosion resistance of the alloy.

[0170] A similar analysis can be performed for the dispersoid forming elements. It is clear that an increase of their concentration leads to a significant improvement of the hardness. Based on the data of the different alloys that contain 7 wt. % Mg and 2 wt. % Mn (alloy Nos. 10-17 and 19-21), the following linear relationship was found between the "sum" of the amounts of the different dispersoid forming elements (V, Cr, Zr, Ti and Co) and the hardness of the alloy:

$$HV = 13.965 \text{ sum} + 171.53,$$

with R² being equal to 0.89.

[0171] Based on this relationship, a minimum amount of 0.5 wt. % of dispersoid forming transition elements would enable to achieve a hardness of about 180HV, or a corresponding yield strength of about 560 MPa.

[0172] The effect on the hardness value (and thus also on the tensile strength of the alloy) does not only depend on the amount of dispersoid forming elements but also on the element itself and on the specific combinations thereof (this is why the R² value for the linear equation is somewhat smaller). For example from alloy Nos. 2 and 3 and also from alloy Nos. 8 and 9 it appears that Cr has a somewhat higher strengthening effect than V.

[0173] It is important to stress out that the dispersoid forming elements do not only have an effect on the strength of the alloy but that they also provide for the required grain stabilizing effect. It is also an important finding, that several dispersoid forming elements can be used simultaneously, each in a smaller content, making it thus possible to decrease the required casting temperature of the alloy (i.e. the liquidus temperature). The highest strengthening effects are obtained with compositions wherein the sum of the dispersoid forming elements is at least equal to 2 wt. %. According to the binary Al—Cr and Al—V phase diagrams, 2 wt. % of Cr would result in an alloy having a liquidus temperature of 780° C. whilst 1.7 wt. % of V would result in an alloy having a liquidus temperature of about 950° C. By using 2% of a combination of Cr and V, the liquidus temperature, and thus the required casting temperature, can be reduced.

[0174] The liquidus temperature can further be reduced by using three or more dispersoid forming elements, such as for example in alloy nos. 14-17 and 19-20. The liquidus temperature can especially be lowered by reducing the amount of vanadium. Using more different dispersoid forming elements appear to produce some synergistic effects on the yield strength. This can be seen when comparing alloy no. 11 with alloy no. 10 and alloy no. 14 with alloy no. 13.

Microstructure of Alloy No. 11: Al-7Mg-2Mn-1V-1Cr

[0175] The microstructure of this alloy is shown in FIGS. 5 to 8. The alloy was rapidly solidified and plastically consolidated by extrusion at a temperature of 400° C. Transmission electron microscopy (TEM) micrographs are showing a high number of dispersoids and a grain size below 1 μm. The dispersoids are mainly located in the grain boundary regions, nonetheless in locations where the supersaturation of the alloy with chemical components was higher dispersoids can also be seen inside the grain regions (FIG.

6). A higher magnification of the alloy microstructure shows that the size of the dispersoid is in the range below 50 nm (FIG. 7). Both the high number and the small size of the dispersoids promotes stabilization of the grains during forming processes. Due to the low diffraction contrast between adjoining grains in the TEM micrographs, the small grain size cannot be seen clearly. Transmission Kikuchi diffraction (TKD) methodology was used in order to extract more detailed information about the grain size.

[0176] The TKD technique is characterized by a high spatial resolution that is required for sub-micron materials and to provide correct statistical information about micro-structure features. A typical microstructure of the Al-7Mg-2Mn-1V-1C alloy acquired using TKD is shown in FIG. 8. The average grain size calculated in accordance with standard ASTM E2627-13 (2019) is 510 nm.

Examples Demonstrating Low Forming Stresses

[0177] The high strength of the alloys according to the disclosure at room temperature is coupled with a low flow stress in hot forming conditions. Exemplary results for the RS alloys are provided in Table 2 and compared to the one from the most common, medium strength, EN-AW 6082T6 alloy, which has been precipitation hardened by a T6-temperature treatment, and the technical purity aluminium EN-AW1050. Data were obtained from compression tests performed with a deformation speed of 10^{-2} s^{-1} . The flow stresses are the maximum flow stresses when compacting a sample of the alloy at the indicated temperature to 50% of its initial height.

[0178] The alloys with the high strength at room temperature shows very low flow stresses in hot forming conditions. Depending on the deformation temperature, the flow stresses are in the range of 10 to 1.5 MPa. At 525° C., the flow stress is only about 1% or even less of the yield strength of the alloy at 20° C. whereas the flow stress of the most common EN-AW 6082T6 alloy is about 5% of its yield strength at room temperature. Moreover, the yield strength of this alloy was determined after being precipitation hardened whilst the flow stress was measured before being precipitation hardened, i.e. when the alloy had a much lower yield strength (in the order of magnitude of 110 MPa). The flow stress was thus equal to about 15% of its yield strength without the T6 thermal treatment.

[0179] It is well known that solid solution strengthening is also present in high temperature forming conditions. The opposite effect for RS alloys that have a high concentration of dispersoid forming elements is a result of the grain boundary sliding phenomena, being active for sub-micron sized materials. In the article of T. Tokarski et al. (2012) described hereabove, a decrease of the flow strength was also achieved by the fine grain size of the alloy. The flow stress at 450° C. could be reduced from 48 MPa to 19 MPa, but the alloy had a lower strength, namely a yield strength at room temperature of only 320 MPa. The high strength aluminium alloy Al-7Mg-2Mn-1V-1Cr had, on the contrary, only a flow stress of about 16 MPa at 450° C., combined with a high yield strength of 624 MPa. Moreover, the high strength aluminium alloy according to the present disclosure has a higher thermal stability, and may moreover loose only a small part of its strength, so that it can be hot deformed at higher temperatures, thus enabling much lower hot forming stresses.

TABLE 2

True yield strength (YS, $R_{0.2}$ in MPa) at room temperature and flow stress (FS in MPa) at elevated temperatures of various alloys deformed at high temperatures					
	Al-7 Mg-2 Mn- 1 V-1 Cr	Al-7 Mg-2 Mn- 1 V-0.5 Cr	Al-7 Mg-2 Mn- 0.5 V-0.5 Cr	EN-AW 6082	EN-AW 1050
YS 20° C.	624	581	584	370 (T6)	60
FS 500° C.	10	8	7		
FS 525° C.	7	5	4	18	10
FS 550° C.	3.5	2	1.5	17	7

Examples of Ductile High Strength Alloys

[0180] The alloys according to the disclosure may not only have a high strength at room temperature but may also have a high elongation combined with a strength that is still high. Table 3 shows the composition and mechanical properties of six rapidly solidified and plastically consolidated alloys according to the present disclosure, including two alloys (alloys nos. 22 and 23) that have an elongation higher than 15% and four alloys (alloys nos. 24 to 27) that have a smaller elongation but a higher yield strength.

TABLE 3

Composition (in wt. %), mechanical strength (YS, $R_{0.2}$: true yield strength in MPa and UTS: ultimate tensile strength in MPa) and percent elongation at fracture (El. in %) of rapidly solidified and plastically consolidated alloys and of two solutionised and artificially aged (T6 treated) standard alloy compositions.

TABLE 3-continued

Composition (in wt. %), mechanical strength (YS, $R_{0.2}$: true yield strength in MPa and UTS: ultimate tensile strength in MPa) and percent elongation at fracture (El. in %) of rapidly solidified and plastically consolidated alloys and of two solutionised and artificially aged (T6 treated) standard alloy compositions.

	Alloy No.						EN-AW 6082 (T6)	EN-AW 6061 (T6)
	22	23	24	25	26	27		
Cr	0.1							
Zr	0.25	0.25	0.25	0.25	0.25	0.25		
Sum	0.85	0.75	0.75	0.75	0.75	0.75		
YS, $R_{0.2}$	350	375	460	490	510	530	370	330
UTS	470	480	555	560	570	585	400	350
El.	19	18	9	4	4	3	8	7.5

[0181] Alloys nos. 22 to 27 all have a yield strength that meets equation (1) and even equation (2), i.e. a yield strength in MPa that is higher than 525 or 530 minus ten times the percent elongation at fracture.

[0182] The heat treated standard alloys EN-AW 6082 and EN-AW 6061 have a similar yield strength as alloys nos. 22 and 23 but have a much lower elongation.

[0183] The alloys according to the present disclosure therefore do not only provide high to very high yield strengths but can even provide a combination of a high elongation with a high yield strength thus providing new opportunities in applications, such as automotive applications, wherein a low weight has to be combined with a high strength and a high ductility.

Examples Demonstrating Improved Thermal Stability

[0184] One of the unique feature of the alloys according to the present disclosure is the possibility to withstand higher forming temperature without significant loss of its mechanical properties. It allows to produce ready to use components in a hardened state. Examples of alloy hardness after hot forming processes is presented in Table 3. A hot forming process was simulated by means of a compression test, wherein the alloys plastically consolidated (extruded) at 400° C. where deformed to 50% of their initial height. The thermal stability tests was carried out in a temperature range of 450° C. to 550° C. at a deformation speed of 10^{-2}s^{-1} . An increase of the forming temperature to 50° C. above the extrusion temperature lead to approximately 5% loss of the initial hardness. As the temperature is increased, loss of hardness is higher; however, it does not exceed approx. 20% in the most extreme case of 550° C. (550° C. is the temperature just below the solidus temperature of the tested alloys). Depending on the desired final strength of the hot formed component a suitable temperature can be chosen, taking into account that an increased temperature will result in lower forces during the shaping process.

TABLE 3

Vickers hardness of the alloys after the hot forming processes at different temperatures.		
Al-7 Mg-2 Mn-1 V-0.5 Cr	Al-7 Mg-2 Mn-0.5 V-0.5 Cr	
As extruded at 400° C.	190	186

TABLE 3-continued

Vickers hardness of the alloys after the hot forming processes at different temperatures.

	Al-7 Mg-2 Mn-1 V-0.5 Cr	Al-7 Mg-2 Mn-0.5 V-0.5 Cr
Compr. 450° C.	181	178
Compr. 500° C.	170	164
Compr. 525° C.	168	156
Compr. 550° C.	150	143

1. An aluminium alloy which is rapidly solidified and plastically consolidated and which comprises aluminium (Al) and:

between 3.00 and 10.00 wt. % of magnesium (Mg);
more than 1.00 and less than 6.00 wt. % of manganese (Mn); and

one or more transition elements which are selected from the group consisting of chromium (Cr), vanadium (V), titanium (Ti), zirconium (Zr), molybdenum (Mo), cobalt (Co) and niobium (Nb), a total amount of the one or more transition elements being at least equal to 0.50 wt. % and the maximum amounts of the transition elements being 1.50 wt. % for Cr, 1.50 wt. % for V, 1.00 wt. % for Ti, 1.00 wt. % for Zr, 1.50 wt. % for Mo, 1.50 wt. % for Co and 1.00 wt. % for Nb,

wherein the aluminium alloy comprises:

no or less than 2.00 wt. %, preferably less than 1.00 wt. % of zinc (Zn), no or less than 2.00 wt. %, preferably less than 1.00 wt. % of silicon (Si), no or less than 1.00 wt. % of scandium (Sc) and no or less than 1.00 wt. % of tungsten (W);

no other elements different from Al, Mg, Mn, Cr, V, Ti, Zr, Mo, Co, Nb, Zn, Si, Sc and W or in total less than 4.00 wt. %, preferably less than 2.0 wt. % and more preferably less than 1.00 wt. % of the other elements; and aluminium as balance.

2. The aluminium alloy according to claim 1, which comprises at most 1.20 wt. %, preferably at most 1.00 wt. % and more preferably at most 0.75 wt. % of V.

3. The aluminium alloy according to claim **1**, which comprises at most 1.40 wt. %, preferably at most 1.30 wt. % and more preferably at most 1.20 wt. % of Cr.

4. The aluminium alloy according to claim **1**, which comprises at least 0.10 wt. %, preferably at least 0.20 wt. %, more preferably at least 0.30 wt. % and most preferably at least 0.40 wt. % of Cr.

5. The aluminium alloy according to claim **1**, which comprises at least 0.10 wt. %, preferably at least 0.20 wt. %, more preferably at least 0.30 wt. % and most preferably at least 0.40 wt. % of V.

6. The aluminium alloy according to claim **1**, which comprises two or more, preferably three or more of the transition elements.

7. The aluminium alloy according to claim **1**, which comprises Cr and V in a total amount of at least 0.50 wt. %, preferably of at least 0.70 wt. % and more preferably of at least 0.90 wt. %.

8. The aluminium alloy according to claim **1**, which comprises Cr and/or V, and additionally at least 0.15 wt. %, preferably at least 0.20 wt. % of Ti, at least 0.15 wt. %, preferably at least 0.20 wt. % of Zr, at least 0.15 wt. %, preferably at least 0.20 wt. % of Mo, at least 0.15 wt. %, preferably at least 0.20 wt. % of Co and/or at least 0.15 wt. %, preferably at least 0.20 wt. % of Nb.

9. The aluminium alloy according to claim **1**, wherein the total amount of the one or more transition elements is larger than 0.70 wt. %, preferably larger than 0.80 wt. %, more preferably larger than 0.90 wt. % and most preferably larger than 1.00 wt. %.

10. The aluminium alloy according to claim **1**, wherein the total amount of the one or more transition elements is smaller than 3.00 wt. %, preferably smaller than 2.75 wt. %, more preferably smaller than 2.50 wt. % and most preferably smaller than 2.25 wt. %.

11. The aluminium alloy according to claim **1**, which comprises at least 1.10 wt. %, preferably at least 1.20 wt. %, more preferably at least 1.30 wt. % and most preferably at least 1.40 wt. % of Mn.

12. The aluminium alloy according to claim **1**, which comprises less than 5.00 wt. %, preferably less than 4.50 wt. % and more preferably less than 4.00 wt. % of Mn.

13. The aluminium alloy according to claim **1**, which comprises at least 4.00 wt. %, preferably at least 4.50 wt. %, more preferably at least 5.00 wt. %, most preferably at least 5.50 wt. % and even more preferably at least 6.00 wt. % of Mg.

14. The aluminium alloy according to claim **1**, which comprises less than 9.00 wt. %, preferably less than 8.00 wt. % and more preferably less than 7.00 wt. % of Mg.

15. The aluminium alloy according to claim **1**, which has a true yield strength (YS, R_{0.2}), measured at 20° C. in accordance with ASTM E8/E8M-13a, of at least 400 MPa, preferably of at least 450 MPa, more preferably of at least 500 MPa and most preferably of at least 550 MPa.

16. The aluminium alloy according to claim **1**, which has a maximum true stress during plastic deformation in a compression test at a true strain rate of 10⁻² per second, of less than 25 MPa, preferably less than 21 MPa and more preferably less than 18 MPa at 450° C., and/or of less than 20 MPa, preferably less than 17 MPa and more preferably less than 14 MPa at 500° C., and/or of less than 17 MPa, preferably less than 14 MPa and more preferably less than 8 MPa at 550° C.

17. The aluminium alloy according to claim **1**, which has a true yield strength (YS, R_{0.2}), measured at 20° C. in accordance with ASTM E8/E8M-13a, and a maximum true stress during plastic deformation in a compression test to 50% of an initial height at a temperature of 525° C. and at a true strain rate of 10⁻² per second, which maximum true stress, expressed in MPa, is smaller than 3.0%, preferably smaller than 2.0% and more preferably smaller than 1.5% of the true yield strength expressed in MPa.

18. The aluminium alloy according to claim **1**, which comprises between 1.50 and 3.50 wt. %, preferably between 2.00 and 3.00 wt. % of Mn and between 3.00 and 5.00 wt. %, preferably between 3.00 and 4.50 wt. % of Mg, with the total amount of the one or more transition elements being less than 1.00 wt. %, and with Cr being preferably either absent or present in an amount of at most 0.3 wt. % or preferably of at most 0.2 wt. %.

19. The aluminium alloy according to claim **18**, which has a percent elongation at fracture (el) of at least 15%, preferably of at least 18%, and a true yield strength (YS, R_{0.2}), of at least 320 MPa, preferably of at least 340 MPa, both measured at 20° C. in accordance with ASTM E8/E8M-13a, the true yield strength being preferably further defined by the following equation:

$$\text{YS} > 525 - 10 \cdot el \quad (\text{equation 1})$$

20. The aluminium alloy according to claim **1**, which comprises no or less than 3.00 wt. % of Fe, preferably no or less than 2.00 wt. % of Fe, and more preferably no or less than 1.00% of Fe, and no or less than 2.00 wt. % of Cu, preferably no or less than 1.00 wt. % of Cu, and no or less than 0.40 wt. % of Ni, preferably no or less than 0.30 wt. % of Ni, and no or less than 0.40 wt. % of Bi, preferably no or less than 0.30 wt. % of Bi, and no or less than 0.40 wt. % of Sn, preferably no or less than 0.30 wt. % of Sn, and no or less than 0.40 wt. % of Pb, preferably no or less than 0.30 wt. % of Pb.

21. The aluminium alloy according to claim **1**, which has a liquidus temperature lower than 950° C., preferably lower than 900° C. and more preferably lower than 850° C.

22. The aluminium alloy according to claim **1**, which has, as plastically consolidated, an average grain size, measured in accordance with standard ASTM E2627-13 (2019), of less than 2000 nm, preferably of less than 1000 nm, more preferably of less than 800 nm and most preferably of less than 600 nm.

23. The aluminium alloy according to claim **1**, which is hot forged, in particular die forged, and which has an average grain size, measured in accordance with standard ASTM E2627-13 (2019), of less than 4000 nm, preferably of less than 2000 nm, more preferably of less than 1500 nm and most preferably of less than 1200 nm.

24. The aluminium alloy according to claim **1**, which is substantially free of primary intermetallic phases and preferably also of dendrites.

25. A method for producing an aluminium alloy according to claim **1**, which aluminium alloy has a liquidus and a solidus temperature and a predetermined difference between the liquidus and the solidus temperature, in which method a molten aluminium alloy composition is made having a composition as defined in any one of the claims **1** to **24**, wherein the molten aluminium alloy composition is rapidly solidified in a form of pieces of rapidly solidified material,

and wherein the pieces of the rapidly solidified material are plastically consolidated to produce the plastically consolidated aluminium alloy.

26. The method according to claim **25**, wherein the liquidus temperature is lower than 950° C., preferably lower than 900° C. and more preferably lower than 850° C.

27. The method according to claim **25**, wherein the molten aluminium alloy composition is ejected from at least one nozzle to be rapidly solidified, the molten aluminium alloy composition exiting the at least one nozzle having a predetermined temperature upon exiting the at least one nozzle and is rapidly solidified by being cooled down within a predetermined period of time to the solidus temperature of the aluminium alloy at an average cooling rate which is determined as a ratio of the difference between the predetermined temperature and the solidus temperature over the predetermined period of time and which is higher than 10 000° C./sec.

28. The method according to claim **27**, wherein the predetermined temperature is at least 75% of the liquidus/solidus temperature difference higher than the solidus temperature, the predetermined temperature being preferably equal to or higher than the liquidus temperature, and being more preferably at least 10° C. higher than the liquidus temperature.

29. The method according to claim **27**, wherein the molten aluminium alloy composition is supplied through a piping to the at least one nozzle and is additionally heated in the piping before being ejected out of the nozzle, the piping being preferably substantially completely filled with the aluminium alloy composition.

30. The method according to claim **25**, wherein the pieces of rapidly solidified material are plastically consolidated by

plastic deformation under pressure at a temperature of at least 300° C., preferably of at least 400° C. and more preferably of at least 450° C.

31. The method according to claim **25**, wherein the pieces of the rapidly solidified material are plastically consolidated at a temperature which is lower than the solidus temperature of the alloy, preferably at least 10° C. lower than the solidus temperature, but higher than 350° C., preferably higher than 375° C. and more preferably higher than 400° C., the temperature being most preferably comprised between 400° C. and 550° C.

32. The method according to claim **25**, wherein the pieces of rapidly solidified material are consolidated by plastic deformation to reduce an average grain size of the consolidated aluminium alloy, measured in accordance with standard ASTM E2627-13 (2019) to a value of less than 2000 nm, preferably less than 1000 nm, more preferably less than 800 nm and most preferably less than 600 nm.

33. The method according to claim **25**, wherein the pieces of the rapidly solidified material are consolidated by extruding them with a cross section reduction λ of at least 3, preferably of at least 6, more preferably of at least 8 and most preferably of at least 10.

34. The method according to claim **25**, wherein the consolidated aluminium alloy is hot forged, in particular die forged.

35. The method according to claim **34**, wherein the consolidated aluminium alloy is hot forged at a temperature higher than 400° C., preferably higher than 450° C. and more preferably higher than 475° C. or higher than 500° C., but lower than the solidus temperature of the alloy, preferably at least 10° C. lower than the solidus temperature.

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