Nov. 28, 1967 K. E. MANN 3,354,935

MANUFACTURE OF LIGHT-METAL CASTINGS

Filed April 7, 1964 3 Sheets-Sheet 1

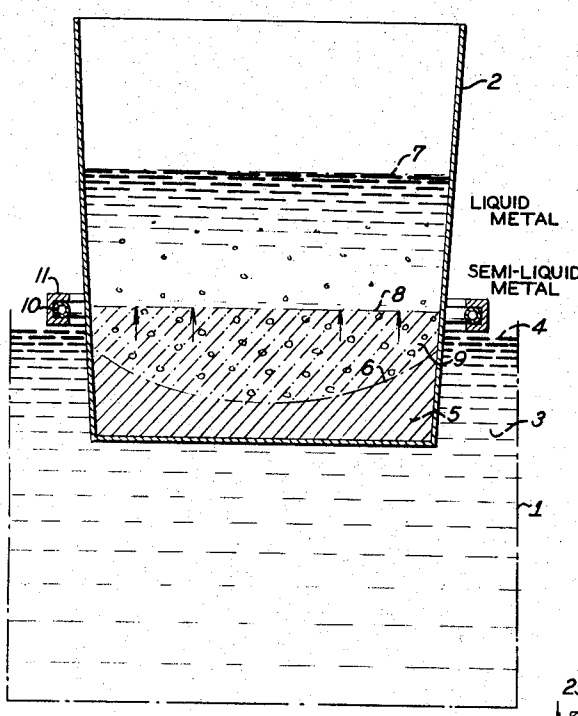

Fig. 1

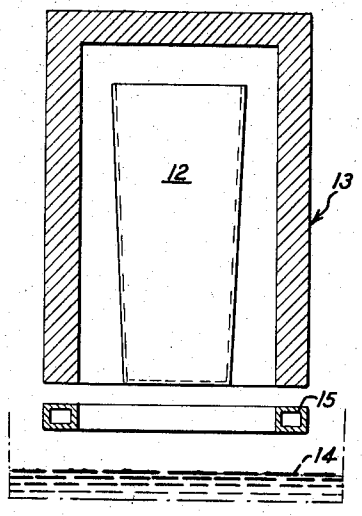

Fig. 3

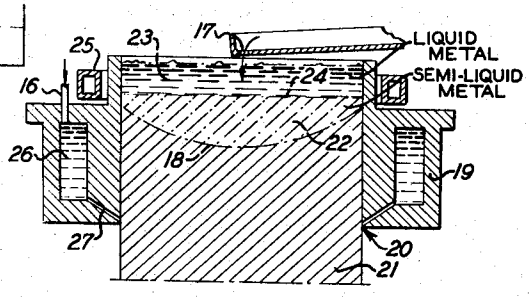

Fig. 10

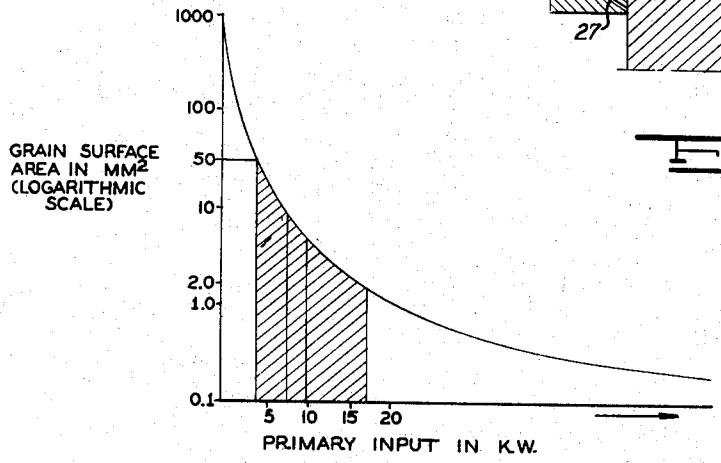

Fig. 2

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Nov. 28, 1967 K. E. MANN 3,354,935

MANUFACTURE OF LIGHT-METAL CASTINGS

Filed April 7, 1964 3 Sheets-Sheet 2

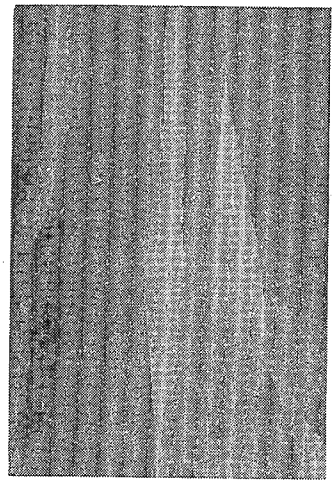

Fig.4

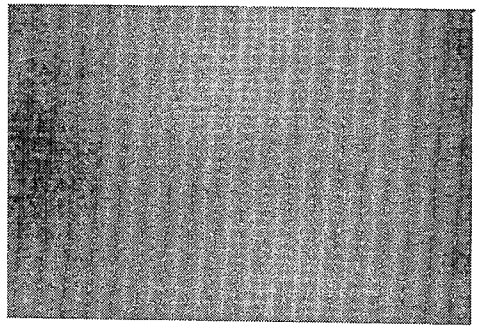

Fig.5

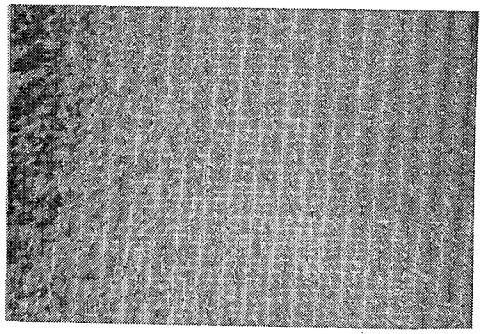

Fig.6

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MANUFACTURE OF LIGHT-METAL CASTINGS

Filed April 7, 1964 3 Sheets-Sheet 3

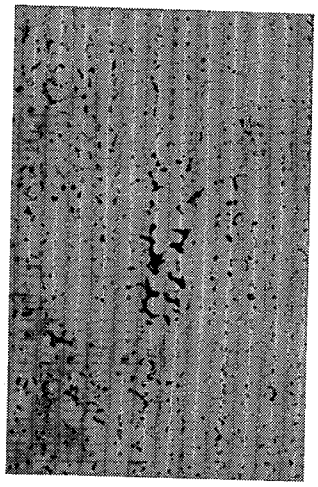

Fig.7

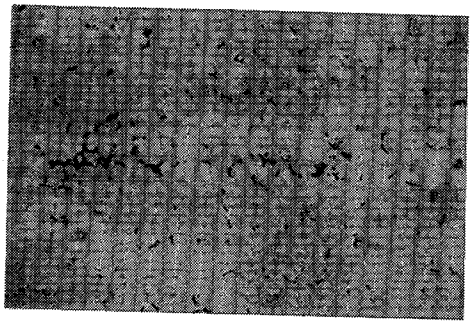

Fig.8

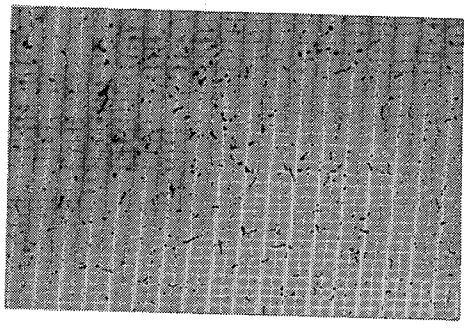

Fig.9

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United States Patent Office 3,354,935
Patented Nov. 28, 1967

3,354,935

MANUFACTURE OF LIGHT-METAL CASTINGS

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Filed Apr. 7, 1964, Ser. No. 358,047
Claims priority, application Germany, Apr. 13, 1963, F 39,487; Apr. 26, 1963, F 39,588
8 Claims. (Cl. 164—48)

The invention relates to improvements in the manufacture of castings of light metals and light-metal alloys, involving directional solidification, such as in tin-can casting and continuous casting. The preferred application of the invention is to the so-called "tin-can casting" and the same will be therefore principally described in connection with this casting method with "directional solidification," following called "directional casting."

The tin-can casting process developed to industrial maturity about 25 years ago has been described in Swiss Patent 221,650. A. Back and K. E. Mann have reported on its historical development, its theoretical basis, and its practice, in an article in the publication "Aluminium," vol. 29 (1953). Apparently there has been no further development beyond the state of the art described in that article, and alongside the highly developed continuous casting processes, the tin-can casting process has been unable to obtain any industrial significance to date, in spite of the good results it has achieved.

In this process the molten metal that is to be cast into ingots is placed in a thin-walled mold—the so-called "tin can"—whose wall thickness expediently amounts to from 1/100 to 1/30 of the diameter of the ingot to be produced, and by a relative movement between a coolant bath and the mold, it is gradually solidified from the bottom up, the metal in the mold above the surface of the coolant being kept in the fully molten state by heating means surrounding the mold. The mold is either lowered gradually into the coolant bath on a descending support or the level of the coolant bath is gradually raised relative to the stationary mold. In the first case, the furnace remains stationary while the mold moves downward, and in the second case it is elevated as the coolant level rises.

The invention as applied to "directional casting" is aimed at the production of metal ingots for shaping by non-cutting methods, especially of light metals and light-metal alloys, these ingots having high and directionally substantially independent, substantially uniform strength characteristics, especially elongations. These ingots are to serve especially for the production of such semifinished products as forgings with or without dies. In the conventional production of such semi-finished products by "directional casting," the same are deficient if the castings are deformed but slightly, or if the semifinished products have differential cross-sectional areas. This is due to the fact that insufficient hot forming of the cast structure takes place. For this reason such semifinished products exhibit unsatisfactory elongation values, particularly in the vertical direction to the direction of casting.

It is therefore an object of the invention to produce castings, and particularly tin-can cast ingots, having high directionally substantially independent, substantially uniform strength characteristics, especially elongations, which approach the characteristics of shaped forgings.

The foregoing and further objects of the invention will be apparent from the following description read in conjunction with the drawings in which:

FIG. 1 is a schematic representation of an arrangement for carrying out my invention as applied to tin-can casting;

FIG. 2 is a graph illustrating the dependency of grain structure of high strength aluminum alloys upon the power of stirring in the partially solid zone in accordance with the invention;

FIG. 3 is a schematic representation of a further arrangement for carrying out my invention as applied to tin-can casting;

FIGS. 4, 5 and 6 represent photomacrographs on a 1:1 scale showing respectively sectional cuts illustrating columnar and globulitic grain structure and grain structure in accordance with the invention;

FIGS. 7, 8 and 9 represent respectively photomicrographs at 50 magnification of parts of the photomacrographs of FIGS. 4, 5 and 6; and

FIG. 10 is a schematic representation of an arrangement for carrying out my invention to continuous casting.

Though it had been hitherto known that ingots produced by tin-can casting or the continuous casting method (both with directional solidification) exhibit different strength characteristics in the direction of casting and transverse thereto, persons skilled in the art have not concerned themselves with the causes of these differences.

American researchers have studied the influence of the cast structure on the mechanical properties of an aluminum alloy containing 4% copper and made of high purity aluminum. In their studies they used a sand mold clad with a steel jacket, which had a height of 15 cm. and an inside diameter of 8.6 cm.; the bottom was closed off with a copper disk with a diameter of 15 cm. The metal was protected against cooling by the air by means of radiant heating means placed over the mold during the solidification process. The solidification was brought about by spraying water against the bottom of the copper disk. At high casting temperatures, columnar crystals were obtained, and at low casting temperatures, fine-grained globulites were obtained. Strength tests on ingots with a pronounced columnar crystallization showed high strength factors, especially elongations parallel to the direction of the solidification, but very low values in a direction perpendicular thereto. The American researchers determined by microsections that these differing strength characteristics are caused by a linear arrangement of gas voids along the columnar crystals.

In my own experiments in which ingots were produced from an aluminum alloy of the AlZnMgCu type, with contents of chromium, iron and manganese, using the tin-can casting process, I was able to confirm the findings of the American researchers. It was further established, however, that in addition to the gas voids, aluminides are present in this case which were also disposed linearly along the columnar crystals, i.e., parallel to the direction of solidification. Insofar as these aluminides remain undissolved even after a homogenizing heat treatment, they bring about a reduction of the elongation values the same as gas voids.

Considerations of the art might have led one to expect that directionally different strength characteristics can be avoided by producing a fine globulite structure free of columnar crystals.

In my further experiments performed in this direction with the above-mentioned alloy, which further had a content of 0.1% titanium, tin-can cast ingots were produced having a fine-grained globulitic structure. Structural studies and strength tests, however, showed that the gas voids and aluminides were now oriented linearly perpendicular to the direction of solidification, and that the strength characteristics, especially the elongations, were better perpendicularly to the direction of solidification, and poorer parallel thereto.

Linearly arranged gas voids which develop during solidification, especially due to hydrogen released in molecular form, and linearly arranged aluminides, such as iron aluminide, which are still present in light metal alloys of technical purity, even after solution heat-

treatment, are therefore the cause of directionally unequal strength characteristics, especially elongations in the ingots in both these cases, the linear orientation of the texture in the one case being precisely 90° different from that of the other.

The invention is based on the discovery that the gas voids and the intermetallic compounds are rendered substantially harmless as regards the achievement of higher, directionally substantially independent and thus more uniform strength characteristics, if they are present in a relatively fine and random distribution. It has been found that such a fine and random distribution of the gas voids and intermetallic compounds in the cast ingots can be achieved by producing a cast metal structure intermediate columnar crystals and globulites. It has been further discovered that the desired intermediate structure can be obtained by causing the melt to contain, shortly before solidification, just so many crystallization nuclei that, on the one hand, columnar crystallization is suppressed while the formation of fine-grained globulites is prevented on the other.

Though casting temperatures and speeds influence the formation of nuclei or seedlings of crystallization, they are conventionally known and thus, for suitable results, predetermined in the art for any given light metal or light metal alloy to be "directionally" cast, such as by conventional tin-can casting or continuous casting. This leaves in the practical application of the invention the obtaining of the requisite number or range of nuclei or seedlings of crystallization in the area adjacent the solidus line.

As for instance schematically illustrated in FIG. 1, mold 2 and coolant tank 1 have been moved relative one another (by conventional means not shown) so that part of the mold is immersed in the coolant, such as water 3, below the level 4. Within mold 2 the solidified metal 5 has its solidus line 6. The liquid metal 7 in the upper part of the mold 2 is maintained at the desired temperature by conventional heating means (not shown) and has its liquidus line 8. The space between solidus line 6 and liquidus line 8 constitutes the pre-solid area 9 in which the dots indicate crystals acting as nuclei or seedlings of crystallization for the metal or metal alloy 7 to crystallize into solidification as the cooling level 4 progresses with respect to the liquid metal 7 with the continuing relative movement between mold and coolant. These seedlings may be constituted, at least in part, by an additive, as more fully set forth below. Alternatively, within the preferred embodiment of the invention, the crystallization nuclei or seedlings may be constituted by crystals of the light weight metal or alloy itself. It is the concentration of these crystals which controls the ultimate structure of the casting, i.e. whether a columnar or a globulitic structure is obtained on the one hand, or an intermediate essentially non-columnar or essentially non-globulitic structure, in accordance with the invention, on the other. If the nuclei of crystallization concentration in the pre-solid area, i.e., the area adjacent the solidus line, is too high, an essentially globulitic structure is formed, while if that concentration is too low, an essentially columnar structure results.

If the nuclei are to be furnished primarily by an additive, as hereafter more fully set forth, given concentrations of the additive will produce desired concentrations of nuclei of crystallization for the obtaining of the desired intermediate structure of the cast.

The preferred method of achieving the desired nuclei of crystallization concentration, however, and giving by far the best and most reliable results in the intermediate metal or metal alloy structure, is by way of controlling the number or range of crystals of the metal or metal alloy itself in the pre-solid area, i.e., the area adjacent the solidus line. This can be advantageously accomplished by distributing the crystals, such as by stirring, preferably in an upward direction, so that more or less of them (as may be required) are brought above the liquidus line. When then maintaining this distribution relatively constant, the concentration of nuclei of crystallization in the area adjacent the solidus line will remain substantially constant.

Since the nature of molten metal does not permit a direct determination or observation of a suitable concentration of nuclei of crystallization, it is necessary to establish the same empirically, i.e., by running a few test samples at different stirring intensities for a given metal or alloy and given casting conditions. If the cast test sample proves essentially columnar in nature, the stirring intensity must be increased, while if the test sample shows an essentially globulitic structure, the stirring intensity must be decreased. Increased stirring, and especially with a stirring movement or effect in an angular direction, and especially of an order substantially vertical to the normal liquidus line, will bring crystallization nuclei from the pre-solid area into the liquidus area, thereby reducing the concentration of such nuclei in the area adjacent the solidus line. On the other hand, decreasing such stirring will bring less crystallization nuclei into the liquidus area and thereby increase the concentration of crystallization nuclei in the presolidification area adjacent the solidus line. It is thus only necessary to establish empirically for given casting conditions and a specific alloy, that intensity of stirring which will give the desired intermediate structure for the casting.

In accordance with a preferred embodiment of the invention, the stirring is best accompli hed inductively. An arrangement for inductively stirring metal is disclosed in U.S. Patent No. 2,861,302. As there shown, a partially solidified, i.e., pre-solid portion of the metal is associated with a water-cooled, annular copper tubing surrounded by a secondary magnetic member formed of a plurality of horseshoe-shaped transformer laminations of magnetic material. The open horseshoe ends face, and are adjacent, the partially solidified metal area. When an alternating current is fed to the copper coil as the primary, so that a magnetic field is established thereabout, substantially all of the lines of flow of the magnetic field flow through the horseshoe-shaped elements of the secondary magnetic member. A substantial amount of this flow then passes from one end of the horseshoe through the partially solidified portion of the metal. Accordingly each of the laminations provides a substantially annular magnetic field having an axis substantially perpendicular to the direction of solidification of the molten material, and the aggregate of these alternating magnetic fields causes stirring motions in the partially solid area in planes substantially parallel to the annular axis of said fields. The object is to obtain temperature uniformity within the presolidification area, a decrease of the grain size and increase in the number of grains for the cast structure, and thereby a finely textured, i.e., globulitic casting.

Such an inductive stirring arrangement may be used for the obtaining of the desired intermediate, i.e., essentially non-columnar or essentially non-globulitic structure in accordance with the invention and so arranging the copper coil and secondary transformer laminations that the magnetic fields beyond the horseshoe ends pass partly through the partially solid area and partly through the liquidus area.

The annular copper coil and eventually a secondary member may be advantageously so placed that the magnetic field induced stirring passes at least through the upper part of the pre-solidification area. By appropriately adjusting the energy input into the primary, a greater or lower order of magnitude of alternating magnetic fields pass through adjacent portions of the pre-solidification and liquidus zones, and thus a greater or lower intensity of stirring effect results in bringing crystallization nuclei from the pre-solidification area into the liquidus zone.

An inductive stirring arrangement by the use of a device similar to that of said U.S. patent is schematically

illustrated in FIG. 1 as applied to tin-can casting, in accordance with the invention. As there shown the annular hollow (for water-cooling) copper tube **10**, connected to a source of alternating current (not shown) is surrounded by the laminated horseshoe-type magnetic laminated transformer sections **11**, facing with their open ends the mold **2**, so that the space therebetween overlaps the liquidus and pre-solidification zones. The current-induced alternating magnetic fields cause a stirring motion, bringing nuclei of crystallization from the pre-solidification zone into the liquidus zone, substantially in the direction of the arrows.

It is now only necessary to cast a few test samples prepared with different current energy inputs and select that current input or input range (for further operations) for which the test casting shows an essentially non-columnar, essentially non-globulitic structure, and preferably within the limits of average surface areas of grains shown on a sectional cut as below defined. This structure will have comparable elongation values, not only much higher than those of its corresponding globulitic structure, but also substantially uniform both parallel and transverse to the direction of casting, i.e., the direction of solidification of the metal or alloy.

The essentially non-columnar, essentially non-globulitic structure to be obtained in accordance with the invention, is for high strength aluminum alloys preferably one showing, on a sectional cut, grains, at least 80% and preferably substantially all of which have individual average surface areas on the order of about 2–50 mm.^{2} and preferably 7–9 mm.^{2}. Such a structure is for instance exemplified in FIG. 6 (Example 4) representing a photograph on a 1:1 scale of a sectional cut through an ingot, in accordance with the invention, polished and etched as is conventional for such cuts. The determination of the average grain size in terms of surface area is made in accordance with the procedure outlined in the 1952 Book of ASTM Standards, Part 2, "Non-Ferrous Metals," pp. 1176 to 1181.

The extent to which the specific intermediate structure differs in order of magnitude of average grain surface areas from columnar and globulitic structures is evident from the fact that columnar structures, as represented by the photographed section (1:1) of FIG. 4 (Example 2), have average grain surface areas of about 500–900 mm.^{2} while comparable globulitic surface areas, as represented by the photographed section (1:1) of FIG. 5 (Example 3), have average grain surface areas of about 0.2–04 mm.^{2}. Fifty times magnification of the cuts shown in FIGS. 4, 5 and 6 are represented by the microphotographs illustrated by FIGS. 7, 8 and 9. While FIGS. 7 and 8 clearly show the substantially linear arrangement of gas voids and intermetallic compounds for the columnar and globulitic structures respectively (one parallel and the other transverse to the direction of solidification as indicated by the arrows), FIG. 9 demonstrates the substantially uniform, random distribution of those voids and compounds throughout the structure in accordance with the invention.

The preferred materials to which the invention is applicable are all Al and Mg alloys conventionally identified in the art as high strength Al and Mg wrought alloys and having, for instance, proof stresses for the Al alloys of at least about 30 kg./mm.^{2} and for Mg alloys of at least about 20 kg./mm.^{2} (for illustrations cf. for example the latest editions of ASTM Standards on Non-Ferrous Metals and of the Metals Handbook). They include alloys with these proof stresses conventionally falling under the alloy types Al Zn Mg, Al Cu Mg, Mg Al Zn and Mg Zn Zr.

For each specific alloy combination, to be cast in accordance with the invention there exists, for the same casting conditions, a specific relationship between average surface area of the grains of the metal structure as above described, and the intensity of the stirring action induced to move nuclei of crystallization from the pre-solidification zone to the liquidus zone. This stirring intensity may be represented in terms of energy input, i.e., in case of an electrically induced stirring action, in terms of electric energy input into the primary circuit. This relationship between average surface area of grains and intensity of stirring action is typified for high strength aluminum alloys by the graph illustrated in FIG. 2, in which the logarithm of average grain surface area is plotted against energy input (kilowatt) to the primary. Of course, the values of energy input shown in FIG. 2 relate to a determined volume of metal, in this case that volume described in Example 2. The broad cross-hatched area shows the stirring intensity range within which grain surface areas of about 2–50 mm.^{2} surface area are obtained, while the narrower area within connotes that stirring range within which the preferred grain surface area of about 7–9 mm.^{2} results.

It is readily apparent from the curve of FIG. 2 that in order to determine the electric energy requirements for the obtaining of the desired grain structure in accordance with the invention, a first test sample cast at an arbitrary selected kw. input will show its corresponding average surface area on a sectional cut of the grain of the cast structure, and thus will give a first point for the specific curve. If this surface size is above about 50 mm.^{2}, a greater kw. input is required while if this surface size is less than about 2 mm.^{2} a lesser kw. input will be needed. Already the surface area grain size of the first test sample may permit a fair interpolation as to where the useable kw. range lies, then permitting in many cases the selection of a kw. within the desired range. Even if this is not the case, 3–4 test casts will ordinarily suffice to furnish a sufficient number of points to substantially define that kw. portion of the graph which is applicable to the grain structure in accordance with the invention.

As a practical matter kw. values of 5, 10, 15 and/or 20 will ordinarly suffice for this purpose. This is for instance exemplified by the graph of FIG. 2, which is specific to the alloy and casting conditions of Example 4, as set forth in the following example:

Example 1

Using the same alloy and the same casting conditions specified in the below Example 4, except for the use of an energy input into the primary of about 15 kw., a first test sample was cast. It shows an average grain surface between 1 and 2 (as indicated at *c*) which is already substantially non-globulitic and indicates by its relative closeness to the lower limit of 2 mm.^{2} that the 15 kw. are close to the useable kw. range, so that 10 kw. should bring it well within that range. Two further test samples are prepared in identical manner except that 5 and 10 kw. are used respectively. Their respective grain surface values (indicated at *a* and *b*) will now (together with the 1000 mm.^{2} point) furnish sufficient points for the curve to substantially define the useable and preferred kw. range. As is apparent, the 8 kw. energy input selected for Example 4 is about midway in the preferred area.

In one embodiment of the invention a seeding agent may provide the required crystallization nuclei. Such seeding agent may be, in known manner, a melt additive that has the effect of reducing grain size such as titanium, titanium carbide and boron carbide for aluminum alloys, and zirconium for aluminum alloys and magnesium alloys, or other known equivalent additives. While 0.1–0.15% by weight of titanium produce globulites and additive amounts below about 0.01% by weight produce columnar crystals, amounts between about 0.04 to 0.06% by weight of titanium may produce in many cases the desired intermediate structure.

The formation of nuclei in the melt can also be brought about by agitating it, in combination if desired, with the use of the above-mentioned additives, which reduce grain size. As to the knowledge of the inventor a very reliable

formation of the intermediate structure in accordance with the invention, however, is achieved advantageously by agitating the melt parallel to the direction of solidification, such as by causing alternating electromagnetic fields, especially induced currents, to work on the melt in the manner for instance set forth above or as hereafter described by way of alternative.

When tin-can casting, the heating applied to the molten metal in the mold above the level of the coolant can also be resistance heating in known manner. The electro-inductive method may be used in combination with resistance heating, if desired, so that by way of further alternative, the amount of vertical agitation of the melt and hence the amount of nucleus formation in the melt, can be thereby appropriately controlled.

Since gas voids and intermetallic compounds are known to be flaws which reduce the strength characteristics on the basis of a notch effect, the more numerous they are and the greater their individual dimensions are, the greater of course will be their effect.

Therefore, the more gas-free the melt is, the better are the results that can be achieved by the process of the invention. However, even though a charge has been carefully degassed in the melting furnace, it exhibits a poor vacuum test after the metal has been cast into the mold because it picks up gas while it is being cast, and gas is produced in it by the reaction of the metal with the parting agent. It may therefore be advantageous to subject the melt to another degassing treatment while in the mold, in a known manner, as for example by chlorination. Tests have shown that the vacuum test after chlorination is perfect, the ultrasonic picture of the ingot was substantially improved, and, in the microsection, the number of voids was smaller, which furthermore resulted in a definite improvement of the strength characteristics of the ingot.

The process of the invention, however, can be further improved by creating a vacuum or an overpressure of an inert gas, like argon, in the space above the surface of the metal in the mold during the progressive solidification of the metal. Since then nothing but a small number of very small gas voids, with a random orientation independent of direction, is produced by the procedure of the invention, optimum results are achieved as regards uniformly high strength characteristics in the ingot.

According to my copending application, Ser. No. 358,308, now U.S. Patent No. 3,268,963, the tin-can casting process can be performed in such a manner that the heating furnace or muffle is eliminated and no relative movement takes place between the electric heating units and the mold. The level of the coolant bath is gradually raised relative to the stationary mold and the stationary heating units since in this manner no raising or lowering apparatus is required for the mold and the heating units. The mold with the heating units surrounding it stands on a base in the coolant tank; the coolant level is raised by the controlled inflow of coolant above the level at the top of the base, where it stands before the beginning of the solidifying process.

The application of this procedure to the process of the invention is especially expedient when the solidification is performed, as described above, under a vacuum or under an inert gas under pressure, since the vacuum or pressure lines can be attached to a cover which is flange-mounted on the "tin can," since the position of the latter does not change during the solidification process.

The compartive tin-can casting tests described below will serve further to illustrate the invention. In these tests tin-can ingots were made, in conventional manner, in a tin can mold of about 3 mm. wall thickness, from an aluminum alloy containing 4.6% zinc, 3.3% magnesium, 0.65% copper, 0.32% iron, 0.30% manganese, 0.17% silicon and 0.18% chromium, and having a structure of columnar crystals in Example 2, of fine-grained globulites in Example 3, and of the intermediate structure of the invention in Example 4.

Example 2

An ingot is made of the above-mentioned aluminum alloy at a casting temperature of 660° C. and a rate of descent of 25 mm. per minute, and having an average diameter of about 145 mm. and a length of about 300 mm.; it has a cast structure that shows a pronounced columnar crystallization. FIG. 4 shows the macrostructure in a scale of 1:1. The arrow denotes the direction of casting.

From this ingot, rod-shaped test specimens were taken, which lay parallel to the direction of solidification in the one case, i.e., parallel to the columnar crystals, and perpendicular to the direction of solidification in the other. The test rods were made in a manner conventional for obtaining tensile strength (σ_B), yield strength ($\sigma_{0.2}$) and elongation ($\delta_{5\%}$) in a given direction and a direction transverse thereto, such as for instance published as German engineering norm DIN 50125 of April 1951. The specimens were subjected to a solution heat treatment (14 hrs. at 460° C. and then 14 hrs. at 480° C.), quenched in water, and subjected in one case to seven days of storage at room temperature, and in the other case they were hardened for 14 hours at 145° C. The strength characteristics for σ_B, $\sigma_{0.2}$, and $\delta_{5\%}$ of the test rods were made in conventional manner, such as for instance published as German engineering norm DIN 50146 of May 1951. The measured strength characteristics of these test specimens are shown in the following

TABLE 1

	Cold-hardened Test Specimens (to direction of solidification)		Hot-hardened Test Specimens (to direction of solidification)	
	Parallel	Perpendicular	Parallel	Perpendicular
σ_B(kp./mm.²)	42.2	30.5	46.3	43.6
$\sigma_{0.2}$(kp./mm.²)	29.3	25.4	42.7	43.2
$\delta_{5\%}$	16	3.5	6	1.5

Example 3

Under the same casting conditions as in Example 2, an ingot was made from the aluminum alloy of the same composition, but with an additional content of 0.1% titanium. This ingot has a fine-grained globulitic structure. FIG. 5 shows the macrostructure on a scale of 1:1, the arrow again indicating the direction of casting.

Following the same procedures outlined in Example 2, test specimens were taken from this ingot parallel and perpendicularly to the direction of casting, were subjected to heat treatments and measured for strength. The measured strength characteristics of these test specimens are set forth in the following

TABLE 2

	Cold-hardened Test Specimens (to direction of solidification)		Hot-hardened Test Specimens (to direction of solidification)	
	Parallel	Perpendicular	Parallel	Perpendicular
σ_B(kp./mm.²)	33.2	36.1	47.0	48.2
$\sigma_{0.2}$(kp./mm.²)	25.7	25.8	45.6	46.1
$\delta_{5\%}$	4	7	1	1.5

Example 4

Under the same casting conditions as in Example 2, an ingot was cast from the same (titanium-free) aluminum alloy, with inductive bath agitation during the solidification process.

In this case an alternating electro inductive stirring arrangement to that represented in FIG. 1 is used and is

schematically exemplified in FIG. 3. As there illustrated, the tin can mold **12** of sheet steel, having an inner mean diameter of about 145 mm. and a wall thickness of about 3 mm. and capable of producing an ingot of about 300 mm. length, is surrounded by a stationary heating muffle or furnace **13**, and may be gradually moved downwardly into progressive immersion below the coolant (such as water) level **14** by conventional, such as hydraulic means (not shown). A substantially rectangular hollow (for water cooling, not shown) copper tubing of 14 x 9 mm. outer and 8 x 3 mm. inner dimensions, is annularly bent to an outer ring diameter of about 210 mm. It is arranged relatively closely below the heating muffle or furnace **13** (about 6–10 mm.) and substantially co-axial with the tin can **12**. The copper tube **15** is mounted with its central horizontal plane about 30 mm. above the coolant level **14**. The copper tube **15** is connected in conventional manner to the secondary spool of a transformer (not shown). When alternating current is supplied to the primary of the transformer, the alternating current induced in the secondary creates alternating magnetic fields around the copper tube **15**, which, as the mold **12** moves gradually downwardly into the coolant, pass through the liquidus and pre-solidification zones in the manner hereinabove explained.

An ingot was cast using the same alloy and casting conditions specified in Example 2, in conjunction with the arrangement represented by FIG. 3 as above described and supplying to the transformer primary a current at 220 v. and 45 amperes. This represents an energy input of about 8 kw. and the current data for the secondary are approximately about 4 v. and 2000 amperes.

FIG. 6 shows on a 1:1 scale the macrostructure of such an ingot, which has average surface areas of grains between about 7 and 9 mm.^{2}, i.e., the structure of the invention. The arrow in FIG. 6 again indicates the direction of casting.

The test specimens taken from the ingot parallel and perpendicularly to the direction of castings were again subjected to the same heat treatments as described in Examples 2 and 3. The strength characteristics measured on these test specimens are shown in the following

TABLE 3

	Cold-hardened Test Specimens (to direction of solidification)		Hot-hardened Test Specimens (to direction of solidification)	
	Parallel	Perpendicular	Parallel	Perpendicular
σ_B(kp./mm.^{2})	37.9	35.8	47.0	46.0
$\sigma_{0.2}$(kp./mm.^{2})	23.8	27.1	42.8	39.0
δ_5%	14.5	15.5	6.5	9

When casting another ingot under the same conditions specified in this example but using a current input of about 20 kw. or higher, a grain structure approximating that of FIG. 5 is obtained, while when using a current input of about 2–3 kw., a grain structure approximately that of FIG. 4 results. This is also apparent from the graph on FIG. 2.

The following is apparent from Tables 1, 2 and 3: In the ingot with the pronounced columnar crystallization (Example 2, Table 1), elongation values are obtained, both in the cold-hardened and in the hot-hardened state, which were appreciably higher in the direction of solidification than perpendicularly thereto.

The microstructure in FIG. 7, which is a 50× enlargement, the arrow again indicating the direction of casting, shows the linear arrangement of the voids and of the intermetallic compounds in the direction of solidification.

In the fine-grained globulitic structure in FIG. 5, the ingots show higher elongation values perpendicularly to the casting direction than in the casting direction. The microstructure in FIG. 8, in which the arrow again indicates the solidification direction, shows the band-like arrangement of the voids in this fine-grained structure perpendicular to the direction of solidification.

The ingot with the intermediate structure of the invention shown in FIG. 6 has practically the same elongation values in the cold-hardened state in the casting direction and perpendicularly thereto. Also, in the hot-hardened state, the elongation values of these ingots are more uniform and substantially higher in both directions than the columnar crystal structure or in the fine-grained globulitic structure. The microstructure shown in 50× enlargement in FIG. 9, in which the arrow again indicates the direction of solidification, shows the random and fine distribution of the voids in an ingot with the intermediate structure according to the invention.

Example 5

Following the procedure outlined for Example 4 except that instead of the alloy there specified any of the high strength Al alloys set forth on p. 984 of the most recent edition of Metals Handbook may be substituted for similar results, i.e., the obtaining of cast structures having average grain surface areas in accordance with the invention and having randomly distributed gas voids and intermetallic compounds, as may be present. Particularly good results are obtained for the Al alloy there defined under the designation #7075.

Example 6

Following the procedure outlined for Example 4 except that there is substituted for the alloy there specified any of the high strength Al alloys recited on pp. 624 and 677 of the 1952 Book of ASTM Standards, part 2, Non-Ferrous Metals, satisfactory ingots with intermediate grain structures in accordance with the invention are obtained.

In general, following the procedure outlined for Example 4, satisfactory ingots with intermediate grain structures in accordance with the invention are obtained using an aluminum alloy composition as follows: MgO to 10%, ZnO to 8%, CuO to 6%, SiO to 5%, MnO to 2%, NiO to 2%, AgO to 1%, FeO to 1%, CrO to 0.5%, BeO to 0.1%, TiO to 0.05%, balance Al, the total amount of all constituents except aluminum not exceeding 12%.

Example 7

Proceeding in accordance with Example 4 except using for the Al alloy there specified a high strength Mg alloy, as for instance one conventionally designated as AZ80 containing 8.2% Al, 0.5% Zn, 0.2% Mn, balance Mg, good high strength ingots with intermediate grain structures in accordance with the invention are obtained.

Inasmuch as the rate of solidification in tin can casting and in continuous casting are comparable, the invention is also applicable to continuous casting.

Example 8

A conventional continuous casting arrangement is used as typified by that schematically illustrated in FIG. 10, in which the metal or alloy melt to be cast is continuously passed through inlet **17** into the interior of the mold **19**, cooled by the water reservoir **26** fed by inlet **16**, and passing cooling water at **20** through the channel **27** to flow along the continuously downwardly moving solidified casting **21**. Above the solidus line **18** is the partially solid zone **22** and above the same the liquidus zone **23**, with the liquidus line **24** therebetween. An annular copper tube **25** of about the same dimensions as given in Example 4 and similarly connected into a secondary transformer circuit, is provided at the upper part of the mold.

An ingot was cast by the just described procedure, using the alloy, casting temperature, electric current conditions and inner mold diameter as specified in Example 4. The resulting cast structure was about of the same order as to grain surface area and strength characteristics as that obtained in accordance with Example 4.

When applying the same continuous casting procedure just described to the alloys of Examples 5, 6 and 7, similar results are obtained.

By way of general observation, it should be stated in connection with the electro inductive stirring in accordance with the invention, that higher energy inputs into the primary are required in the direction of higher casting speeds or higher casting temperatures.

The best mode presently known to the inventor of carrying out the invention is that set forth in Example 4. This and the other examples are, however, to serve for purposes of illustration only and not of limitation.

I claim:

1. In the process of producing ingots of light metal and light metal alloys by casting such by directional solidification of a metallic melt; the improvement which comprises providing in at least the region adjacent both the liquidus and pre-solidus zones nuclei of crystallization in such concentration that at least 80 percent of the grain structure of the ingots formed possess an average grain surface area of about 2 to 50 mm.^{2}.

2. Improvement in accordance with claim 1 in which said ingot is that of a light metal alloy of the Mg and Al type, in which said nuclei of crystallization are obtained by way of a grain forming additive, in amount insufficient to yield an essentially globulitic structure.

3. Improvement in accordance with claim 1 in which there is established and maintained a liquidus zone for said metal, a partially solidified zone between said solidus and said liquidus zone, in which said concentration of nuclei of crystallization is maintained by a stirring action through at least a part of said partially solidified zone into and through at least the adjacent part of said liquidus zone.

4. Improvement according to claim 3 in which said stirring action is electromagnetic field induced, in which said casting process is of the group of the semi-continuous and continuous casting processes and in which said metal is Mg including its alloys.

5. Improvement according to claim 3 in which said stirring action is electromagnetic field induced, in which said casting process is of the group of the semi-continuous and continuous casting processes and in which said metal is Al including its alloys.

6. Improvement according to claim 5 in which said stirring action is maintained at an intensity for which at least 80% of the resulting grain structure of the casting essentially possess an average surface area of grain of substantially about 2–50 mm.^{2}.

7. In the tin-can casting of ingots from the group of high strength Al and Mg alloys in which there is substantially maintained a solidus zone, a liquidus zone, and a thereto intermediate pre-solidus zone, the improvement which comprises substantially subjecting, while casting, at least the adjacent portions of said liquidus and pre-solidus zones, to at least one electromagnetic field passing therethrough in a directional plane substantially parallel to the direction of casting and substantially maintaining the intensity of said field, such that at least 80% of the resulting grain structure of the ingots essentially possess an average surface area of grain of substantially about 2–50 mm.^{2}.

8. Improvement according to claim 7 in which said electromagnetic field is induced by a secondary transformer current and in which said intensity is maintained by a corresponding electric energy of a primary transformer current.

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